# TRANSACTIONS

## of the

## American Foundrymen's Society

Proceedings of the Sixty-Fifth
Annual Meeting, May 8-12, 1961

VOLUME 69

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# OFFICERS AND DIRECTORS

of the

# AMERICAN FOUNDRYMEN'S SOCIETY

(INCORPORATED)

*(Year ending May 11, 1962)*

*President,* ALBERT L. HUNT, Superior Foundry, Inc., Cleveland, Ohio.
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*(In Addition to the President and Vice-President)*

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*N. N. AMRHEIN, Federal Malleable Co., West Allis, Wis.
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*T. T. LLOYD, Albion Malleable Iron Co., Albion, Mich.
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*Technical Director,* S. C. MASSARI
*Secretary-Business Manager,* A. B. SINNETT
*Treasurer,* EDW. R. MAY

*Regional Vice-President

# PROCEEDINGS SUMMARY OF 65th ANNUAL MEETING

The 65th Castings Congress of the American Foundrymen's Society was held in San Francisco, May 8-12, 1961. Over 100 papers were presented at 50 Technical Sessions, which included 6 shop courses.

The Charles Edgar Hoyt Memorial Lecture was delivered by J. B. Caine, Cincinnati, Ohio, on the subject "Cast Metals and Cast Shapes" and is published starting on page 259 of this volume.

A summary of the sessions held follows:

**MONDAY, MAY 8**

**8:30 am — Authors-Chairmen Meeting**

**9:30 am — Light Metals Session**

*Presiding* — M. C. FLEMINGS, Massachusetts Institute of Technology, Cambridge.
H. ROSENTHAL, Frankford Arsenal, Philadelphia.

*Secretary* — S. LIPSON, Frankford Arsenal, Philadelphia.

*Meeting the Challenge of High Reliability Castings* — W. BABINGTON, Bell Telephone Laboratories, Murray Hill, N.J., E. E. MILLER, Western Electric Co., Greensboro, N.C., W. R. CLITHEROW, Western Electric Co., Winston-Salem, N.C.

*A Preliminary Evaluation of M517-T61 as a High Strength Aluminum Alloy Casting Material* — W. A. BAILEY, E. N. BOSSING, F. H. ROEBUCK, Douglas Aircraft Co., Santa Monica, Calif.

**9:30 am — Brass and Bronze Session**

*Presiding* — R. C. STOKES, Specialloy, Inc., Feasterville, Pa.
G. STEWART, East Bay Brass Foundry, Richmond, Calif.

*Secretary* — B. KINSMAN, Camsco Foundry, Inc., Azusa, Calif.

*Weld Repair of High Strength Yellow Brass Castings* — D. K. FOX, Westinghouse Electric Corp., East Pittsburgh, Pa.

*Compositions and Microstructures of Consistently Weldable As-Cast Nickel-Aluminum Bronze* — S. GOLDSPIEL, M. L. FOSTER, New York Naval Shipyard, Brooklyn, N.Y.

**9:30 am — Pattern Session**

*Presiding* — J. M. KREINER, National Malleable & Steel Casting Co., Cleveland.
R. L. OLSON, Dike-O-Seal Inc., Chicago.

*Secretary* — W. E. MASON, Westinghouse Air Brake Co., Wilmerding, Pa.

*Machined Patterns Can Be Competitive* — R. OLSON, Southern Precision Pattern Works, Inc., Birmingham, Ala.

*Low Melting Alloys for Pattern Fabrication and Other Uses in Pattern Shops* — O. J. SEEDS, Cerro Sales Corp., New York.

**2:00 pm — Sand Session**

*Presiding* — E. H. KING, The Hill & Griffith Co., Cincinnati.
C. A. SANDERS, American Colloid Co., Skokie, Ill.

*Secretary* — L. E. TAYLOR, Ottawa Silica Co., Ottawa, Ill.

*Preliminary Evaluation of Calcined Fluid Coke* — E. G. GENTRY, C. L. LEAR, Humble Oil & Refining Co., Detroit.

*Improvements in the $CO_2$ Process* — D. A. TAYLOR, British Cast Iron Research Association, Alvechurch, Birmingham, England.

*A New Approach to Testing the High Temperature Strength of Molding and Core Sands* — R. W. HEINE, Univ. of Wisconsin, Madison, J. S. SCHUMACHER, E. H. KING, The Hill and Griffith Co., Cincinnati.

**2:00 pm — Brass and Bronze Session**

*Presiding* — V. D. ALLEN, JR., American Smelting & Refining Co., Newark, N.J.
C. W. WARD, JR., American Standard Co., Cincinnati.

*Secretary* — R. B. FISCHER, Ingersoll-Rand Co., Phillipsburg, N.J.

*Hard Spots in Aluminum Bronzes* — N. A. BIRCH, Albion Malleable Iron Co., Albion, Mich.

*Electric Melting and Quality Control of Bronze Alloys* — R. V. BARONE, T. F. GODFREY, R. A. ROSENBERG, Walworth Co., S. Braintree, Mass.

**2:00 pm — Pattern Session**

*Presiding* — R. W. SMITH, Dike-O-Seal, Inc., Chicago.
M. K. YOUNG, U. S. Gypsum Co., Chicago.

*New Concept in Plastic Patternmaking (Polyurethanes)* — H. BURTON, Canadian Steel Foundries, Montreal, Canada, Presented by R. L. OLSON, Dike-O-Seal, Inc., Chicago.

Motion Picture: "Application of Polyurethanes for Patternmaking."

**2:00 pm — Heat Transfer Session**

*Presiding* — R. C. SHNAY, Canada Iron Foundries, Ltd., Toronto, Canada.
R. W. RUDDLE, Foseco, Inc., Cleveland.

*Factors Affecting Casting Soundness in Certain Alloys with Long and Short Freezing Ranges* — J. T. BERRY, T. WATMOUGH, Armour Research Foundation, Chicago.

*Thermal Properties of Mold Materials* — M. R. SESHADRI, A. RAMACHANDRAN, Indian Institute of Science, Bangalore, India.

*How Much of Your Superheat Is Being Lost in the Runner?* — J. W. HLINKA, V. PASCHKIS, F. S. PUHR, Columbia University, N.Y.

*Heat Transfer Committee Report* — W. K. BOCK, National Malleable & Steel Castings Co., Cleveland.

**4:00 pm — Brass and Bronze Session**

*Presiding* — J. E. GOTHERIDGE, Foseco, Inc., Cleveland.
D. C. CAUDRON, Pacific Brass Foundry of San Francisco.

*Secretary* — N. R. BARNETT, M. Greenberg's Sons, San Francisco.

*Properties and Microstructure of Cast Nickel-Containing Aluminum Bronze Alloys* — E. BELKIN, Westinghouse Electric Corp., East Pittsburgh, Pa.

*Effect of Chemical Composition on the Structure and Properties of Aluminum Bronzes Containing Manganese, Nickel and Iron* — J. O. EDWARDS, D. A. WHITTAKER, Department of Mines and Technical Surveys, Ottawa, Ont., Canada.

**4:00 pm — Light Metals Session**

*Presiding* — J. G. MEZOFF, The Dow Metal Products Co., Midland, Mich.
R. E. EDELMAN, Frankford Arsenal, Philadelphia.

*Secretary* — S. LIPSON, Frankford Arsenal, Philadelphia.

*The Effects of Centrifugal Force on Sub-Structure of Aluminum Alloy 356 + Be Permanent Mold Castings* — A. J. ILER, Northrop Corp., Norair Div., Hawthorne, Calif.

*Influence of Mold Variables on Fluidity of Aluminum; Doubling Casting Ductility by Fluidity Control* — M. C. FLEMINGS, F. R. MOLLARD, H. F. TAYLOR, Massachusetts Institute of Technology, Cambridge.

*Magnesium-Silver — Rare Earth — Zirconium Casting Alloys* — D. L. WHITEHEAD, Magnesium Elektron, Ltd., Manchester, England.

**4:00 pm — Safety, Hygiene and Air Pollution Session**

*Presiding* — R. L. McILVAINE, National Engineering Co., Chicago.
K. M. SMITH, Caterpillar Tractor Co., Peoria, Ill.

*Secretary* — H. J. WEBER, American Foundrymen's Society, Des Plaines, Ill.

*Foundry Air Pollution in the San Francisco Bay Area* — B. LINSKY, G. AASE, Bay Area Pollution Control District, San Francisco.

*Control of Dusts and Solvent Vapors from Shell Molding Operations* — E. B. HENBY, National Dust Collector Corp., Chicago.

Motion Picture: *It's Up to You* — National Society for the Prevention of Blindness, New York.

**4:00 pm — Sand Session**

*Presiding* — J. M. THOMAS, Lincoln Foundry Corp., Vernon, Calif.
L. E. TAYLOR, Ottawa Silica Co., Ottawa, Ill.

*Secretary* — E. J. PASSMAN, Frederic B. Stevens, Inc., Detroit.

*Granular Movement During Squeezing* — D. C. WILLIAMS, Ohio State University, Columbus, Ohio.

*A Systematic Approach to Sand Design and Control, Effect of Mulling* — G. J. VINGAS, A. H. ZRIMSEK, Magnet Cove Barium Corp., Arlington Heights, Ill.

**8:00 pm — Sand Shop Course**

*Presiding* — C. A. SANDERS, American Colloid Co., Skokie, Ill.

*The Most Important Green Sand Properties* — A. A. BELUSKO, Electric Steel Foundry Co., Portland, Ore., T. E. BARLOW, International Minerals & Chemical Co., Skokie, Ill., V. ROWELL, Archer-Daniels-Midland Co., Cleveland.

**8:00 pm — Pattern Shop Course**

*Presiding* — J. M. KREINER, National Malleable & Steel Castings Co., Cleveland.
R. L. OLSON, Dike-O-Seal, Inc., Chicago.

*The Application and Working Characteristics of Hard and Soft Plasters* — M. K. YOUNG, U. S. Gypsum Co., Chicago.

Motion Picture: *Industrial Tooling and Patternmaking with Plaster.*

**TUESDAY, MAY 9**

**8:30 am — Authors-Chairmen Meeting**

**9:30 am — Brass and Bronze Session**

*Presiding* — F. L. RIDDELL, H. Kramer & Co., Chicago.
D. L. HARRISON, Metro Metal Products, Inc., San Jose, Calif.

*Secretary* — R. A. COLTON, American Smelting & Refining Co., Houston, Tex.

*The Vertical Gating of Bronze Valves in Shell Molds* — F. E. MURPHY, G. J. JACKSON, R. A. ROSENBERG, Walworth Co., S. Braintree, Mass.

*Correlation of Physical Properties and Void Volume of Copper-Base Alloys with Morphology of Solidification* — Brass & Bronze Research Progress Report — R. A. FLINN, H. KUNSMANN, University of Michigan, Ann Arbor, Mich.

*A New High Strength Dezincification Resistant Bronze Alloy* — A. H. HESSE, R. Lavin & Sons, Chicago.

**9:30 am — Malleable Session**

*Presiding* — L. WININGS, Wagner Castings Co., Decatur, Ill.

*Secretary* — J. H. LANSING, Consultant, Shaker Heights, Ohio.

*Casting and Annealing Heavy White Iron Sections, Malleable Research Progress Report* — R. W. HEINE, T. W. MUELLER, J. W. WIDMEYER, University of Wisconsin, Madison, Wis.

*Shooter, Ceroxide Cermet or Slag* — C. A. SANDERS, American Colloid Co., Skokie, Ill., H. J. HEINE, Malleable Founders' Society, Cleveland, R. F. MARANDE, Ohio Malleable Div., Dayton Malleable Iron Co., Columbus, Ohio.

**9:30 am — Light Metals Session**

*Presiding* — H. ROSENTHAL, Frankford Arsenal, Philadelphia.
M. C. FLEMINGS, Massachusetts, Institute of Technology, Cambridge, Mass.

*Secretary* — N. DAVIDSON, Arwood Corp., Brooklyn, N.Y.

*Melting Casting and Heat Treatment Techniques for Structural Uranium Alloys* — G. D. CHANDLEY, D. G. FLECK, Watertown Arsenal, Watertown, Mass.

*Centrifugal Casting of Aluminum-Uranium Alloys* — N. E. DANIEL, E. L. FOSTER, R. F. DICKERSON, Battelle Memorial Institute, Columbus, Ohio.

*A New Type Light Weight Cellular Metal* — L. POLONSKY, S. LIPSON, H. MARKUS, Frankford Arsenal, Philadelphia.

**9:30 am — Pattern Session**

*Presiding* — R. L. OLSON, Dike-O-Seal, Inc., Chicago.
W. E. MASON, Westinghouse Air Brake Co., Wilmerding, Pa.

*Secretary* — R. OLSON, Southern Precision Pattern Works, Inc., Birmingham, Ala.

*Design and Engineering of Hot Core Boxes* — E. E. HARKESS, Cleveland Foundry, Ford Motor Co., Berea, Ohio.

*Aluminum Pattern Castings* — W. E. SICHA, Aluminum Co. of America, Cleveland.

**9:30 am — AFS Institute Trustees**

**12:00 Noon — Brass and Bronze Round Table Luncheon**

*Presiding* — R. B. FISCHER, Ingersoll-Rand Co., Phillipsburg, N.J.

*Problems of Running a Bronze Foundry in 1961*, L. H. DURDIN, President, Dixie Bronze Co., Birmingham, Ala.

**12:00 Noon — Pattern Round Table Luncheon**

*Presiding* — J. M. KREINER, National Malleable & Steel Casting Co., Cleveland.

*Judging Apprentice Contest Patterns*, G. E. GARVEY, Garvey Pattern & Mfg. Co., South Bend, Ind.

**12:00 Noon — AFS Board of Directors Luncheon and Business Meeting**

**2:00 pm — Gray Iron Session**

*Presiding* — W. W. HOLDEN, Pickands Mather Co., Chicago.
H. E. HENDERSON, Lynchburg Foundry Co., Lynchburg, Va.

*Secretary* — R. W. CARPENTER, Hanna Furnace Corp., Buffalo, N.Y.

*The Dwight Lloyd McWane Process for Making Foundry Iron* — T. E. BAN, C. D. THOMPSON, McDowell Co., Cleveland, B. W. WORTHINGTON, McWane Cast Iron Pipe Co., Birmingham, Ala.

*The Use of Calcium Carbide in Acid Cupolas* — R. SCHULZE, Knapsack-Griesheim A. G., Knapsack Bei Köln, West Germany.

**2:00 pm — Industrial Engineering and Cost Session**

*Presiding* — K. T. RINDERLE, Farrell-Cheek Steel Co., Sandusky, Ohio.
E. C. REID, Ford Motor Co. of Canada, Windsor, Ont., Canada.

*Report on Casting Survey, Industrial Engineering and Cost Committee* — J. A. WESTOVER, Westover Corp., Milwaukee.

Motion Picture: *Establishing Work Standards by Sampling* — Introduction by E. C. KEACHIE, University of California, Berkeley, Calif.

**2:00 pm — Light Metals Session**

*Presiding* — K. E. NELSON, Wellman Bronze & Aluminum Co., Bay City, Mich.
D. L. LAVELLE, American Smelting & Refining Co., South Plainfield, N.J.

*Secretary* — K. HERRICK, Arwood Corp., Tilton, N.H.

*Initial Bubble Test for the Determination of Hydrogen Content of Molten Aluminum Alloys* — D. J. NEIL, A. C. BURR, Aluminum Laboratories Ltd., Kingston, Ont., Canada.

*Supercooling Measurements and Grain Size Control in Light Alloys* — V. B. KURFMAN, The Dow Metal Products Co., Midland, Mich.

*Phosphorus Refinement of Hypereutectic Aluminum-Silicon Casting Alloys* — F. L. ARNOLD, J. S. PRESTLEY, Reynolds Metals Co., Richmond, Va.

**2:00 pm — Sand Session**

*Presiding* — F. S. BREWSTER, Brumley-Donaldson Co., Huntington Park, Calif.
T. E. BARLOW, International Minerals & Chemical Co., Skokie, Ill.

*Secretary* — W. A. SNYDER, University of Washington, Seattle.

*A Microscopic Study of the Bonding of Steel Foundry Sands by Clay* — J. B. CAINE, Consultant, E. H. KING, J. S. SCHUMACHER, Hill & Griffith Co., Cincinnati.

*A Systematic Approach to Sand Design and Control, Dry Properties of Clay Bonded Sand* — G. J. VINGAS, A. H. ZRIMSEK, Magnet Cove Barium Corp., Arlington Heights, Ill.

*Precision in the Foundry* — A. SHORT, Rolls-Royce Ltd., Derby, England. Officially Sponsored paper from the Institute of British Foundrymen, London.

**4:00 pm — Gray Iron Session**

*Presiding* — E. J. SMITH, JR., Semet-Solvay Div., Allied Chemical Corp., Cincinnati, Ohio.
D. E. KRAUSE, Gray Iron Research Institute, Columbus, Ohio.

*Influence of Inoculation on the Structure of Gray Iron at Various Equivalences* — Gray Iron Research Progress Report — H. D. MERCHANT, L. E. TORIELLO, J. F. WALLACE, Case Institute of Technology, Cleveland.

*Progress of Solidification in Inoculated and Uninoculated Gray Iron* — Gray Iron Research Progress Report — H. D. MERCHANT, J. F. WALLACE, Case Institute of Technology, Cleveland.

**4:00 pm — Light Metals Session**

*Presiding* — J. W. MEIER, Department of Mines and Technical Surveys, Ottawa, Ont., Canada.
W. E. SICHA, Aluminum Co. of America, Cleveland.

*Secretary* — A. B. DEROSS, Kaiser-Aluminum & Chemical Sales, Inc., Chicago.

*Premium Quality Magnesium Sand Castings* — K. E. NELSON, Wellman Bronze & Aluminum Co., Bay City, Mich.

*Progress in High Strength Alloy Castings for Airframes* — W. A. BAILEY, E. N. BOSSING, Douglas Aircraft Co., Inc., Santa Monica, Calif.

*Investment Cast Properties of Four Magnesium Base Alloys* — K. HERRICK, Arwood Corp., Tilton, N.H.

**4:00 pm — Malleable Session**

*Presiding* — G. B. MANNWEILER, The Eastern Co., Naugatuck, Conn.

*Secretary* — F. W. JACOBS, Texas Foundries, Inc., Lufkin, Texas.

*As-Cast Compact Graphite in Cast Iron* — Malleable Research Progress Report — R. W. HEINE, T. W. MUELLER, University of Wisconsin, Madison, Wis.

*Influence of Some Trace Elements on the Graphitization Kinetics of Malleable Irons* — G. SANDOZ, B. F. BROWN, Naval Research Laboratory, Washington, D.C., W. A. PENNINGTON, University of Maryland, College Park, Md.

**4:00 pm — Steel Session**

*Presiding* — A. J. KIESLER, General Electric Co., Research Laboratory, Schenectady, N.Y.
C. K. DONOHO, American Cast Iron Pipe Co., Birmingham, Ala.

*Secretary* — W. J. PARANA, Waukesha Foundry Co., Waukesha, Wis.

*Use of the Betatron for Radiographic Examination of Heavy Section Castings* — L. VENNE, Electric Steel Foundry Co., Portland, Ore.

*Radiography of Cast Nuclear Components by Means of Kilocurie Cobalt Sources* — R. LENT, Pacific Southern Foundries, Inc., Long Beach, Calif.

*Use of Direct Reading Spectrograph for Analyzing Castings* — E. MANG, Electric Steel Foundry Co., Portland, Ore.

**8:00 pm — Gray Iron Shop Course**

*Presiding* — H. H. WILDER, Vanadium Corp. of America, Chicago.
W. W. HOLDEN, Pickands Mather & Co., Cleveland.
J. L. LOWE, General Motors Institute, Flint, Mich.

*Materials and Chemical Control in Cupola Operation*, H. E. HENDERSON, Lynchburg Foundry Co., Lynchburg, Va.

**8:00 pm — Sand Shop Course**

*Presiding* — V. ROWELL, Archer-Daniels-Midland Co., Cleveland.

*Furfural Binders in the Foundry*, W. H. BUELL, Aristo Corp., Detroit.

**WEDNESDAY, MAY 10**

**8:30 am — Authors-Chairmen Meeting**

**9:30 am — AFS Annual Business Meeting**

*Presiding* — AFS President NORMAN J. DUNBECK.

President Dunbeck called upon General Manager Maloney who reported on the nomination of Officers and Directors for the coming year, and stated that no additional nominees had been received in accordance with the procedure prescribed in Art. XI of the Society By-Laws. He therefore cast a unanimous ballot of the membership of AFS for the election of the following:

*President (to serve one year):*
ALBERT L. HUNT, Superior Foundry, Inc., Cleveland.

*Vice-President (to serve one year):*
JOHN A. WAGNER, Wagner Castings Co., Decatur, Ill.

*Directors (to serve three years):*
D. E. BEST, Bethlehem Steel Co., Bethlehem, Pa.
R. J. HAGEBOECK, Frank Foundries Corp., Moline, Ill.
D. HALL, Oklahoma Steel Castings Co., Tulsa, Okla.
C. J. LONNEE, Alloyed Grairon Castings Corp., Ravenna, Mich.

C. OSSENFORT, Fairbanks, Morse & Co., Kansas City, Kans.
C. O. SCHOPP, Link Belt Co., Indianapolis.

*Director-at-Large (to serve three years):*
M. E. NEVINS, Wisconsin Centrifugal, Inc., Waukesha, Wis.

*Director (to serve one year):*
N. J. DUNBECK, International Mineral & Chemical Corp., Skokie, Ill.

The winners of the 1961 Robert F. Kennedy Apprentice Contest were:

**Wood Patternmaking**
1st — K. E. MCGARR, Poppenger & Howell Co., Akron, Ohio.
2nd — D. E. KUENZEL, Central Pattern Co., St. Louis.
3rd — A. E. LARSON, Progress Pattern Co., Detroit.

**Metal Patternmaking**
1st — J. S. ARNOLD, Progress Pattern Co., Detroit.
2nd — H. J. THIEL, Caterpillar Tractor Co., Peoria, Ill.
3rd — R. J. LAUDI, Aluminum Co. of America, Cleveland.

**Iron Molding**
1st — W. MULVEY, Brown & Sharpe Co., Providence, R.I.
2nd — J. V. BOGAERT, International Harvester Co., Indianapolis.
3rd — S. L. PURVIS, Rock Island Arsenal, Rock Island, Ill.

**Steel Molding**
1st — D. A. SPRAGUE, Columbia Steel Casting Co., Portland, Ore.
2nd — D. L. MALLDER, Rock Island Arsenal, Rock Island, Ill.
3rd — S. G. MUELLER, Electrocast Steel Foundry, Chicago.

**Non-Ferrous Molding**
1st — W. BELANGER, Bird Aluminum Foundry, Ltd., Vancouver, British Columbia.
2nd — W. MAXEY, Wisconsin Foundry Co., Manitowac, Wis.
3rd — E. D. LASECKI, Zandt Brass Foundry, Inc., Seattle, Wash.

President Dunbeck called upon Past President Nelson to present the AFS Service Citations and Awards of Scientific Merit.

*The AFS Service Citation*
J. TOTH, Harry W. Dietert Co., Detroit.
K. M. SMITH, Caterpillar Tractor Co., Peoria, Ill.
A. D. BARCZAK, Superior Foundry, Inc., Cleveland.

*The Award of Scientific Merit*
H. E. HENDERSON, Lynchburg Foundry Co., Lynchburg, Va.
H. J. WEBER, American Foundrymen's Society, Des Plaines, Ill.
T. R. SCHROEDER, General Motors Corp., Pontiac, Mich.

**10:30 am — Charles Edgar Hoyt Memorial Lecture**

*Cast Metals and Cast Shapes*, J. B. CAINE, Consultant, Cincinnati, Ohio.

**12:00 Noon — Joint Ductile, Gray and Malleable Iron Round Table Luncheon**

*Presiding* — R. A. CLARK, Union Carbide Metals Co., Cleveland.
E. A. WELANDER, John Deere Malleable Works, East Moline, Ill.

*The Foundry's Place in Our Future Economy*, B. L. SIMPSON, National Engineering Co., Chicago.

**12:00 Noon — Light Metals Round Table Luncheon**

*Presiding* — J. G. MEZOFF, The Dow Metal Products Co., Midland, Mich.

*The Manufacture and Use of Premium Quality Light Alloy Castings as High Integrity Structural Components*, PANEL: J. H. RIZLEY, Convair Div., General Dynamics Corp., Pomona, Calif., R. A. ZUECH, Rocketdyne Div., North American Aviation, Inc., Canoga Park, Calif., W. A. BAILEY, Douglas Aircraft Corp., Santa Monica, Calif., A. H. PETERSON, Lockheed Aircraft Corp., Sunnyvale, Calif. and T. E. PIPER, Hughes Aircraft Co., Fullerton, Calif.

**2:00 pm — Industrial Engineering and Cost Session**

*Presiding* — W. E. BOSWELL, Glamorgan Pipe & Foundry Co., Lynchburg, Va.
E. C. REID, Ford Motor Co. of Canada, Windsor, Ont., Canada.

*Cost Control Through Application of Process Control* — C. E. Haney, Esco Corp., Portland, Ore.

*Developing Standard Times for Use in the Job Foundry* — D. L. RAPHAEL, Pennsylvania State University, University Park, Pa.

**2:00 pm — Plant and Plant Equipment Session**

*Presiding* — J. THOMPSON, East Chicago, Ind.
H. W. JOHNSON, Wells Mfg. Co., Skokie, Ill.

*Maintenance Materials Inventory Control and Materials Disbursing Systems* — K. M. SMITH, Caterpillar Tractor Co., Peoria, Ill.

**2:00 pm — Steel Session**

*Steel Castings in the Aircraft Industry*

*Presiding* — D. N. ROSENBLATT, American Foundry & Machine Co., Salt Lake City, Utah.
G. A. COLLIGAN, United Aircraft Corp., East Hartford, Conn.

*Secretary* — A. J. KIESLER, General Electric Co. Research Laboratories, Schenectady, N.Y.

*Design Problems in Missile Castings* — R. G. BASSETT, Boeing Airplane Co.

*Challenge or Compromise with Aircraft — Missile Steel Castings* — S. A. MCCARTHY, McDonnell Aircraft Corp., St. Louis, Mo.

*Horizon Requirements for Castings* — S. R. CARPENTER, Convair, Div. of General Dynamics Corp., San Diego, Calif.

*High Strength Structural Steel Castings for Aerospace Applications* — W. R. ROSER, Northrop Corp., Norair Div., Hawthorne, Calif.

*Report on the AMC Steel Casting Program to Obtain High-Quality High-Strength Steel Castings for Aircraft and Missile Applications* — Y. J. ELIZONDO, Vought Aeronautics, Div. of Chance Vought Aircraft, Dallas, Texas.

**3:00 pm — Die Casting and Permanent Mold Session**

*Presiding* — D. L. COLWELL, Apex Smelting Co., Cleveland.
E. V. BLACKMUN, Aluminum Co., of America, Pittsburgh.

*Maximizing Soundness in Aluminum Die Castings* — L. M. ELIJAH, The George Sall Metals Co., Philadelphia.

*Fundamental Mechanical and Thermal Aspects of Die Casting* — A. B. DRAPER, Pennsylvania State University, University Park, Pa.

**4:00 pm — Steel Session**

*Presiding* — D. N. ROSENBLATT, American Foundry & Machine Co., Salt Lake City, Utah.
G. A. COLLIGAN, United Aircraft Corp., East Hartford, Conn.

*Secretary* — A. J. KIESLER, General Electric Co., Research Laboratories, Schenectady, N.Y.

*Press Forging of High Alloy Steel Castings* — T. WATMOUGH, J. T. BERRY, P. R. GOUWENS, Armour Research Foundation of Illinois Institute of Technology, Chicago.

*Development of Foundry Procedures for High Integrity Steel Castings* — R. J. ELY, J. E. FRIES, American Brake Shoe Co., Mahwah, N.J.

*Procurement of GSE Missile Castings* — L. W. SMITH, Symington-Wayne Corp., Depew, N.Y.

**4:00 pm — Gray Iron Session**

*Presiding* — J. N. WESSEL, Puget Sound Naval Shipyard, Bremerton, Wash.
W. C. JEFFERY, McWane Cast Iron Pipe Co., Birmingham, Ala.

*Secretary* — S. S. PHILLIPS, Ohio Ferro Alloys, South Bend, Ind.

*Some Factors Affecting the Chill Behavior of a Roll-Type Cast Iron* — R. W. LINDSAY, K. E. PINNOW, Pennsylvania State University, University Park, Pa.

*Development of Engine Castings by Experimental Stress Analysis* — L. A. GROTTO, International Harvester Co., Melrose Park, Ill.

**4:00 pm — Sand Session**

*Presiding* — D. C. CAUDRON, M. Greenberg & Sons, Berkeley, Calif.
J. B. CAINE, Consultant, Cincinnati.

*Secretary* — R. A. FULLER, Stanley Foundries, Huntington Park, Calif.

*Progress Report, Castings Defects Handbook Revision Committee (8-S), Sand Division* — T. E. BARLOW, International Minerals & Chemical Co., Skokie, Ill.

**7:00 pm — AFS Annual Banquet**

*Presiding* — AFS President NORMAN J. DUNBECK.

The AFS Annual Banquet was called to order by President Dunbeck. Presentation of AFS Gold Medal Awards was made as follows:

The Peter L. Simpson Gold Medal
Awarded to MERTON C. FLEMINGS, JR., Massachusetts Institute of Technology, Cambridge, Mass. . . . *"For outstanding contributions to the industry in the practical application of basic research to production of quality castings, particularly in the light metals field."*

The John A. Penton Gold Medal
Awarded to WILLIAM S. PELLINI, Naval Research Laboratories, Washington, D.C. . . . *"For exceptional contributions to the science of metal casting through leadership in fundamental foundry research, particularly in the field of heat transfer and flow of metals."*

Following the presentation of awards, President Dunbeck introduced the program of entertainment, consisting of three vocal quartets: The Californians, The Bay Town Four and The Jassabelles.

## THURSDAY, MAY 11

**8:30 am — Authors-Chairmen Meeting**

**9:30 am — Die Casting and Permanent Mold Session**

*Presiding* — R. T. MEYER, Kaiser Aluminum & Chemical Corp., Oakland, Calif.
F. C. BENNETT, The Dow Metal Products Co., Midland, Mich.

*Permanent Mold Design* — J. WALL, Permanent Mold Die Co., Hazel Park, Mich.

*Tilt Pouring Permanent Mold Castings* — G. W. STAHL, Stahl Specialty Co., Kingsville, Mo.

*Aluminum Furnace Refractories* — C. H. SCHWEINSBERG, J. L. DOLPH, Harbison-Walker Refractories Co., Pittsburgh, Pa.

**9:30 am — Ductile Iron Session**

*Presiding* — K. D. MILLIS, International Nickel Co., New York.
J. SCHUYTEN, Vulcan Foundry Co., Oakland, Calif.

*Secretary* — K. H. HAPGOOD, H. C. Macaulay Foundry Co., Berkeley, Calif.

*Reliability of Magnesium Analysis as a Criterion of Ductile Iron Quality* — Report of Ductile Iron Research Committee — A. H. RAUCH, Deere & Co., Moline, Ill., H. G. HAINES, Woodruff & Edwards, Inc., Elgin, Ill.

*Graphite Structures in Heavy Castings of Nickel Alloyed Austenitic Ductile Iron* — I. KARSAY, R. D. SCHELLENG, International Nickel Co., Bayonne, N.J.

**9:30 am — Fundamental Papers Session**

*Presiding* — J. F. WALLACE, Case Institute of Technology, Cleveland.
W. A. MATEJKA, Kelsey-Hayes Co., Utica, N.Y.

*Secretary* — F. R. MOLLARD, Massachusetts Institute of Technology, Cambridge, Mass.

*Structure in Rapidly Solidified Alloys* — P. E. BROWN, C. M. ADAMS, JR., Massachusetts Institute of Technology, Cambridge, Mass.

*Grain Refinement of Steel Castings and Weld Deposits* — G. K. TURNBULL, D. M. PATTON, G. W. FORM, J. F. WALLACE, Case Institute of Technology, Cleveland.

*Solidification Studies on Steel Castings and Ingots* — M. C. FLEMINGS, R. V. BARONE, S. Z. URAM, H. F. TAYLOR, Massachusetts Institute of Technology, Cambridge, Mass.

**9:30 am — Steel Session**

*Presiding* — A. J. KIESLER, General Electric Co., Research Laboratories, Schenectady, N.Y.
C. K. DONOHO, American Cast Iron Pipe Co., Birmingham, Ala.

*Secretary* — W. J. PARANA, Waukesha Foundry Co., Waukesha, Wis.

*Meeting the Challenge of Controlling Quality* — L. S. KRUEGER, Pelton Steel Casting Co., Milwaukee.

*Factors Affecting Metal-Mold Reactions* — G. A. COLLIGAN, United Aircraft Corp., East Hartford, Conn., L. H. VANVLACK, R. A. FLINN, University of Michigan, Ann Arbor, Mich.

*A Test Program for Improving the Performance of Ladle Refractories* — P J. NEFF, J. J. DOWNS, J. T. BAKER, American Steel Foundries, E. Chicago. Ind.

*Feeding Distance of Bars in Investment Molds* — H. PRESENT, Arwood Corp., Brooklyn, N.Y., H. ROSENTHAL, Frankford Arsenal, Philadelphia.

**12:00 Noon — Die Casting and Permanent Mold Round Table Luncheon**

*Presiding* — F. C. BENNETT, The Dow Metal Products Co., Midland, Mich.

*Metals in Space*, J. C. MCDONALD, Lockheed Missile & Space Div., Lockheed Aircraft Corp., Sunnyvale, Calif.

**12:00 Noon — Steel Round Table Luncheon**

*Presiding* — D. N. ROSENBLATT, American Foundry & Machine Co., Salt Lake City.
J. A. RASSENFOSS, American Steel Foundries, E. Chicago, Ind.

Steel Research Progress Reports:

*Sources and Prevention of Non-Metallic Macro-inclusions in Cast Steel* — R. A. FLINN, W. B. PIERCE, L. H. VANVLACK, University of Michigan, Ann Arbor, Mich.

*Evaluation of Refractories as Sources of Macro-inclusions* — L. H. VANVLACK, J. E. BROKLOFF, R. A. FLINN, University of Michigan.

**12:00 Noon — Past Presidents' Luncheon**

*Presiding* — L. H. DURDIN, Dixie Bronze Co., Birmingham, Ala.

**2:00 pm — Sand Session**

*Presiding* — G. S. SCHALLER, University of Washington, Seattle.
V. ROWELL, Archer-Daniels Midland Co., Cleveland.

*Secretary* — B. H. BOOTH, Carpenter Bros., Inc., Milwaukee.

*Core Production in "Hot Boxes" at the Renault Works* — P. JASSON, Renault, Billancourt Seine, Paris, France.

*Jolt Impact as a Measurement of Sand Toughness* — T. E. BARLOW, International Minerals & Chemical Co., Skokie, Ill., H. W. DIETERT, H. W. Dietert Co., Detroit.

*The Effect of Median Grain Size on the Properties of Green Sand* — A. B. DRAPER, Pennsylvania State University, University Park, Pa., H. A. KNAPPENBERGER, North Carolina State University, Raleigh, N.C.

**2:00 pm — Symposium on Vacuum Melting and Casting, Jointly Sponsored by Steel Division and Fundamental Papers Committee**

*Presiding* — M. C. FLEMINGS, Massachusetts Institute of Technology, Cambridge, Mass.
A. J. KIESLER, General Electric Co., Research Laboratory, Schenectady, N.Y.

*Secretary* — W. KOPPI, International Nickel Co., New York.

*Vacuum Induction Melting and Casting* — W. A. MATEJKA, Kelsey Hayes Co., Utica, N.Y.

*Specialty Melting as a Service Function in a Research Laboratory* — W. F. MOORE, General Electric Co. Research Laboratory, Schenectady, N.Y.

*Quality Requirements of Alloys for Use in Vacuum Induction Melting* — T. F. KAVENEY, Union Carbide Metals Co., Cleveland.

*Effect of Sulphur and Carbon on Tensile Properties of Vacuum Cast High Strength Steel* — J. B. DABNEY, American Brake Shoe Co., Mahwah, N.J., M. C. FLEMINGS, H. F. TAYLOR, Massachusetts Institute of Technology, Cambridge, Mass.

*300 Kilowatt Induction Melting and Degassing Installation* — O. E. ERICKSON, Hanford Foundry Co., San Bernardino, Calif.

*Some Theoretical and Practical Considerations in Vacuum Induction Melting Uranium Alloys* — P. S. SCHAFFER, Watertown Arsenal, Watertown, Mass.

*Ladle Degassing for Foundry Use* — H. S. PHILBRICK, John Mohr & Sons, Chicago.

**3:00 pm — Die Casting and Permanent Mold Session**

*Presiding* — H. J. HEATH, Aluminum Co. of America, Vernon, Calif.
R. P. DUNN, Lindberg Engineering Co., Chicago.

*Inclusions in Aluminum Alloy Castings* — J. M. FOX, Aluminum Co. of America, Cleveland.

*Minor Elements in Aluminum Alloys for Die & Permanent Mold Casting* — D. L. COLWELL, R. J. KISSLING, Apex Smelting Co., Cleveland.

*Iron: The Problematic Factor in the Quality of Aluminum Alloy Die Castings* — W. BONSACK, Aluminum & Magnesium Inc., Sandusky, Ohio.

**4:00 pm — Ductile Iron Session**

*Presiding* — F. F. BRENDLER, Stanley Foundries, Inc., Huntington Park, Calif.
H. W. RUF, Grede Foundries, Inc., Milwaukee.

*Secretary* — J. R. LUDWIG, Soundcast Co., Newport Beach, Calif.

*Classification of Graphite Forms in Nodular or Ductile Irons* — C. K. DONOHO, American Cast Iron Pipe Co., Birmingham, Ala.

*Effect of Chemical Composition Upon Graphite Structures in Heavy Ductile Iron Castings* — I. KARSAY, R. D. SCHELLENG, International Nickel Co., Bayonne, N.J.

**6:00 pm — AFS Alumni Dinner**

*Presiding* — C. E. NELSON, The Dow Metal Products Co., Midland, Mich.

**8:00 pm — Ductile Iron Shop Course**

*Presiding* — E. A. WELANDER, John Deere Malleable Works, East Moline, Ill.

*Raw Material, Melting and Controls for the Production of Ductile Iron* — H. W. RUF, Grede Foundries, Inc., Milwaukee, Wis.

**8:00 pm — Gray Iron Shop Course**

*Presiding* — W. W. HOLDEN, Pickands Mather & Co., Cleveland.
J. L. LOWE, General Motors Institute, Flint, Mich.
H. E. BARNUM, Vanadium Corp. of America, Detroit.

*How to Make Different Irons from a Single Cupola Heat* — H. H. WILDER, Vanadium Corp. of America, Chicago.

## FRIDAY, MAY 12

**8:30 am — Authors-Chairmen Meeting**

**8:30 am — Registration Opens**

**9:30 am — Ductile Iron Session**

*Presiding* — K. D. MILLIS, International Nickel Co., New York.
E. A. WELANDER, John Deere Malleable Works, E. Moline, Ill.

*Secretary* — D. MATTER, Ohio Ferro Alloys Corp., Canton, Ohio.

*The Formation of Flake Spheroidal, Film and Compact Graphite During Solidification of Cast Iron* — C. R. LOPER, JR., R. W. HEINE, University of Wisconsin, Madison, Wis.

*Post Inoculation of Ductile Iron* — W. C. JEFFERY, McWane Cast Iron Pipe Co., Birmingham, Ala., H. D. BRADSHAW, University of Alabama, University, Ala.

**9:30 am — Die Casting and Permanent Mold Session**

*Presiding* — A. B. DEROSS, Kaiser Aluminum & Chemical Sales, Chicago.
A. W. SIMPSON, III, Western Die Casting Co., Emeryville, Calif.

*Die Casting Trim Die Procedure* — D. HANNAH, Hamco Products, Warren, Mich.

*A Study of the Nature of Porosity in Zinc Die Castings* — G. P. LEWIS, D. A. CRAW, R. C. BELL, Consolidated Mining & Smelting Co. of Canada, Ltd., Trail, B.C.

**9:30 am — Fundamental Papers Session**

*Presiding* — P. J. AHEARN, Watertown Arsenal, Watertown, Mass.
W. A. MATEJKA, Kelsey Hayes Co., Utica, N.Y.

*Secretary* — F. R. MOLLARD, Massachusetts Institute of Technology, Cambridge, Mass.

*An Experimental and Quantitative Evaluation of the Fluidity of Aluminum Alloys* — M. C. FLEMINGS, E. NIIYAMA, H. F. TAYLOR, Massachusetts Institute of Technology, Cambridge, Mass.

*Heat Extraction from External Corners of Metal Castings* — K. H. COATS, University of Michigan, Ann Arbor, Mich.

**9:30 am — Sand Session**

*Presiding* — M. E. GINTY, Vulcan Foundry Co., Oakland, Calif.
B. FARROW, Barker Foundry Supply Co., Los Angeles, Calif.

*Secretary* — O. H. ROSENTRETER, O. H. Rosentreter Co., Bell Gardens, Calif.

*A Systematic Approach to Sand Design and Control, Wood Flour* — A. H. ZRIMSEK, G. J. VINGAS, Magnet Cove Barium Corp., Arlington Heights, Ill.

*A Systematic Approach to Sand Design — Interpretation of Data* — A. H. ZRIMSEK, G. J. VINGAS, Magnet Cove Barium Corp., Arlington Heights, Ill.

*The Carbon Dioxide Process* — G. D. HALEY, Waimet Alloys Co., Detroit, J. L. LEACH, University of Illinois, Urbana.

**9:30 am — Steel Session**

*Presiding* — G. A. COLLIGAN, United Aircraft Corp., E. Hartford, Conn.
W. J. PARANA, Waukesha Foundry Co., Waukesha, Wis.

*Secretary* — W. A. KOPPI, International Nickel Co., New York.

*Improve Fatigue Life of Steel Castings by Controlled Gas Atmosphere Heat Treatment* — R. D. BLOCK, Barber Coleman Co., Rockford, Ill.

*Rapid Gas Heating Techniques for Centrifugally Cast Steel Cylinders* — A. AYVAZIAN, Rodman Laboratory, Watertown Arsenal, Watertown, Mass.

*High Alumina Roofs for Electric Arc Furnaces* — R. E. GRAY, Texas Steel Co., Fort Worth, Texas.

*Rapid Desulfurization to 0.002% Sulfur* — E. J. DUNN, JR., Crucible Steel Co., Pittsburgh, Pa.

*A Flux-Cored $CO_2$ Welding Process for Repairing Steel Castings* — M. B. PRISUTA, National Cylinder Gas, Div. Chemetron Corp., Chicago.

**12:00 Noon — 65th AFS CASTINGS CONGRESS Adjourned**

## MONDAY, MAY 15

**6:00 pm — Closing Session**

*Presiding* — AFS President NORMAN J. DUNBECK.

*The Industrial Future of Hawaii* — GEORGE MASON, Director, Department of Economic Development, State of Hawaii.

# ANNUAL REPORT OF AFS GENERAL MANAGER

*(Fiscal Year 1960-1961)*

This report covers overall Administrative and Policy matters and is supported by separate reports of the Secretary, the Treasurer, and the Technical Director.

## SURVEY OF YEAR

The fiscal year 1960-61, which started out to be an Exhibit year, turned out as a non-Exhibit year with a substantial loss. While not as large a loss as in 1958-59 ($187,900), the loss of $149,800 for the past year nevertheless pointed up sharply the choice between increased revenues and decreased expenses (activities).

It should be recalled that the decision to hold an Exhibit in San Francisco was based chiefly on these considerations: (1) the decision to meet in San Francisco had been made two years earlier and announced; (2) no AFS Convention had ever been held west of St. Louis; (3) no exclusively foundry Exhibit had ever been held on the West Coast; (4) no "policy" for holding an annual AFS Show was involved or considered; and (5) a 1961 Exhibit was believed sound for these reasons and essential to avoid another large financial loss as in 1958-59.

Cancellation of the 1961 Exhibit, followed by one of the smallest attendances at any previous nonexhibit Convention, clearly demonstrates that future annual meetings must be held where greatest participation can be expected. Furthermore, the severe budget cuts made last January, as essential as they were, are damaging to the Society's long-standing image of good management.

The following table shows the extent of mid-year budget revisions and how the year 1960-61 turned out.

| | Income | Expense | Excess Expense |
|---|---|---|---|
| Original Budget | $851,200. | $928,700. | $ 77,500. |
| Revised Budget | 671,700. | 820,000. | 148,300. |
| Incr. or (Decr.) | (179,500.) | (108,700.) | 70,800. |
| Actual for Year | $688,200. | $838,000. | $149,800. |

The net result of 1960-61 finances means, of course, that the "austerity" program necessarily adopted for 1959-60 and continued more drastically in 1960-61, must now be a guiding principle for at least the 2-year period 1961-63. Society officers will naturally be anxious to restore the Investment Trust Account (depleted by $150,000 in March 1961) through conservative Expense budgets and increased Income so as to produce a substantial Net Income by June 30, 1963. Recommendations of the Finance Committee for 1961-62 and 1962-63 budgets are to that end.

## "MODERN CASTINGS"

It should be noted that the magazine MODERN CASTINGS continues to show more professionalism in content, readership and acceptance. It can no longer be classed as strictly a "house organ" since it now is definitely a first-class, competitive, commercial publication.

To do this entails certain obligations, decisions and policy adjustments, not all of which will be understood or accepted by the rank-and-file membership. To produce "x" number of copies of an issue containing "y" number of pages will result in "z" cost. Adding more pages for Chapter news, technical committee reports and various AFS activities (e.g., Conventions and Exhibits, T&RI, Publications ads, reports on official meetings important to AFS) simply means less pages for other material important to maintaining a "professional image."

In other words, AFS cannot have in MODERN CASTINGS all that is essential to a full-fledged "house organ," and at the same time build a commercially competitive magazine required to produce maximum net income through greater advertising acceptance. Within expense budgets approved, the Managing Director of MODERN CASTINGS now has the right to adjust the magazine's appearance, format, content and coverage to conform to his proved experience as to what a professional and commercially competitive book should be . . . without, of course, violating basic AFS policy.

MODERN CASTINGS (and its predecessor AMERICAN FOUNDRYMEN) appeared as vol. 1, no. 1 in January 1938 as strictly a house organ . . . a "means of communication between the National organization and the Chapters." Advertising was not accepted until May 1945. Since then the magazine has gone through three distinct phases of development, and now is in the fourth (and, we trust, final) phase of growth. These 4 phases may be described thus:

*1945-1951* — Advertising "acceptance" based initially on novelty appeal, then on "support" of AFS, except for 1946 pre-Convention issue (Golden Anniversary year). Considerable house organ appeal to members, circulation limited to members.

*1951-1954* — Accepted as permanent, advertising materially increased during "good times" by intensive selling effort, largely by Staff personnel. Pre-Convention issue of 1952 (International Congress year) outstanding, caliber of articles much improved, house organ features continued. Audit of circulation begun.

*1954-1959* — Period of hard sell against aroused competition, ground substantially lost by professional comparisons. "Con-

# Breakdown of AFS Membership

## As of June 30, 1961

| Chapter | Sustaining | Company | Personal | Affiliate | Associate | Junior | Honorary | Life Serv. | Total |
|---|---|---|---|---|---|---|---|---|---|
| Birmingham | 9 | 15 | 117 | 248 | 6 | 9 | 7 | 0 | 411 |
| British Columbia | 0 | 5 | 62 | 9 | 7 | 6 | 0 | 0 | 89 |
| Canton District | 0 | 16 | 60 | 105 | 2 | 2 | 0 | 0 | 185 |
| Central Illinois | 2 | 7 | 39 | 170 | 4 | 13 | 2 | 0 | 237 |
| Central Indiana | 3 | 24 | 59 | 248 | 5 | 10 | 1 | 0 | 350 |
| Central Michigan | 1 | 6 | 48 | 91 | 11 | 11 | 2 | 0 | 170 |
| Central New York | 0 | 17 | 57 | 146 | 2 | 1 | 0 | 0 | 223 |
| Central Ohio | 2 | 20 | 60 | 124 | 7 | 5 | 4 | 0 | 222 |
| Chesapeake | 2 | 13 | 55 | 37 | 20 | 0 | 4 | 1 | 132 |
| Chicago | 17 | 54 | 175 | 374 | 10 | 70 | 8 | 5 | 713 |
| Cincinnati District | 7 | 34 | 94 | 145 | 8 | 14 | 3 | 0 | 305 |
| Connecticut | 3 | 10 | 73 | 38 | 4 | 1 | 0 | 0 | 129 |
| Corn Belt | 0 | 3 | 10 | 23 | 2 | 1 | 0 | 0 | 39 |
| Detroit | 7 | 34 | 151 | 312 | 13 | 16 | 9 | 3 | 545 |
| Eastern Canada | 5 | 23 | 101 | 154 | 19 | 11 | 1 | 0 | 314 |
| Eastern New York | 1 | 6 | 18 | 42 | 4 | 1 | 0 | 0 | 72 |
| Metropolitan | 9 | 35 | 151 | 214 | 19 | 5 | 8 | 0 | 441 |
| Mexico | 0 | 3 | 53 | 10 | 0 | 5 | 0 | 0 | 71 |
| Michiana | 3 | 18 | 77 | 148 | 4 | 4 | 0 | 0 | 254 |
| Mid-South | 0 | 1 | 22 | 35 | 1 | 1 | 0 | 0 | 60 |
| Mo-Kan | 1 | 5 | 44 | 24 | 4 | 1 | 0 | 0 | 79 |
| New England | 1 | 22 | 125 | 64 | 20 | 6 | 4 | 0 | 242 |
| Northeastern Ohio | 14 | 50 | 162 | 384 | 14 | 22 | 9 | 2 | 657 |
| Northern Calif. | 0 | 17 | 116 | 132 | 10 | 0 | 0 | 1 | 276 |
| No. Ill. & So. Wis. | 0 | 13 | 29 | 68 | 1 | 1 | 0 | 1 | 113 |
| Northwestern Pa. | 0 | 14 | 53 | 85 | 2 | 2 | 3 | 0 | 159 |
| Ontario | 2 | 48 | 165 | 220 | 9 | 14 | 1 | 1 | 460 |
| Oregon | 1 | 8 | 44 | 65 | 4 | 3 | 0 | 0 | 125 |
| Philadelphia | 7 | 49 | 135 | 199 | 21 | 11 | 5 | 1 | 428 |
| Piedmont | 1 | 12 | 55 | 47 | 11 | 0 | 4 | 1 | 131 |
| Pittsburgh | 6 | 20 | 152 | 135 | 8 | 3 | 1 | 0 | 325 |
| Quad City | 3 | 15 | 40 | 134 | 6 | 2 | 1 | 1 | 202 |
| Rochester | 0 | 3 | 41 | 26 | 2 | 0 | 0 | 0 | 72 |
| Saginaw Valley | 6 | 11 | 36 | 248 | 5 | 59 | 4 | 0 | 369 |
| St. Louis District | 1 | 19 | 98 | 91 | 7 | 2 | 0 | 0 | 218 |
| Southern Calif. | 1 | 34 | 167 | 113 | 8 | 11 | 0 | 0 | 334 |
| Tennessee | 0 | 10 | 62 | 214 | 1 | 8 | 0 | 1 | 296 |
| Texas | 1 | 22 | 106 | 112 | 4 | 7 | 2 | 0 | 254 |
| Timberline | 0 | 0 | 37 | 9 | 5 | 0 | 0 | 0 | 51 |
| Toledo | 1 | 11 | 27 | 93 | 2 | 3 | 1 | 0 | 138 |
| Tri-State | 0 | 5 | 57 | 46 | 0 | 5 | 1 | 0 | 114 |
| Twin City | 2 | 19 | 82 | 105 | 8 | 3 | 0 | 0 | 219 |
| Utah | 0 | 4 | 18 | 51 | 1 | 2 | 0 | 0 | 76 |
| Washington | 0 | 4 | 46 | 19 | 15 | 5 | 0 | 0 | 89 |
| Western Michigan | 1 | 19 | 63 | 182 | 3 | 7 | 1 | 0 | 276 |
| Western New York | 1 | 16 | 71 | 129 | 2 | 8 | 2 | 2 | 231 |
| Wisconsin | 8 | 66 | 121 | 459 | 8 | 18 | 4 | 1 | 685 |
| TOTAL IN REGULAR CHAPTERS | 129 | 860 | 3,634 | 6,127 | 329 | 389 | 92 | 21 | 11,581 |
| STUDENT CHAPTERS | | | | | | | | | |
| Massachusetts Institute of Technology | | | | | | 9 | | | 9 |
| Michigan State University | | | | | | 25 | | | 25 |
| Missouri School of Mines | | | | | 2 | 63 | | | 65 |
| Ohio State University | | | | | | 17 | | | 17 |
| Oregon State University | | | | | | 24 | | | 24 |
| Pennsylvania State University | | | | | | 18 | | | 18 |
| Polytechnic Institute of Brooklyn | | | | | | 16 | | | 16 |
| Ryerson Institute | | | | | | 22 | | | 22 |
| Texas A. & M. College | | | | | 1 | 35 | | | 36 |
| University of Alabama | | | | | | 40 | | | 40 |
| University of Illinois | | | | | | 62 | | | 62 |
| University of Michigan | | | | | | 6 | | | 6 |
| University of Wisconsin | | | | | | 29 | | | 29 |
| Wentworth Institute | | | | | | 37 | | | 37 |
| TOTAL IN STUDENT CHAPTERS | | | | | 3 | 403 | | | 406 |
| TOTAL IN ALL CHAPTERS | 129 | 860 | 3,634 | 6,127 | 332 | 792 | 92 | 21 | 11,987 |
| Foreign | | 21 | 423 | 8 | 34 | 35 | 2 | 0 | 523 |
| Non-Chapter | 1 | 3 | 15 | 1 | 11 | 0 | 2 | 0 | 33 |
| In Military Service | | | | | | | | | 11 |
| GRAND TOTAL | 130 | 884 | 4,072 | 6,136 | 377 | 827 | 96 | 21 | 12,554 |

trolled" circulation begun to all non-member foundries. TRANSACTIONS papers begun in Jan. 1958.

*1959 to Date* — Professional phase begun in May 1960. Changes made to de-emphasize house organ flavor, build up commercially competitive medium. "Controlled" circulation increased to current 22,000 to meet competitive distribution methods.

It should be realized that at least six rate changes occurred during this 16-year period, or a rate change every 32 months on the average. Thus, while Avg. Adv. Income and Avg. Adv. Income per Page have increased steadily, the Avg. No. of Advertising Pages per year has steadily declined since the 1951-1954 phase. At the same time, Avg. Total Expense (direct) and Avg. Expense per Page have steadily increased. The result has been an exceedingly erratic record of Net Income.

During the next two years it is intended to concentrate on 4 major factors: Increased gross income through intensive selling of a superior product; growth through increased Advertising volume (pages); holding the line on Expense by maintaining more realistic ratios between Editorial and Advertising content per issue; all toward the end of substantially increased Net Income.

## Membership Matters

At the recent Chapter Officers Conference it was decided that no membership "Targets" would be established for 1961-62 in view of uncertainty surrounding the new Dues schedule effective July 1. Our estimates of membership for the current year contemplate a possible loss of 15 per cent of Sustaining, Company and Personal members in 1961-62, and a recovery of one-third in 1962-63. While the Staff has prepared visual data in support of the dues action, we believe that no apology is necessary in light of AFS activities and their value to the industry.

A folder "Blueprint for Progress," explaining the necessity of a dues increase, has been mailed to all Chapter Officers and Directors and will go to the entire membership. Each member will also receive, with invoice for dues renewal, a second copy. President Hunt has written all Chapter Chairmen and those who attended the Chapter Conference a personal letter requesting Chapter cooperation.

A personally addressed letter will be sent shortly by the General Manager to all Sustaining and Company members, suggesting Research Patron membership. With the letter will go a folder describing briefly the 16 research projects currently recommended. Four Research Patron members are already on record, and our goal is 100 during the next several years.

## Chapters

The Chapter Contacts Manual for 1961-62 contains six charts illustrating major reasons for the dues increase. It is suggested that AFS Directors be familiar with these charts before visiting a Chapter and meeting with a Chapter Board.

As agreed, only one delegate per Chapter was accepted at full AFS expense for the 1961 Chapter Officers Conference; other delegates were invoiced for $45 each to cover meals and misc. expense during the meeting. Eighty Chapter Officers attended (same number as in 1960), representing all 47 Chapters.

## Conventions & Exhibits

During the past year the General Manager and Exhibit Manager have spent much time developing formalized "Conditions" to govern the servicing and conduct of future AFS Expositions. These "Conditions" have been broadcast throughout the country and acclaimed with almost universal praise. Special "ground rules" were also set up to govern our 1962 Exposition at Cobo Hall, Detroit, our position being that acceptance of definite ground rules is a definite condition to our signing a hall contract. These "Conditions" form Appendices "B", "C" and "D" to the February 1961 Board meeting Minutes.

On July 6 a public signing of Cobo Hall "ground rules," essentially as AFS wrote them, took place in Detroit. Attending were: President Hunt, Regional Vice-Pres. Lloyd, Director Jeter, Detroit's Mayor Miriani, the president of F.E.M.A., the General Chairman and Honorary Chairman of the 1962 Detroit Convention Committees, Chairman of the Detroit Chapter, the Cobo Hall Commissioner and the Hall Manager, the head of Detroit Convention & Tourist Bureau, and the presidents of Detroit's Building Trades Council, the Electrical Workers union and the Carpenters union. We consider this event a major breakthrough in getting more common sense and fair play into the servicing and handling of industrial expositions.

All effort is now under way to make the 1962 Exposition the most successful in AFS history. To do so we must overcome timidity over current business conditions, and fears over the Detroit labor setup. However, Cobo Hall is the finest in North America, Michigan is the largest castings producer, and the International Foundry Congress is an added attraction in our estimates of 16,000 attendance.

## Educational Activities

It is hoped that recommendations of T&RI Trustees and the AFS Finance Committee will be accepted to restore the Foundry Instructors Seminar in 1962. This event, to keep instructors in foundry and patternmaking at high schools and vocational schools up-to-date on foundry developments, is an important part of AFS-T&RI educational work. Four successful Seminars were held (in 1956, 1957, 1958 and 1959) before it was deferred as an economy measure.

## SH&AP Program

The Technical Director's report covers activities of the SH&AP Director during the past year. This valuable service to foundry groups and individual plants cannot be stressed too strongly as an AFS activity of direct value to Foundry Management. It offers practically the only unbiased, impartial source of information and counsel to which foundries can turn for assistance in safety, plant environment and air pollution matters.

## Training and Research Institute

T&RI has finally been granted full tax exemption as an organization exempt under Sec. 501 (c) (3) of the 1954 Internal Revenue Code. As a result, the T&RI Trustees have again recommended Board decision on the proposed National Foundry Training Center building, on which AFS has already spent over $80,000 in necessary preliminary steps. In the opinion of the General Manager, this project presents AFS with one of the greatest opportunities it has ever had to serve the training and research needs of the industry as they must be served if the industry is to progress. Admittedly, it is a big project, but one the industry has already shown willingness to support.

## International Affairs

AFS was officially represented at the 1961 International Foundry Congress in Vienna by Past President Walter L. Seelbach and the AFS exchange paper author A. J. Kiesler of General Electric Co. Some 15 Americans attended, including wives.

Next year AFS again will be host to the International Foundry Congress, in Detroit. Simultaneous 3-language translation will have to be provided at some of the sessions, and the schedule of International meetings is now being worked out. One AFS obligation in 1962 will be the arranging of pre-Congress Productivity Tours for those planning to attend. In this we will seek Chapter and individual plant cooperation.

## General Administration

Staff personnel was reduced by six persons when the budget was revised in January, and this has necessarily slowed some activity development. Budgets for 1961-62 recommend re-employment of four of these six . . . on Publications' editing, in Membership and Addressograph work, and as temporary assistance on 1962 Advance Registrations.

We expect 1961-62 to be a year of intense AFS activity in many directions, and it is hoped that full consideration will be given anticipated later needs for additional personnel as the year develops, especially in stenographic and clerical facilities.

Respectfully submitted,
WM. W. MALONEY
*General Manager*

# ANNUAL REPORT OF AFS SECRETARY

*(For Fiscal Year 1960-1961)*

This report covers developments of the past year in Membership and Chapter Activities and prospects for progressing these areas during 1961-62.

## Membership

Membership on June 30, 1961 totaled 12,554 . . . a net loss of 236 or 1.8 per cent below the total of 12,790 on June 30 a year ago. The loss in members represented a 3.1 per cent or $9,000 loss in dues income, due mainly to conversions to lesser classes of membership, resignations of Company members, and nonrenewals of delinquent members.

In consideration of the state of business during the past year the Chapters did a commendable job in retaining such a high percentage of the membership (98.2 per cent). By Regions, five showed a net loss for the year, one a gain, and one ended the year with the same total memberhip: Region 1 – 24 loss, Region 2 – 22 gain, Region 3 – even, Region 4 – 50 loss, Region 5 – 79 loss, Region 6 – 82 loss, Region 7 – 21 loss.

### Membership Record as of June 30, 1961

| | | | |
|---|---|---|---|
| Total Active Membership as of May 31, 1961 | | | 12,646 |
| Members Added June 1 to June 30, 1961 | | | |
| New Members | 79 | | |
| Members Reinstated | 56 | 135 | |
| Members Dropped | | | |
| June 1 to June 30, 1961 | | | |
| Resignations | 17 | | |
| Deaths | 1 | | |
| Delinquents Dropped | 209 | 227 | |
| Net Membership Loss June 30, 1961 | | | 92 |
| Total Active Membership as of June 30, 1961 | | | 12,554 |

### Fiscal Year to Date

| | | | |
|---|---|---|---|
| Total Active Membership as of June 30, 1960 | | | 12,790 |
| Members Added, | | | |
| July 1, 1960 to June 30, 1961 | | | |
| New Members | | 1,533 | |
| Adjustment | | 7 | |
| | | 1,540 | |
| Members Dropped, | | | |
| July 1, 1960 to June 30, 1961 | | | |
| Resignations | 385 | | |
| Deaths | 38 | | |
| Delinquents Dropped | 2,074 | | |
| Less Reinstated | 721 | 1,353 | 1,776 |
| Net Membership Loss, 12 Months | | | 236 |
| Total Active Membership as of June 30, 1961 | | | 12,554 |

### Modern Castings Circulation

| | |
|---|---|
| Members | 12,554 |
| Subscribers | 1,299 |
| Prospective Members | 6,568 |
| Other Circulation | 1,024 |
| | 21,445 |

Due to the fact that recessed business conditions continued, the membership targets were left at 13,650 for 1960-61, the same as for 1959-60. Targets have been dropped for 1961-62 as agreed to by the Chapters at the Chapter Officers Conference in June. However, a comparative report on each Chapter will be issued each month in lieu of the Target report used for the past four years. The report will give each Chapter its standing compared to the previous month and to June 30, 1961, as well as to other Chapters in the Region, and to all 47 Chapters.

## Membership Program for 1961-62

Membership activity for 1961-62 can only be generally forecasted due to possible effect of the new dues structure. Past experience in dues increases resulted in estimates of a 15 per cent loss during 1961-62 and the return of 5 per cent of this loss during 1962-63. However, the new structure is expected to materially increase dues income to the Society . . . 24.5 per cent for the first year and an additional 7.1 per cent the second year, based on actual income for 1960-61.

Included in the new dues structure is a flat 20 per cent dues refund program to Chapters on all classes of membership, a return to a policy of several years ago. It will net the Chapters a slightly greater income and increase their interest in obtaining larger memberships. The new policy also includes pro-rated refunds of Sustaining memberships based on the amount of Affiliate participation.

## Chapters

No new regular Chapters were established nor Chapter petitions filed during the past fiscal year. However, the Southern Tier Section of the Central New York Chapter is proceeding to obtain necessary approvals of surrounding Chapters in regard to boundaries for new "Southern Tier Chapter." By early fall this should be completed and a petition received for formation of a Chapter from this Section.

On August 1, 1961 the new Student Chapter at General Motors Institute was installed. This is an extremely active group of enthusiastic young men under the leadership of John Lowe, who is principally responsible for the formation of the Chapter.

Chapter interest is again being shown in the Winnipeg, Canada, area. H. R. Grout of Winnipeg attended the Chapter Officers Conference in June and agreed to discuss Chapter possibilities with the industry of the area.

The 18th Annual Chapter Officers Conference was held June 15-16 with a total attendance of 85 including 78 delegates representing all 47 Chapters. Workshop sessions were held the first day of the Conference at the Central Office Headquarters in Des Plaines. General sessions were held the second day at the LaSalle Hotel, Chicago. Five new National Directors (Messrs. Best, Hageboeck, Hall, Lonnee, and Schopp) were present for the two-day meeting. President Hunt presided as Conference Chairman.

Respectfully submitted,
ASHLEY B. SINNETT
*Secretary*

# ANNUAL REPORT OF AFS TREASURER

*(For Fiscal Year 1960-1961)*

The following documents are a part of this report: Annual Audit Report as of June 30, 1961; Comparative Statements of Itemized Income and Expense, Budgets vs. Actuals for fiscal years 1957-58 through 1960-61; Recommended Budgets of Itemized Income and Expense for 1961-62 and 1962-63; and are being forwarded to the Board of Directors for study prior to the Annual Board Meeting, August 10-11.

## Summary Analysis

*Annual Audit* — Following is a comparison of the financial position of the General Fund of the Society, 1961 vs. 1960:

| | 1961 | 1960 |
|---|---|---|
| *Assets* | | |
| Cash | $ 7,269. | $ 40,392. |
| Investment Agency Trust | 339,287. | 453,043. |
| Accounts Receivable | 13,986. | 14,523. |
| Inventory of Publications | 45,449. | 44,151. |
| Fixed Assets (net) | 355,112. | 350,661. |
| Deferred Charges | 9,591. | 6,922. |
| | $770,694. | $909,692. |
| *Liabilities and Principal* | | |
| Current Liabilities | $ 30,608. | $ 19,836. |
| Fund Principal | 740,086. | 889,856. |
| | $770,694. | $909,692. |

The major action affecting the Balance Sheet was the Board's action in February 1961 to withdraw approximately $153,000 from the AFS Investment Agent Trust Account . . . to liquidate (then) existing bank loans . . . and to provide working capital, in order to begin the 2-year budgetary period 1961-1963 without outstanding obligations. This action was precipitated over a period of years in which AFS expanded services, but anticipated income therefrom did not materialize. Also, continuous investments in the present Headquarters Building and land acquisitions for future use have totaled $80,000.

The Finance Committee is standing firm on a policy to stabilize the AFS financial position and keep expenditures within the amount of available funds.

## Operations Summary — 3 Years

| | 1961 | | 1960 | 1959 |
|---|---|---|---|---|
| | Rev. Budget | Actual | Actual | Actual |
| Income Activities (net) | $210,700. | $211,130. | $535,421. | $199,495. |
| Expense Activities (net) | 359,000. | 360,900. | 383,800. | 387,362. |
| Net Income or (Expense) | ($148,300.) | ($149,770.) | $151,621. | ($187,867.) |

Net expense in 1960-61 was off only $1,470 from the Budget as revised in December 1960. Naturally, many programs were started which required termination periods and expense when the 1961 Exposition was cancelled. Basically, the revised Budget was achieved.

It should be noted that some misunderstanding may arise from placing labels on accounts used in order to evaluate costs, thereby separating them from their true purpose as "technical activities" or the basic purposes of the Society. Rather than review the accounts in the usual way, we present below the cost in terms of their technical relationship per se. This not only proves their necessity but the contrast between what the membership "gives" versus what AFS must "earn" in order to provide a forum for the industry.

## Based on Average of Fiscal Years 1959-60 and 1960-61

| | | |
|---|---|---|
| Investment — By Members (gross) | $287,800. | |
| — By Income Activity (less direct sales expense) | 377,300. | $665,200. |
| Technical Activity — | | |
| Research and Committee Work (1) | $103,100.* | |
| Publications (2) | 259,200. | |
| Chapter Mtgs. and Convention (1) | 211,600. | |
| Exhibits and Educational Programs (3) | 90,400. | $664,300. |
| Excess Investment over Activity | | $ 900. |

*Direct expense only; AFS nor any Society could ever estimate the investment of man-hours involved in Committee work or cooperation of others to stretch our Research dollar for infinite projects.

In the above, we have deleted the direct cost of producing the applicable income in order that such cost will not distort the proportionate part that activity plays in the necessary programs of AFS; (1) Discussion and Development, (2) Dissemination, and (3) Demonstration.

## Operations Summary

*1960-61 vs. 1959-60 and 1958-59*

| | 1961 | 1960 | 1959 |
|---|---|---|---|
| *Income* | | | |
| Exhibit (net) | $ (44,463.) | $250,259. | $ (11,397.) |
| Membership Dues | 190,544. | 209,546. | 196,604. |
| Modern Castings (net) | 12,723. | 41,703. | (575.) |
| Buyers Directory | (488.) | 7,045. | (13,213.) |
| Income from Investments | 51,333. | 26,192. | 26,952. |
| Miscellaneous | 1,481. | 673. | 1,123. |
| Total | $ 211,131. | $535,421. | $ 199,494. |
| *Expense* | | | |
| General Administration | $ 87,638. | $ 96,976. | $ 94,454. |
| Technical Activities | 100,843. | 112,094. | 120,394. |
| Chapter Operations | 89,514. | 97,502. | 93,731. |
| General Publications (net) | 30,301. | 25,223. | 28,658. |
| Special Publications (net) | 3,479. | 11,892. | 15,615. |
| Convention (net) | 49,124. | 40,110. | 34,507. |
| Total | $ 360,901. | $383,799. | $ 387,362. |
| Net Income or (Expense) | $(149,770.) | $151,621. | $(187,867.) |

## Condensed Operating Statement of Income & Expense
### Fiscal Years 1959-1961
*(All Expenses Distributed to Major Activities — as Audited)*

| Net Income Activities | Net Income Amount | % of Total |
|---|---|---|
| Membership | $400,090. | 53.6 |
| "Modern Castings" | 54,426. | 7.3 |
| Exhibit | 205,796. | 27.6 |
| "Buyers Directory" | 6,557. | .9 |
| Investments and Miscellaneous | 79,679. | 10.6 |
| Total | $746,552. | 100.0 |

| Net Expense Activities | Net Expense Amount | % of Total |
|---|---|---|
| Technical Activities | $212,937. | 28.6 |
| Publications | 70,895. | 9.5 |
| Chapter Operations | 187,016. | 25.1 |
| Convention | 89,234. | 12.0 |
| General Administration | 184,614. | 24.8 |
| Total | $744,700. | 100.0 |
| Net Income, 2 Years | $ 1,851. | |

## 2-Year Income and Expense
*1959-1961 vs. Prior Periods*

| 2-Year Period | Excess Income or (Expense) |
|---|---|
| 1959-1961 Actual | $ 1,851. |
| 1957-1959 Actual | (60,388.) |
| 1955-1957 Actual | 41,963. |
| 1953-1955 Actual | 47,889. |
| 1951-1953 Actual | 126,040. |
| Average, five 2-year Periods | $ 31,471. |

## Annual Dues Income vs. Member Expense
*8 Years, 1953-54 to 1960-61, Inclusive*

| | No. Members June 30 | Avg. Dues Inc. Per Member | Avg. Net Exp. Per Member | Avg. Excess Expense Per Member |
|---|---|---|---|---|
| 1960-61 | 12,554 | $22.45 | $28.75 | $6.30 |
| 1959-60 | 12,790 | 22.98 | 30.71 | 7.73 |
| 1958-59 | 12,449 | 22.99 | 31.11 | 8.12 |
| 1957-58 | 12,880 | 23.15 | 29.00 | 5.85 |
| 1956-57 | 13,340 | 23.11 | 25.20 | 2.09 |
| 1955-56 | 12,488 | 23.79 | 25.65 | 1.86 |
| 1954-55 | 10,965 | 24.50 | 29.98 | 5.48 |
| 1953-54 | 11,551 | 20.01 | 22.83 | 2.82 |
| 8-yr. avg. | 12,377 | $22.87 | $27.90 | $5.03 |
| 8-yr. Total Excess Expense | | | | $62,256. |

## Budgets 1961-1963

Recommendations of the Finance Committee for the coming 2-year period are based on present conditions. As decided a year ago, the Committee again will meet for a six month review in order that adjustments may be made based on information at that time. All activities have been budgeted with caution, but the Committee does not want to delay proposed activities if the funds are available.

## 2-Year Dues Income vs. Expense Per Member
*1953-54 to 1960-61, Inclusive*

| 2-Year Periods | Avg. Members | Avg. Dues Inc. Per Member | Avg. Net Exp. Per Member | Avg. Excess Expense Per Member |
|---|---|---|---|---|
| 1959-1961 | 12,672 | $22.72 | $29.73 | $7.01 |
| 1957-1959 | 12,664 | 23.07 | 30.05 | 6.98 |
| 1955-1957 | 12,914 | 23.45 | 25.42 | 1.97 |
| 1953-1955 | 11,258 | 22.26 | 26.40 | 4.15 |
| Avg. 2-Yr. Periods | 12,377 | $22.87 | $27.90 | $5.03 |

## Investments as of June 30, 1961

| Investment Income | 1961 | 1960 | 1959 |
|---|---|---|---|
| Interest on U.S. Government Bonds | $ 1,339.55 | $ 608.29 | $ 331.07 |
| Interest on Canadian Government Bonds | 1,006.25 | 1,006.25 | 831.25 |
| Interest on Industrial Bonds | 8,195.69 | 8,506.17 | 7,702.29 |
| Dividends on Common Stocks | 8,680.84 | 9,081.18 | 9,282.46 |
| Discount on Treasury Bills | | | (170.67) |
| Gain or (Loss) on Securities Sold | 33.357.83 | 8,922.95 | 8,764.99 |
| Discount or (Premium) on Bonds Purchased | | 525.00 | (325.00) |
| Building Rental | 1,602.50 | 1,620.00 | 1,410.64 |
| Interest on Bank Loans | (2,839.03) | (4,076.94) | (874.58) |
| Total | $51,333.63 | $26,192.90 | $26,952.45 |

| Investments | Book Value | Redemption or Market Value June 30 | Increase or (Decrease) Redemption or Market Value over Book Value: % Incr. or (Decr.) | % of Total (Mkt.) |
|---|---|---|---|---|
| U.S. Government Bonds | $ 34,000.00 | $ 34,168.13 | .49 | 8.10 |
| Canadian Government Bonds | 25,000.00 | 23,437.50 | (6.25) | 5.56 |
| Industrial Bonds | 141,000.00 | 132,793.75 | (5.82) | 31.49 |
| Industrial Stocks | 134,291.39 | 231,331.13 | 72.26 | 54.85 |
| Total | $334,291.39 | $421,730.51 | 26.16 | 100.00 |

## Comparative Condensed Balance Sheets
*June 1958 to June 1961, Inclusive (As Audited)*

| Assets | June 30 1958 | June 30 1959 | June 30 1960 | June 30 1961 | Incr. or (Decr.) Over 1960 | 4-Year Incr. or (Decr.) |
|---|---|---|---|---|---|---|
| Cash | $ 127,573. | $ 28,858. | $ 48,867. | $ 22,468. | $ (26,399.) | $ (105,105.) |
| Investment Securities | 465,298. | 483,180. | 488,043. | 381,287. | (106,756.) | ( 84,011.) |
| Inventories | 47,974. | 44,605. | 44,151. | 45,449. | 1,298. | ( 2,525.) |
| Accounts Receivable | 13,647. | 10,459. | 14,522. | 13,986. | (536.) | 339. |
| Deferred & Prepaid Items | 18,928. | 16,885. | 6,922. | 9,590. | 2,668. | ( 9,338.) |
| Furniture & Fixtures (net) | 34,500. | 30,651. | 27,617. | 38,434. | 10,817. | 3,934. |
| Land & Buildings | 316,528. | 310,943. | 323,044. | 316,676. | (6,368.) | 148. |
| Total | $1,024,448. | $920,581. | $953,167. | $827,892. | $ (125,275.) | $ (196,556.) |
| Liabilities | | | | | | |
| Current Liabilities | $ 58,739. | $146,356. | $ 19,836. | $ 35,594. | $ 15,758. | $ (23,145.) |
| Fund Principals | 965,709. | 774,225. | 933,331. | 792,298. | (141,033.) | (173,411.) |
| Total | $1,024,448. | $920,581. | $953,167. | $827,892. | $ (125,275.) | $ (196,556.) |

### Assets — 10 Years

*1952-1961, Inclusive*

| As of June 30 | Investments (At Cost) | Fixed Assets | Inventories | Fund Principals |
|---|---|---|---|---|
| 1961 | $339,287. | $355,112. | $45,449. | $792,299. |
| 1960 | 453,043. | 350,661. | 44,151. | 933,332. |
| 1959 | 449,180. | 341,594. | 44,605. | 774,225. |
| 1958 | 465,298. | 351,028. | 47,974. | 965,710. |
| 1957 | 476,848. | 301,230. | 57,067. | 838,349. |
| 1956 | 560,960. | 307,087. | 52,218. | 908,625. |
| 1955 | 456,904. | 304,490. | 52,323. | 801,021. |
| 1954 | 564,573. | 13,046. | 37,135. | 912,455. |
| 1953 | 448,761. | 14,616. | 39,668. | 679,974. |
| 1952 | 330,600. | 15,560. | 32,470. | 622,775. |
| 10-Year Increase | $ 8,687. | $339,552. | $12,979. | $169,524. |

The Board and Chapter Officers agree that activities of AFS must become self-supporting. Such action, if realized, will in itself prove the activity necessary and worthwhile.

### Conclusion

The Staff of AFS are taking a greater interest and showing more responsibility as they learn more about their position, through being active in a number of areas, because of our austerity budgets. We trust the Board and membership are also encouraged toward broader participation in the many facets of AFS.

Respectfully submitted,
EDW. R. MAY
*Treasurer*

# ANNUAL REPORT OF AFS TECHNICAL DIRECTOR

*(Fiscal Year Ended June 30, 1961)*

This report covers the important technical activities of the Society during the past fiscal year.

### Convention Program

Although registered attendance during the 65th Castings Congress in San Francisco was comparatively poor, the technical program was unquestionably among the best in recent years. Many who attended commented favorably on the technical quality and variety of papers presented at the various sessions.

The program comprised a total of 44 technical sessions, 6 shop courses and 6 roundtable luncheons. A total of 126 papers were offered; following careful review by the Program & Papers Committees, 15 were rejected as incompatible with our publications policies, leaving a net of 111 papers on the Official Program. Although few who attended realized it, a total of 336 presiding officers and authors of papers were on the Official Program.

Again this year, preprinting of the papers was initiated in the TRANSACTIONS section of the January issue of MODERN CASTINGS and continued each month, 29 papers being so published in advance of the congress. Practically all papers remaining will be published in the magazine until December 1961. This broad distribution of the technical papers has resulted in a growing interest on the part of foundrymen and a marked increase in the number of written discussions received. Frequently such discussions, followed by the author's rebuttal, are equally as important as the material contained in the paper.

The Charles Edgar Hoyt Memorial Lecture was presented this year by John B. Caine, Consultant, Cincinnati, on the subject of "Cast Metals and Cast Shapes." This lecture was published in the July issue of MODERN CASTINGS and an additional quantity in separate pamphlet form is available for sale.

A Tentative Program has already been prepared for the 66th Castings Congress & Exposition to be held in Detroit, May 7-11, 1962, in conjunction with the 29th International Foundry Congress. Announcements in the three official languages – English, French and German – are being sent to 22 foreign foundry associations to stimulate attendance at the International Foundry Congress and the submittal of Official Exchange Papers from the members of the International Committee, as well as additional foreign papers.

Certain official functions, a part of the International Congress involving the International Committee, are not shown on the Tentative Program, but will be included as soon as definite information is received. Special arrangements will have to be made for some functions to provide simultaneous translation in the three official languages. Steps are being taken to develop sources for such facilities.

### Apprentice Contest

The 1961 Robert E. Kennedy Memorial Apprentice Contest proved successful. A total of 20 Chapters participated and 171 companies sponsored entries. Of the 466 total entries (461 in 1960), 129 were submitted in the National Contest following elimination at the local Chapter level, in the fields of gray iron, nonferrous and steel castings as well as wood and metal pattern-making. Because of unavailable space and cancellation of the exhibit in San Francisco, the pattern entries were not exhibited but will be included along with 1962 contest entries in Detroit next year.

### Publications Program

*General Publications.* By the end of the fiscal year a total of 2,114 copies of the Annual TRANSACTIONS for 1960 (vol. 68) had been shipped to all who ordered them. Of this total, 1,084 were sold to domestic and foreign purchasers and 756 gratis copies sent to Company, Sustaining and Honorary Life members on request. Of a total of 2,300 copies printed, an inventory of 186 is on hand to satisfy future demands. Total sales of General Publications, which includes sales of current and previous TRANSACTIONS, amounted to $18,683.50, representing the highest annual sales for the past 10-year period.

*Special Publications.* During the fiscal year, gross sales of Special Publications totaled $37,733, and royalties $1,455, resulting in gross income of $39,189. Deducting dealers' discounts of $3,868, this leaves net sales of Special Publications $35,321, compared with $22,987 the previous fiscal year. Combined net sales of both General and Special Publications during the past fiscal year amounted to $54,004, highest since 1958, and compared with an 11-year average of about $45,400.

The following publications were produced during the year: 1960 TRANSACTIONS (vol. 68); Gating & Risering of Castings

(Gray Iron Research Report); Design of Die Castings (reprint); Bibliography of Shell Molding; Symposium on Solidification; Patternmaker's Manual (2nd Edition); Classification of Foundry Cost Factors (Reprint); Cast Metals and Cast Shapes (1961 Hoyt Lecture); Cast Metals Handbook (balance of original printing bound).

A revised edition of the PATTERNMAKER'S MANUAL was published and initially offered for sale in September. By the end of the fiscal year, of the original 3,000 printed approximately 1,400 remain in stock, representing sales of 1,579 copies in a comparatively few months.

A new book, "Molding Methods & Materials," is being edited by the Staff, but the remaining material yet to be prepared by the committee will still require several months for completion. Final draft of the "Manual of Foundry Refractories" has been completed and now the illustrations must be acquired and it will be ready for publication. The revised edition of the SAND HANDBOOK is still in process of preparation by the Handbook Committee of the Sand Division, but it is hoped that the committee will have completed its task and the book can be likewise published this year.

The 10-Year Index to the Annual TRANSACTIONS is ready for publication. A revised edition of "Recommended Practices for Sand Casting of Light Alloys" is practically complete and should be published the coming year.

## Research Projects

All current research projects are progressing in a satisfactory manner under sponsorship of the AFS Training and Research Institute. Several annual progress reports were presented during the Castings Congress in San Francisco. These investigations are a continuing source of new technological development to assist the industry in improving its product.

## Training and Research Institute

Since the beginning of the calendar year, the AFS Training and Research Institute has given nine courses, as follows: Gating & Risering of Castings, Cupola Melting of Iron, Sand Control & Technology, Production of Ductile Iron, Preventive Maintenance, Shell Molds & Cores, Economical Purchasing of Foundry Materials, Core Practices, Sand Testing.

Registration in these courses involved a total of 301 students in nine courses. During the remainder of the year, courses are to be given on Gating & Risering of Castings, Scrap Causes & Remedies, Metallurgy of Gray Iron, Sand Control & Technology, Cleaning Room Operation, Production Scheduling & Control, and New Core Techniques, making a total of 16 courses during the year 1961. Of these five have been given in Regional areas in cooperation with the Texas, Tennessee, Ontario and Birmingham Chapters. Since the inception of the T&RI program, a total of 1814 registrants have attended 60 courses.

## Safety, Hygiene & Air Pollution Activities

The Director of the SH&AP program has been successful in preparing and having published in the Journal of the Air Pollution Control Association, seven informative reports on foundry air pollution problems. These present the foundry side, and since the Air Pollution Control Association is a control officials' organization, the reports will be accepted in their entirety and will have the advantage of the cloak of approval by the Control officials themselves. This will provide a valuable weapon in handling favorably future legal developments.

The Federal appropriation for increased control over water pollution has resulted in a new AFS Committee on Water Pollution. The purpose of this committee is to help foundrymen who will soon be confronted with the problem of disposing of foundry wastes, especially those containing phenols and formaldehydes.

It is encouraging to note that the other technical divisions of the Society are becoming increasingly conscious of safety, hygiene and air pollution. The Light Metals Division, for example, will publish a chapter prepared by the Director on Safety and Hygiene in handling and processing light alloys, as part of the book "Recommended Practices for Sand Casting of Light Alloys." Similarly, the Shell Mold & Core Committee of the Sand Division will publish a chapter prepared by the Director on the health aspects of materials used in shell mold and coremaking.

There continue to be an increasing number of requests from member foundries for consultation on problems in this field. The Director also prepares a monthly column for MODERN CASTINGS which is profitably being used to assist in obtaining an increased volume of advertising from suppliers of SH&AP equipment. The Director has served as an instructor in several AFS Training and Research Institute courses during the year where the subjects are included in the courses.

## Technical Inquiries

It appears that the Society is receiving an increasing number of technical inquiries. During the past fiscal year, well over 300 inquiries have been received both from members of the Society and others. Unquestionably, it is excellent tangible evidence of the value of membership in AFS. In addition, they provide an entree to induce non-members to acquire membership.

Respectfully submitted,
S. C. MASSARI
*Technical Director*

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# Minutes
# First Meeting of AFS Board of Directors 1960-61
# Sheraton Hotel, Philadelphia — May 14, 1960

ROLL CALL: President N. J. Dunbeck, *presiding;* Vice-President A. L. Hunt.

*Directors (Exp. 1961):* D. W. Boyd, R. R. Deas, Jr., Jake Dee, W. L. Kammerer, D. E. Matthieu, C. E. Nelson, H. M. Patton, C. A. Sanders.

*Directors (Exp. 1962):* N. N. Amrhein, D. L. Colwell, E. C. Jeter, R. E. Mittlestead, A. J. Moore, J. N. Wessel.

*Directors (Exp. 1963):* W. C. Jeffery, T. T. Lloyd, J. T. Moore, W. E. Sicha, D. E. Webster.

*Staff:* Wm. W. Maloney, Gen. Mgr.; S. C. Massari, Tech. Dir.; A. B. Sinnett, Secy.; E. R. May, Treas.

*Directors Absent:* W. H. Oliver (Exp. 1962), R. R. Ashley (1963), A. E. Falk (1963).

## Election of Regional Vice-Presidents

A quorum having been established, the President announced that, all 1960-61 Directors having voted, the following Regional Vice-Presidents had been selected in their respective regions, and asked confirmation by the Board:

Region 1 — D. E. Matthieu
Region 2 — A. J. Moore
Region 3 — D. L. Colwell
Region 4 — D. W. Boyd
Region 5 — H. M. Patton
Region 6 — Jake Dee
Region 7 — J. N. Wessel

On motion duly made, seconded and carried, Regional Vice-Presidents for 1960-61 were elected as recommended.

## Formation of Executive and Finance Committees

The President stated that, under the by-laws, the Executive Committee comprises the President, Vice-President, immediate Past President and all elected Regional Vice Presidents. He also stated that the Finance Committee comprises the President, Vice-President, immediate Past President, General Manager and Treasurer.

## Appointment of Board Committees

The President stated that, as empowered by the by-laws, it was his duty to appoint Board and other official committees for the fiscal year 1960-61. On motion duly made, seconded and carried, appointment of the following bodies was approved:

*Regional Vice-Presidents* — Region 1 — D. E. Matthieu, Region 2 — A. J. Moore, Region 3 — D. L. Colwell, Region 4 — D. W. Boyd, Region 5 — H. M. Patton, Region 6 — Jake Dee, Region 7 — J. N. Wessel.

*Executive Committee* — N. J. Dunbeck, *Chairman;* D. W. Boyd, D. L. Colwell, J. Dee, A. L. Hunt, D. E. Matthieu, A. J. Moore, C. E. Nelson, H. M. Patton, J. N. Wessel.

*Finance Committee* — N. J. Dunbeck, *Chairman;* A. L. Hunt, C. E. Nelson, Wm. W. Maloney, Edw. R. May.

*Nominating Committee* — C. E. Nelson, *Chairman;* L. H. Durdin; Region 1 — H. Reitinger, Sr.; Region 4 — R. A. Payne, J. A. Barrett; Region 5 — L. H. Brogley; Region 6 — J. C. Henderson, E. W. O'Brien.

*Chapter Contacts Committee* — A. L. Hunt, *Chairman;* all Regional Vice-Presidents and all Directors, Members of Committee.

*T&RI Trustees* — H. Bornstein (1957-1961), *Chairman;* N. J. Dunbeck (1959-1962), L. H. Durdin (1960-1964), A. L. Hunt (1960-1963), C. E. Nelson (1958-1961), R. A. Oster (1958-1962), I. R. Wagner (1959-1963).

*Retirement Fund Trustees* — Robt. S. Hammond, *Chairman;* Collins L. Carter, Ralph L. Lee, Frank W. Shipley.

*Annual Lecture Committee* — H. H. Wilder, *Chairman;* M. E. Brooks, Wm. Romanoff, F. B. Rote, T. W. Seaton, V. E. Zang.

*Publications Committee* — T. T. Lloyd, *Chairman;* T. E. Barlow, F. C. Bennett, Charles Locke, F. L. Riddell, H. H. Wilder.

*Board Nominating Committee* — A. L. Hunt, *Chairman;* W. H. Oliver, H. M. Patton.

*National Castings Council Representatives* — N. J. Dunbeck, A. L. Hunt.

*Foundry Educational Foundation Trustees* — W. L. Kammerer (1959-1961), N. N. Amrhein (1960-1962).

*International Representatives* — N. J. Dunbeck, *Frank J. Dost, *Wm. W. Maloney, Dr. A. B. Everest, European Repr.; W. A. Gibson, Australian Repr.

## Organization for Chapter Contacts

President Dunbeck urged close contact with the Chapters by the Directors, because of the size of the Society and the inability of the President and Vice-President to visit all the Chapters. The following schedule of Chapter Contacts was presented and accepted:

### Region 1

*Regional Vice-President:* D. E. Matthieu.
*Chapter Group A*
R. R. Ashley (Connecticut, New England, Massachusetts Institute of Technology, Wentworth Institute).
D. E. Matthieu (Metropolitan, Brooklyn Polytechnic Institute).
*Chapter Group B*
D. E. Matthieu (Chesapeake, Philadelphia, Piedmont).

*Official Delegate to 1960 International Congress.

### Region 2

*Regional Vice-President:* A. J. Moore.
*Chapter Group C*
D. E. Webster (Central New York, Eastern New York, Rochester).
*Chapter Group D*
A. J. Moore (Eastern Canada, Ontario).
*Chapter Group E*
W. H. Oliver (Northwestern Pennsylvania, Western New York, Pennsylvania State University).
D. E. Webster (Pittsburgh).

### Region 3

*Regional Vice-President:* D. L. Colwell.
*Chapter Group F*
W. E. Sicha (Northeastern Ohio, Ohio State University).
*Chapter Group G*
D. L. Colwell (Toledo).
R. E. Middlestead (Canton District, Central Ohio).

### Region 4

*Regional Vice-President:* D. W. Boyd.
*Chapter Group H*
E. C. Jeter (Detroit, University of Michigan).
C. E. Nelson (Saginaw Valley).
*Chapter Group I*
T. T. Lloyd (Michiana, Michigan State University).
D. W. Boyd (Western Michigan, Central Michigan).
*Chapter Group J*
R. R. Deas, Jr. (Central Indiana, Cincinnati District).

### Region 5

*Regional Vice-President:* H. M. Patton.
*Chapter Group K*
J. T. Moore (Chicago).
*Chapter Group L*
N. N. Amrhein (Wisconsin, University of Wisconsin).
*Chapter Group M*
H. M. Patton (Twin City).
J. T. Moore (N. Illinois-S. Wisconsin).
C. A. Sanders (Central Illinois, Quad City, University of Illinois).

### Region 6

*Regional Vice-President:* Jake Dee.
*Chapter Group N*
W. L. Kammerer (Corn Belt, Mo-Kan, St. Louis District, Timberline, University of Missouri).
*Chapter Group O*
W. C. Jeffery (Birmingham District, Mid-South, Tennessee, University of Alabama).
*Chapter Group P*
Jake Dee (Mexico, Texas, Tri-State, Texas A.&M. College).

### Region 7

*Regional Vice-President:* J. N. Wessel.
*Chapter Group Q*
A. E. Falk (Northern California, Southern California, Utah).
*Chapter Group R*
J. N. Wessel (British Columbia, Oregon, Washington, Oregon State College).

## Schedule of Regional Administration Meetings 1960-61

The President announced that the Regional Vice-Presidents, as in the past, should assume responsibility for their respective Regional Administration Meetings in 1960-61, and to function as the Society's "Official Representatives" in their respective Regions. Dates for all 1960-61 Regional Administration Meetings were approved. It was agreed that all meeting details would be finalized by the Central Office Staff, working with the Regional Vice-Presidents and the Directors and Chapter Officers.

## Approval of Convention and Exhibit Cities

The General Manager presented for Board approval the following schedule of Convention & Exhibit cities and dates for

1963 and 1964: May 1963 – St. Louis (Technical Congress with no Exhibit); May 1964 – Atlantic City (Technical Congress and unlimited Exhibit). The General Manager stressed the difficulty in retaining hotels and Convention halls in future years if best facilities are to be obtained by the Society. Following discussion, the recommended schedule was approved on motion made, seconded and carried.

### Next Meeting of Board of Directors

The President announced that the next meeting of the Board of Directors would be the Annual Board Meeting to be held August 11-12 in Chicago.

There being no further business to be considered, the first meeting of the 1960-61 Board of Directors was declared adjourned.

Respectfully submitted,
ASHLEY B. SINNETT
*Secretary*

APPROVED:
N. J. DUNBECK, *President*
June 22, 1960

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# Minutes
# Annual Meeting AFS Board of Directors
# Ambassador West Hotel, Chicago — August 11-12, 1960

ROLL CALL: President N. J. Dunbeck, *presiding;* Vice-President A. L. Hunt.

*Directors (exp. 1961):* D. W. Boyd, Jake Dee, W. L. Kammerer, D. E. Matthieu, C. E. Nelson, H. M. Patton, C. A. Sanders.
*Directors (exp. 1962):* N. N. Amrhein, D. L. Colwell, R. E. Mittlestead, A. J. Moore, W. H. Oliver.
*Directors (exp. 1963):* A. E. Falk, W. C. Jeffery, T. T. Lloyd, J. T. Moore, W. E. Sicha, D. E. Webster.
*Staff:* Gen. Mgr. Wm. W. Maloney, Tech. Dir. S. C. Massari, Sec. A. B. Sinnett, Treas. E. R. May.
*Directors Absent:* R. R. Deas, Jr. (exp. 1961); E. C. Jeter, J. N. Wessel (1962); R. R. Ashley (1963).

A quorum having been established, President Dunbeck called the Annual Meeting of the AFS Board of Directors to order.

### Reading of Minutes

Minutes of the final meeting of the 1959-60 Board of Directors held May 10, 1960, and of the first meeting of the 1960-61 Board of Directors held May 14, were presented in brief, having been approved by letter ballot of the Directors.

### Staff Reports

Reports of the General Manager, Secretary, Treasurer, and Technical Director were briefed, having been mailed in advance. All reports were accepted on motion, and were made a part of the minutes.

### Reports of Board Committees

*F.E.F. Trustees.* Director Amrhein presented a report on the F.E.F. meeting held May 11, 1960, with particular reference to a recent F.E.F. survey concerning Engineers and Technicians. It was recommended that F.E.F. allocations to 17 present cooperating schools should be increased. A list of 21 schools was submitted for consideration as future F.E.F. schools and affiliates, and it was proposed that a technical institute program be established at 10 technical institutes. It was pointed out that if the goal is accomplished it would produce approximately 1100 F.E.F. registrants, 260 full scholarships and 840 participating students available for summer work, which should make available 320 to 340 college graduates and 125 technician graduates per year.

*Membership Committee.* President Dunbeck reported the committee would now be known as the Membership Committee: Chairman, Director Amrhein, and including all Regional Vice-Presidents. Activity will basically follow the same pattern as last year in an effort to develop more Company and Sustaining members.

*N.C.C. Representatives.* The President reported on the National Castings Council meeting held May 12 during the 1960 Castings Congress, attended by himself and Past President Nelson. He stated that the Council is intended mainly to develop unity between the various foundry associations and for service to the industry in the event of another national emergency. Report included comments on matters discussed at the May meeting, including the problem of foreign competition.

*Self-Appraisal Committee.* The President stated that this committee would not be appointed during 1960-61 until its purposes could be more fully specified.

*Technical Council.* The Technical Director presented minutes of the Technical Council meeting held June 7, containing reports of Divisions and General Interest Committees on 1959-60 activities and programs for 1960-61. Without discussion, the minutes were accepted.

### Training and Research Institute

The meeting of T&RI Trustees held May 10 was briefly reviewed by the Technical Director. Discussion concerned the conducting of courses similar to T&RI by other organizations and their possible effect on the T&RI program. The Trustees approved partial payment of accumulated fees to the attorneys working on T&RI tax classification, action subject to approval of the AFS Finance Committee.

*Current Tax Classification.* The General Manager reported on the T&RI tax exemption request, stating that Internal Revenue rules governing scientific organizations was still under study and had not yet been approved, that action on behalf of T&RI must await approval of these rules. Although the initial notice on these rules was issued in December 1959, they still have not been finalized. The General Manager presented a sequence of meetings and appeals dating back to December 1957, when the initial application was made.

The attorneys now state, he reported, that the primary problem holding up rulings on scientific organizations is the treatment of organizations that do research for a fee. While it is recognized that this does not involve AFS, it appears that no action will be taken on the T&RI case until this section of the rules is resolved.

*Status of Proposed T&RI Building.* President Dunbeck reported that the Finance Committee gave thorough study to construction of the proposed T&RI Foundry Training Center building, as requested by the Board. It was the Committee's recommendation that no action on construction should take place until such time as the problem of tax exemption for T&RI is resolved.

*Progress of T&RI Activities.* Brief reviews on the present

T&RI program were given by the General Manager and Technical Director. It is apparent that the T&RI program, from its inception and for the 3-year period, is comparable with engineering institutes of long standing.

Ensuing discussion resulted in the Directors recommending that a program of publicity be developed on the T&RI program. Information should be circulated to the industry giving the complete facts as to how the program was conceived and its purpose, facts about the building and its benefits to the industry. On unanimous approval of the Board, without vote, the President instructed the Staff to develop immediately a brochure to be circulated to the entire industry covering all aspects of the T&RI program.

### Report on Discussions With F.E.M.A.

President Dunbeck reported on discussions with officials of F.E.M.A. in regard to exhibit frequency. While the situation was not completely resolved, he said, a much better understanding now exists within F.E.M.A. as to recent actions taken by AFS in regard to exhibits. Further meetings will be held between the officers of AFS and F.E.M.A. in an attempt to resolve any remaining differences between the two groups.

### Delegates to 1960 International Congress

President Dunbeck requested official appointment of F. J. Dost and W. W. Maloney as Official Delegates of AFS to the 1960 International Foundry Congress in Zurich. On motion duly made, seconded and carried, approved.

### Approval of Annual Audit

The Annual Audit of the Society's financial books as of June 30, 1960, prepared by George V. Rountree & Co., Chicago, certified public accountants, was presented, discussed in detail and approved by motion duly made, seconded and carried.

### Election of Staff Officers

In Executive session, the Board of Directors re-elected for one-year terms the following Staff officers: General Manager Wm. W. Maloney; Secretary A. B. Sinnett; Treasurer E. R. May; Technical Director S. C. Massari.

### Income and Expense Budgets 1960-61

The President pointed out the continuing necessity to restore AFS finances to a sound, even keel. He stated that expected improvement in general industrial conditions and the foundry industry had not materialized and that 1960-61 budgets therefore should be set conservatively. It was agreed, he said, that the Finance Committee would reconvene in mid-year to consider whether business conditions and budget developments in the first six months warrant reconsideration of Staff-recommended activities not presently undertakable.

The Finance Committee called attention to actual Excess Income totaling $151,622 in 1959-60 . . . and budgetary proposals for Excess Expense of $162,200 in 1960-61 (as established in July 1959 when a 2-year budget was developed for 1959-61). The President stated that the Committee felt it essential to materially revise 1960-61 budget estimates so as to produce a recommended Excess Expense of $106,600 (instead of $162,200) . . . and a 2-year Excess Income of $45,022 (instead of $1,000 as budgeted in 1959). Detailed recommendations of the Committee were presented affecting budgetary proposals, and approved on motion made, seconded and carried.

### Approval of Interim Financing

On motion made, seconded and carried, the Board approved the following interim actions of the Finance Committee: (a) action re bank loans totaling $100,000, on the schedule of: $40,000 in July 1960; $30,000 in August, and $30,000 in September; (b) action to accomplish amortization and liquidation of present and proposed Bank Loans per Committee recommendations.

### Chapter Contacts 1960-61

Vice-President Hunt called attention to the Chapter Contacts Manual provided all Directors and requested the Directors to study carefully the manual and arrange Chapter visits at the earliest possible date. Regional Administration Meetings were discussed as an excellent media for reaching all chapter officers, and Directors were urged to see that their respective Chapters attend these meetings. The Vice-President strongly stressed membership activity through the Board's Membership Committee, commenting that the Regional Vice-Presidents are responsible to follow through both Chapter contacts and the Membership program.

### Appointment of F.E.F. Trustee

The President announced appointment of Director Warren C. Jeffery to serve a term of two years (March 1961-March 1963) as an AFS Trustee to F.E.F., succeeding Director W. L. Kammerer, Director N. N. Amrhein to remain as the second Trustee for the year 1961-62.

### AFS Film

The President stated that he was considering development of a short color-sound film at the expense of his company, for showing to AFS Chapters because of his inability to make frequent Chapter visits. A tape of the initially proposed script for such a film was presented and the Directors expressed approval of the plan. The General Manager pointed out the need for continuity of such a program each year, expressing the hope that the cost of such a film might be borne by AFS in the future if Chapter reception indicates the advisability.

### Appointment of 1960-61 Nominating Committee

The President announced formation of the 1960-61 Nominating Committee at the Executive Committee meeting held August 11.

### Announcement of Next Board Meeting

The President announced that the next meeting of the Board of Directors would be held at the Cloister, Sea Island, Ga., February 20-21-22, 1961.

There being no further business to be considered, the Annual Meeting of the Board of Directors was declared adjourned.

Respectfully submitted,
ASHLEY B. SINNETT
*Secretary*

APPROVED:
NORMAN J. DUNBECK, *President*
November 9, 1960

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# Minutes
# Meeting of AFS Board of Directors
# The Cloister, Sea Island, Ga. — Feb. 20-21, 1961

ROLL CALL: Presiding, President N. J. Dunbeck; Vice-President A. L. Hunt.

*Directors (exp. 1961):* D. W. Boyd, R. R. Deas, Jr., Jake Dee, W. L. Kammerer, D. E. Matthieu, C. E. Nelson, C. A. Sanders.

*Directors (exp. 1962):* N. N. Amrhein, D. L. Colwell, E. C. Jeter, R. E. Mittlestead, W. H. Oliver, J. N. Wessel.

*Directors (exp. 1963):* A. E. Falk, W. C. Jeffery, T. T. Lloyd, J. T. Moore.

*Headquarters Staff:* Wm. W. Maloney, Gen. Mgr.
*Absent:* Directors H. M. Patton, A. J. Moore, R. R. Ashley, W. E. Sicha, D. E. Webster.

## Approval of Minutes

A quorum having been established, President Dunbeck stated that Minutes of the Executive Committee meeting held Aug. 6, 1960, and of the Annual Meeting of the Board of Directors held Aug. 6-7, 1960, had been approved by letter ballot; Minutes of the Board of Awards meeting Dec. 12, 1960, and of the Finance Committee meeting held Dec. 14, to be presented for approval at current meeting. No request being made for further reading of Minutes, reading was dispensed with.

## Staff Reports

Six-month reports of the General Manager, Secretary, Treasurer, and Technical Director were briefed, having been mailed in advance. All reports were accepted on motion, and were made a part of the Minutes.

## Reports of Board Committees

*Annual Lecture Committee.* The Hoyt Memorial Lecturer for 1961 will be John B. Caine, Cincinnati, on "Cast Metals and Cast Shapes."

*Chapter Contacts.* Chairman Hunt distributed requests for all Directors to record Chapter contacts made to date, and asked each Director to write their Chapters after visits. The General Manager recommended that provision be made for a Staff visit to the Mexico Chapter at least annually. It was the consensus that 1961-62 budgets should include such a Staff visit, and that the Directors should try to visit the Mexico Chapter whenever possible.

*1960 International Delegates.* The General Manager reported briefly on his attendance at the International Foundry Congress in Zurich, Switzerland, as an AFS Official Delegate with past President F. J. Dost. He also reported on visits to foundry associations in England, Norway, Sweden, Denmark, Belgium, Germany and Ireland, and stated strong interest was expressed for attending the 1962 International in Detroit.

*F.E.F. Trustees.* Director Amrhein presented a written report on meeting of F.E.F. Trustees held Oct. 28. Highlights: technical institute scholarships established at Western Michigan University, Erie County Technical Institute, Wentworth Institute, with funds provided by Dow Chemical Co., American Brake Shoe Co. and Griffin Wheel Co.; scholarship fund established at Western Michigan University by Central Michigan Chapter AFS, under F.E.F. administration; no conflict apparent between aims of F.E.F. technical institute program and T&RI programs.

Further highlights: 7 candidates selected for Wheelabrator Corp. fellowships; of 168 graduates in 1960, only 59 employed by F.E.F. members due to business conditions; proposed blueprint for F.E.F. future in next decade presented and discussed.

Question was raised as to whether AFS Chapters should be encouraged to contribute Chapter funds to F.E.F. The General Manager agreed to look up prior Board discussions on the point for possible determination of AFS policy at a future Board meeting.

*Membership Committee.* Chairman Amrhein reported on activities toward increasing Company and Sustaining memberships through Chapter committees working with AFS Regional Vice-Presidents and Directors. Report of Committee was made a part of the Minutes as Appendix "A." In discussion, recommendation was made that AFS should consider, at the earliest practicable date, employment of a "top-notch man to solicit foundry management in personal contacts.

*Nominating Committee.* Director Nelson, as Chairman of the 1960-61 Nominating Committee, announced the names of Officers and Directors nominated for election at San Francisco in May 1961.

*Publications Committee.* The General Manager stated that the Staff planned to recommend to the Publications Committee that prices of the annual Bound Volume of TRANSACTIONS be substantially increased. It does not appear, he said, that higher prices can be expected to materially decrease sales, because of the volume's great reference value, and that he hoped to reduce substantially the present average annual loss of approximately $12,000 on each new volume.

## Recommendations of Executive Committee

The President reported on the Executive Committee meeting held Feb. 20 to review main points developed in Regional Administration Meetings held during 1960-61. The following recommendations and suggestions were presented for Board consideration: (1) that more effort be made to encourage participation and membership of the Die Casting industry; (2) that the Chapter Officers Conference be held in 1961 as planned, on substantially the same expense basis as in 1960; (3) that the call for Chapters to name candidates for appointment to the Nominating Committee, and for consideration as National Directors, be incorporated in a single request to help eliminate confusion; (4) that a program of promotion and public relations be instituted, specifically to inform top management of the full AFS program of activities; (5) that a comprehensive brochure on the T&RI program be developed immediately and circulated to the entire industry, particularly to top management; (6) that AFS investigate possible use of a "blank check" invoice for paying member dues, enabling members to fill in amount, provide name of bank and to sign, in an effort to speed up dues payments; (7) that AFS develop a standard pocket badge for identification at Chapter meetings, and make them available to all Chapters; (8) that TRANSACTIONS papers published in MODERN CASTINGS contain a one-page abstract for quick reference; (9) that MODERN CASTINGS develop a strong "Question & Answer" page or section monthly.

It was agreed that all Directors should receive the full minutes of all Regional Administration Meetings held in all 7 Regions, and that all Chapters be advised of any Board actions taken on recommendations emanating from these meetings.

## Report of Board Nominating Committee

*Election of Director-at-Large.* On motion made, seconded and carried the Board of Directors unanimously approved recommendation of the Committee for election of the following Director-at-Large, to serve 3-year term 1961-1964: M. E. Nevins, President, Wisconsin Centrifugal Foundry, Inc., Waukesha, Wis.

*AFS Trustee to F.E.F.* On motion made, seconded and unanimously carried, recommendation was approved by the Board of Directors for appointment of the following to serve as AFS Trustee to F.E.F. for 2-year term 1961-1963: Warren C. Jeffery, McWane Cast Iron Pipe Co., Birmingham, Ala. The President pointed out that Director Jeffery would replace Director Kammerer as a Trustee, effective with the F.E.F. meeting April 1961.

*T&RI Trustee.* On motion made, seconded and unanimously carried, recommendation was approved by the Board of Directors for election of the following to serve as a Trustee of the AFS Training and Research Institute for 4-year term 1961-1965: Webb L. Kammerer, Midvale Mining & Mfg. Co., St. Louis. (To replace H. Bornstein, term 1957-1961.) The President pointed out that the seven T&RI Trustees comprise the following: AFS President, Vice-President, and immediate past President, terms of office during incumbency (3 years); and 4 additional Trustees elected by the Board for terms of 4 years each.

*AFS Retirement Trust Fund Trustee.* On motion made, seconded and carried, recommendation was approved by the Board of Directors for election of the following to serve as a Trustee of the AFS Retirement Trust Fund for indefinite term: H. A. Forsberg, Foundry & Mill Machinery Div., Blaw-Knox Co., East Chicago, Ind. (To replace R. S. Hammond, resigned.)

## Reports on Training & Research Institute

The President reviewed the 3d Annual T&RI Report, including audited financial statement as of Dec. 20, 1960, and minutes of T&RI Trustees meeting held Dec. 14. He stated that no present action was contemplated on the proposed T&RI Foundry Training Center building until T&RI tax exemption is resolved, and that any future action on the building could be undertaken only upon full approval of the AFS Board of Directors.

The General Manager reported on status of T&RI tax exemption efforts. New regulations for "scientific" organizations, he said, were published Dec. 1, 1960, and approved in early January without change. Although the Internal Revenue Service had promised to consider first the T&RI case, disagreements arose within the Service, he said, on interpretations of the approved new regulations, and no further action could be reported at this time.

## Recommendations of Finance Committee

The President stated that, following unanimous Board agreement on cancellation of the 1961 Exposition in San Francisco,

the Finance Committee had developed and forwarded to the Board specific recommendations on a number of important matters.

*Revised Budgets 1960-61.* The President stated that because of cancellation of the 1961 Exposition, due to lack of exhibitor response, the Finance Committee had drastically revised the Income and Expense budgets for 1960-61, calling for a decrease of over $93,000 in Expense during the 6 months January-June. Many steps were already taken to this end, he said, as advised to the Board. At his request, motion was made, seconded and carried, approving the following 1960-61 budget revisions:

| | Original Budget | Revised Budget | Incr. or (Decr.) |
|---|---|---|---|
| Total Income | $851,200. | $686,700. | ($164,500.) |
| Total Expense | 928,700. | 835,000. | ( 93,700.) |
| Excess Expense | $ 77,500. | $148,300. | $ 70,800. |

The General Manager stated that if further Expense cuts proved necessary as the year progressed, he was prepared to act promptly with Finance Committee approval.

It was pointed out that the May Board meeting announced for San Francisco had been scheduled, for economy purposes, to exclude a group luncheon. Following suggestion, it was agreed without vote that the Board luncheon would be scheduled and the expense prorated among all present. Director Sanders then requested opportunity to personally finance cocktails preceding the luncheon. Without vote, his offer was accepted and President Dunbeck expressed the Board's thanks for his generosity.

In response to question on plans for the 1961 Apprentice Contest, the General Manager stated that the Chairman of the Education Division had agreed to the following: winning patterns and castings to be displayed at Detroit in 1962 instead of at San Francisco 1961; 1st Prize winners to be taken at AFS expense to San Francisco, if they insist, but suggestion being made that they attend instead the International Foundry Congress in 1962; all 1961 cash awards and certificates to be distributed as soon as possible after the contest judging on April 7, 1961. On suggestion, it was agreed that a poster listing all 1961 Contest winners would be displayed at San Francisco.

*Deficit Financing.* The Finance Committee, the President said, was unanimously in favor of ending deficit financing of AFS by June 30, 1961, and henceforth budgeting and financing each 2-year budgetary period on a pay-as-you-go basis. To accomplish the first step, the Finance Committee recommended sale of AFS securities sufficient to liquidate existing bank loans and to provide working capital for balance of 1960-61 year.

Following discussion, motion was made, seconded and carried (3 Directors dissenting), as follows:

> Authorization is hereby granted, as recommended by the Finance Committee, for the withdrawal of approximately $153,000 from the AFS Investment Agent Trust Account by the sale of sufficient securities to liquidate existing bank loans of the Society and to provide working capital, so as to conclude the Society's fiscal year to June 30, 1961, without outstanding deficits.

*Existing Bank Loan.* The President requested Board approval of Finance Committee action for renewal of a $100,000 existing bank loan to March 31, 1961 (if necessary), until funds from the sale of securities authorized become available. Approved on motion made, seconded and carried.

*Working Capital.* The President requested Board approval of Finance Committee action authorizing the negotiation of a second bank loan of $10,000 (if necessary), until funds from the sale of securities authorized are available. Approved on motion made, seconded and carried.

*Convention and Exhibit Locations.* In order to insure maximum attendance at future Conventions and Exhibits of the Society, and to conserve Society resources, Finance Committee recommendation was approved by the Board of Directors on motion made, seconded and unanimously carried, as follows:

> The location of future AFS Castings Congresses and Expositions should henceforth be decided by the Board of Directors in strict terms of Society financing and estimated attendance. While AFS should at all times welcome invitations from the Chapters and Membership to hold these events in specific localities, all such invitations should be referred to the Finance Committee for study and recommendations prior to action by the Board of Directors.

*Chapter Officers Conference.* In compliance with recommendation of the Executive Committee and Chapter officers attending Regional Administration Meetings, the Finance Committee recommended that the 1961 Chapter Officers Conference be held on the following basis: One delegate per Chapter at full AFS expense, as in 1960; all other delegates to be invoiced for prorata share of expenses while at the Conference, other expenses to be assumed by delegates, their Chapters or their companies; delegates to receive an advance estimate of per diem expenses for Conference attendance. It was the consenus of the Board that more precise definitions of acceptable Conference expenses be developed and provided to Chapter delegates in advance, as a guide in submitting expense accounts.

*Non-Exhibit Conventions.* The question of holding future non-Exhibit AFS Conventions was discussed, it being the consensus that AFS as a technical Society is obligated to provide annually a vehicle for the presentation of technical advancements, whether or not in conjunction with an AFS Exposition.

*1962 Exhibit Space Rentals.* In keeping with expected unusual expense of sponsoring an International Foundry Congress in 1962, the Finance Committee recommended that space rental rates for the 1962 AFS Exposition in Detroit be $4.50 per square foot average. Approved by the Board of Directors on motion made, seconded and unanimously carried.

The General Manager reported on negotiations to develop proper labor conditions and service charges at the new Cobo Hall, Detroit, where the 1962 Congress and Exposition are scheduled. He stated that no contract would be signed unless the formalized Exhibit "Conditions" developed by AFS are approved and accepted by all parties concerned, and that every effort was being made to assure the Society's exhibitors favorable conditions.

It was the consensus of the Board that special effort should be made to attract exhibits of die casting equipment in 1962, and that the foundry trade associations who are members of the National Castings Council should be offered exhibit space gratis.

*Member Dues Schedule.* The Finance Committee pointed out that the total of Net Expense items (Technical Activities, General and Special Publications, Educational Activities, Chapter Operations, Safety Hygiene & Air Pollution Control Program, and General Administration) currently exceeds total Dues Income by approximately $100,000 annually. The Committee also stated that Net Expense per Member was $30.60 (2-yr. average 1958-60), whereas Net Income per Member was only $22.98 in the same period.

The Committee called attention to the fact that the last four dues increases became effective on July 1 of 1944, 1945 and 1948, and on Dec. 1, 1953, and that operating costs of AFS and all industry have materially increased since 1953. The effects of the last four dues increases on total Membership and Dues Income were shown as follows:

| | Total Membership | | Total Dues Income | |
|---|---|---|---|---|
| Date of Dues Increase | On Eff. Date of Incr. | 3 Yrs. Later | On Eff. Date of Incr. | 3 Yrs. Later |
| July 1, 1944 | 6,620 | 9,683 | $ 87,700. | $158,400. |
| July 1, 1945 | 7,815 | 10,403 | $103,300. | $167,100. |
| July 1, 1948 | 10,403 | 9,220[1] | $167,100. | $205,500. |
| July 1, 1954 | 11,551[2] | 13,340 | $231,200. | $308,300. |

[1]Membership June 30, 1951 – 9,220; on June 30, 1952 – 10,033.
[2]Effect of Dec. 1, 1953 increase not fully operative until June 30, 1954.

The Finance Committee expressed the belief that the Society's future progress, coupled with increased operating costs of the past seven years and the probability of further inflationary trends ahead, indicated the advisability of an early dues increase. The Committee therefore recommended a new Dues schedule effective July 1, 1961, including a new "Research Patron" class of membership, as authorized under the by-laws of the Society. As recommended, Research Patron member dues in excess of Sustaining member dues would be earmarked specifically for Research purposes.

The new Dues schedule recommended by the Finance Committee, in comparison with current dues:

## AFS Dues Schedule — Effective July 1, 1961

| Member Class | Current Annual Dues, $ | Annual Dues Recom. Eff. July 1, 1961, $ |
|---|---|---|
| Research Patron (new) | — | 1000.00 |
| Sustaining | 300.00 min. | 400.00 |
| Company | 100.00 | 150.00 |
| Personal | 20.00 | 30.00 |
| Affiliate | 10.00 | 15.00 |
| Associate | 10.00 | 15.00 |
| Junior | 4.00 | 5.00 |

Following full discussion, the new Dues rates recommended to become effective July 1, 1961, were approved by the Board of Directors on motion made, seconded and carried (5 Directors dissenting). At the same time, Board approval incorporated the following conditions:

(a) Research Patron dues in excess of $400 annually to be earmarked for AFS Research projects.

(b) Employees of Research Patron members, residing anywhere in Continental North America, to be permitted to hold Personal-Affiliate membership with minimum annual dues $15.

(c) Dues Refunds on Research Patron memberships to be made only on first $400 of dues paid.

(d) Sustaining member dues to be flat $400 annually, not $400 minimum.

(e) Employees of Sustaining and Company members to be permitted to hold Personal-Affiliate membership with minimum annual dues $15, with residential limitations prescribed in AFS by-laws as amended July 1, 1958.

(f) Sustaining and Company members to continue to receive the annual Bound Volume of "Transactions" gratis, on request.

(g) No limitations to be made on number of Affiliate members permissible for Research Patron, Sustaining and Company members.

(h) Dues Refunds to be made to Chapters as a straight 20 percent of dues applicable. Refunds on Research Patron (up to $400) and Sustaining member dues to be pro-rated to Chapters on the basis of number of Affiliate members per Chapter for each Research Patron and Sustaining member.

It was the consensus of the Board that the next AFS by-laws amendment should give serious consideration to (1) removing present territorial restrictions on the holding of Affiliate memberships at minimum dues; and (2) changing qualifications of Junior members to apply strictly to students, apprentices or trainees, regardless of age. The General Manager pointed out that while the by-laws authorize the Board to determine member dues, membership qualifications and privileges can only be changed by means of a letter ballot of the voting membership.

President Dunbeck thanked the Directors for their consideration of a knotty problem, and stated that he fully appreciated the reasons why some Directors felt impelled to withhold approval on increasing dues under present business conditions. He expressed the Finance Committee belief, however, that future AFS progress is dependent on establishing sound policies of financing and control of expenditures, and that the 1961 dues increase is an integral part of this financial planning. As soon as possible, he said, announcement of the new dues structure will be made to the membership. He then expressed confidence that all Board members would support the Board action in discussions with Chapters and individual members.

### Society of Die Casting Engineers

President Dunbeck reported discussions held with the President and Vice-President of the Society of Die Casting Engineers and the AFS General Manager, relating to possible affiliation of AFS and S.D.C.E. He stated that AFS had made positive proposals relating to joint membership, Chapter activities, Board representation, technical affiliation, convention and exhibit programs, and a professional Staff to advance die casting interests. Complete proposals on the advantages of amalgamation had been forwarded to S.D.C.E. officials, he said, for consideration by that society's Board of Directors. It was the consensus that AFS officials should continue to seek affiliation with the die casting society in view of common interests, activities and purposes.

### Reduction of Exhibit Costs

The General Manager reported that while all returns were in from the four questionnaires distributed following the 1960 Convention & Exhibit in Philadelphia, final analysis had not been completed but was scheduled for prompt conclusion and distribution.

The General Manager also reported on steps taken to reduce the cost of exhibiting at future AFS Expositions by development of formalized rules and regulations to govern service charges, service labor and other exhibit requirements. He stated that rules and regulations developed have been widely distributed to exhibitors, exhibit halls, convention bureaus, service organizations and exhibit sponsors and have been widely acclaimed as a "pioneering" approach to the problem. The new rules, he said, would be enforced for the first time at the 1962 AFS Exposition in Detroit. (These Rules and Regulations are shown as Appendices "B," "C" and "D.")

### Selection of Auditor

Recommendation of the Finance Committee for appointment of George V. Rountree & Co., certified public accountants, to audit the Society's financial accounts as of June 30, 1961, was approved by the Board of Directors on motion made, seconded and carried.

### AFS Administrative Film

The President pointed out that his company, International Minerals & Chemical Corp., had financed preparation of the AFS film "What's in It for You?", depicting the President and General Manager and shown to the Board the evening of Feb. 20. The film is to be shown at all AFS Chapters and Chapter reactions obtained as a guide to consider in 1961-62 budgets. (Cost of film $6500, time 10 minutes, color and sound.)

### New AFS Student Chapter

The General Manager announced that a petition was momentarily expected for establishing the 14th AFS Student Chapter at Ryerson Institute of Technology, Toronto, Ont. The petition, he said, would be signed initially by 22 students to become Junior members; and the new group was being formed in accordance with AFS Student Chapter policy and with the backing of the Ontario Chapter. In order to expedite formation of the group, the General Manager recommended Board approval of the Ryerson Institute Student Chapter petition when received and in keeping with AFS policy. On motion made, seconded and unanimously carried, recommendation approved.

### Announcement of Next Board Meeting

President Dunbeck announced that the next meeting of the Board of Directors would be held May 9, 1961, at the Sheraton-Palace Hotel in San Francisco, and requested all Officers and Directors to attend if at all possible.

There being no further business to be considered, the meeting was declared adjourned.

Respectfully submitted,
WM. W. MALONEY
*General Manager*

APPROVED:
N. J. DUNBECK, *President*
March 24, 1961

# APPENDIX "A"

## Report of Board Membership Committee to AFS Board Meeting Feb. 20-21, 1961

Meeting of the Membership Committee was held Feb. 20 at Sea Island, Ga., with 6 of 8 members attending. The objective of the committee, to stimulate Company and Sustaining Memberships in the Society, was reviewed.

Chairman Amrhein reported on results of a survey taken during the early part of February, each Regional Vice-President being asked for an up-to-date report on activities in their respective areas. From discussion it appeared that the organization-

al setup is functioning in each Chapter but showed that the degree of effort and enthusiasm varies from passive activity to Chapters agressively pursuing Company and Sustaining Memberships. It was obvious that best results are coming from Chapters that have established a high-level select committee made up of alumni and prominent foundry executives, organized for the single purpose of making personal contacts with the proper men in management of non-member companies.

Over-all results of efforts to date are frankly discouraging to the Committee. However, the impact of economic conditions during the past eight to ten months cannot be ignored. Many companies are facing the same sort of financial circumstances as AFS has been deliberating, and are looking for ways and means to cut expenses, with the result that all types of association memberships are vulnerable.

The Committee agreed that, despite discouraging results to date, it is necessary that its efforts be redoubled and that we continue to move ahead with the existing program. It was decided to write each Chapter Chairman again to encourage continued effort, with particular emphasis on forming select committees to promptly investigate and attempt to forestall or resell any doubtful Company or Sustaining member. It was further recommended that the Chapters make special effort to perpetuate their select committees for another year so as to assure continuity of effort, since it is always difficult to get new committees activated and moving promptly.

The Committee gave careful consideration to the dues increase proposed by the Finance Committee. While it is fearful of the impact of the increases on borderline memberships, and the probable resulting handicap in attracting new memberships, the Committee expressed itself as generally in favor of the proposed dues increase.

However, the Committee went on record that it is in disagreement with the existing member classifications and dues structure, especially as applied to Company memberships, and strongly urges the Executive Committee to carefully examine the definitions of each member class with particular emphasis on how many Affiliate members should be allowed Company and Sustaining members.

It is also the consensus of the Committee that there is presently too great a dues spread between the Personal and Company member classes, creating a problem for small jobbing foundries and pattern shops who, while willing to recognize that a Personal membership does not reflect proper support of AFS, cannot justify the cost of Company membership.

It would seem advisable, the Committee feels, for AFS to establish one or two additional classifications of Company memberships on a graduated dues basis, with the specific understanding that all companies must hold Company membership according to their size of operation, and not be eligible to hold Personal membership, the Personal class being limited to individuals only.

The Committee also went on record that when changes in policy at the Central Office occur, such moves should be properly explained so as not to be misinterpreted. To do otherwise puts the Society in a bad light when it approaches these members or prospective members to sell the need for increased dues and revenue.

When it is realized that less than 50 percent of the foundries in the United States now hold membership in AFS, the Committee recognizes that there is fertile territory to be explored and, therefore, is willing to accept the challenge of trying to "put the foundry back into AFS."

Respectfully submitted,
BOARD MEMBERSHIP COMMITTEE
NORMAN N. AMRHEIN, *Chairman*

# APPENDIX "B"

## Statement of Policy

## Governing the Holding and Servicing of AFS Expositions

The American Foundrymen's Society (AFS) is the international technical society for the entire Metal Casting Industry. Its sole purpose is to advance the arts and science of metal casting, exclusively through education, research and scientific means.

Since 1915 AFS has sponsored the important Industrial Exposition known as the "Foundry Show," which consistently occupies nearly 150,000 sq. ft. of floor space. These Expositions are among the largest, most complex and most heavily powered Industrial Expositions held in the United States.

The American Industrial Exposition is in large degree responsible for the superiority of American efficiency and productivity. Hundreds of thousands of new and improved materials, methods, processes and mechanical devices have been introduced at such Expositions, encouraging the acquisition of billions of dollars of capital investment and stimulating adoption of countless labor-saving, cost-saving and safe-practice methods. The resulting flood of quality consumer goods, priced for maximum market, is the direct cause of America's high standard of living.

In recognition of the economic value of Industrial Expositions, facilities for these events have been tremendously improved during the past 30 years. In recent years, still finer facilities have been developed (or are now being developed) in at least 12 major U. S. Cities:

| | |
|---|---|
| Atlanta, Ga. | Las Vegas, Nev. |
| Atlantic City, N. J. | Los Angeles, Calif. |
| Boston, Mass. | Miami, Fla. |
| Chicago, Ill. | New York, N. Y. |
| Dallas, Texas | Pittsburgh, Pa. |
| Detroit, Mich. | San Francisco, Calif. |

The Convention-Exhibit facilities in these 12 cities are practically all municipally-financed and municipally-owned . . . with a total investment of at least 2 billion dollars. This tremendous investment could not be justified in terms of "civic pride" . . . it was made for one purpose, to bring to each community a larger share of the 3 billion dollars spent annually at American Conventions and Exhibits.

Today these investments face an uncertain future because the American Industrial Exposition is being challenged by those who have made it possible: the Exhibitors themselves.

Over the past 15 years the cost of exhibiting at Industrial Expositions has constantly increased to a point where most Exhibitors now consider it an accepted principle that they will be overcharged for practically every service they require. Those who recognize the value of the Industrial Exposition accept this "principle" and continue to exhibit in spite of excessive costs; those who do not, may exhibit once or twice and thereafter refrain, or else fear the rumors (not unfounded) of high costs and do not exhibit at all.

Unless the constantly growing cost of exhibiting can be brought under reasonable control, those who sponsor or service Industrial Expositions can expect two actions by Exhibitors and by organized Exhibitor Groups: Fewer quality operating and dramatic exhibits, the backbone and main function of an Exposition; and increasing demands for holding Expositions less frequently. Neither probability will be good for Industry, for the Exhibitors and their customers, for Exhibit Halls, Convention Bureaus, Service Organizations, Sponsoring Organizations, nor for the many local enterprises (hotels, restaurants, air and rail lines, stores, taxi services, etc.) which look to frequent Conventions and Exhibits for much revenue.

The problem is complex, but in final analysis is the direct result of an almost universal policy of charging "all the traffic will bear." Little or no control of this condition has been exercised by those in a position to do so. Sponsoring Organizations accept and publish basic charges but make no attempt to control their application. Exclusively appointed Service Organizations assume little or no responsibility for the quality of labor provided, and exercise no control over work standards or performances.

Some Exhibit Hall managements maintain only a skeleton supervisory force which has neither the authority nor the desire to control actions of union labor or their employers, the Service Organizations. In some Exhibit Halls the effect of published basic service rates is almost completely negated by practices of the labor unions, operating as they see fit and with the employing Service Organizations "playing ball."

The net result is that Exhibitors at many expositions, caught in a "squeeze" by the necessity of completing installations for Opening Day, will approve service charges regardless of cost, even those clearly excessive or even outrageous. The aftermath, however, is caustic criticism of all parties who failed to protect their interests as "customers" of those who sponsored and serviced the event.

Causes of the current appalling situation are many . . . some obvious, some difficult to detect, and some the result of failure to exercise normal business obligations. Among the more obvious causes of excessive exhibit costs are the following:

(1) Failure of Exhibitors to provide advance data on their actual requirements . . . essential to minimizing costs. The remarkable efficiency of most Service Organizations in meeting last-minute Exhibitor demands is usually (and properly) reflected in high installation costs from overtime labor.

(2) Excessive "profit markups" in Service Charges for labor, equipment and other services. All too frequently, Service Charges are based on noncompetitive agreements instead of actual operating costs.

(3) Abuse of "Time and Materials" provisions in most Service contracts, in lieu of accurate, reasonable, advance cost estimates. Many exhibit jobs cannot be accurately estimated, but on many jobs there is no justification in quoting a catch-all "time and materials" basis.

(4) Lack of work standards and performance standards for exhibit labor. Relatively few Service Organizations maintain sufficient labor and supervision of their own to service all Expositions. Transient labor obtained from union "hiring halls" is frequently incompetent, unruly and indifferent (or rebellious) to supervision of production and performance.

(5) Abuse of Overtime Rate provisions during installation. The tendency of many exhibit workers to stretch out daytime work in order to make overtime wages is a familiar practice at most Expositions today. So too is the shortage of several trades on week days . . . and their eager availability on Saturdays and Sundays.

(6) Failure of local officials to prevent work stoppages due to jurisdictional labor disputes. It is inconceivable that a community with an exhibit property costing millions cannot or will not negotiate lasting (or even temporary) agreements governing the jurisdictions of exhibit service labor.

(7) Tendency of some Exhibitors to "buy" special treatment or consideration from both labor and Service Organizations. Exhibitors who believe (as many do) that the only way to get their exhibits installed on time is to "pay off" create an uncontrollable situation that results in higher costs for every Exhibitor.

(8) Failure of Sponsoring Organizations to act promptly on behalf of Exhibitors when special conditions arise. Mistakenly, or deliberately, some Exhibit Managers fail to assist Exhibitors in negotiating special or last-minute requirements at fair or reasonable cost.

(9) Deliberate "padding" of Service Charges for labor, materials and use of equipment. Fortunately, such instances are few indeed and soon become widely known. Equally well known is the fact that the practice does exist.

From the above examples of existing conditions it is obvious that exhibit abuses and excessive costs have no single source. They exist because all parties to the Industrial Exposition have failed to see (or have ignored) their basic obligations and responsibilities. Without exception all parties concerned . . . Sponsor Organizations, Service Organizations, Labor Unions, Exhibitors, Exhibit Halls, Convention Bureaus, and Civic Officials . . . have accepted the development of conditions and practices which threaten the American Industrial Exposition as we have known it.

Restoration and maintenance of reasonable exhibit costs *is not impossible*. It is dependent, however, on three fundamental points:

*First*—Acceptance of full responsibility by all parties concerned to create and maintain a favorable atmosphere of conditions, costs and fair dealing for Industrial Conventions and Expositions.

*Second*—Establishment and acceptance of primary obligations by those who Sponsor and Service the Industrial Convention and Exposition.

*Third*—Development of a system for daily control and adjustment of all exhibit Service Charges as they occur.

In the belief that these three fundamentals must and can be developed if exhibit costs are to be controlled, the American Foundrymen's Society has adopted a program of Rules and Regulations for the Conduct and Servicing of Expositions of the Metal Casting Industry which this Society sponsors.

This program is the result of extensive discussions with Service Organizations, Exhibitors, Exhibit Halls, Convention Bureaus and others who recognize the value of the Industrial Exposition. These Rules and Regulations will govern the Society's decisions on the holding of all future Conventions and Expositions of the Society, and the appointment of all Service Organizations necessary to their installation, operation and removal.

AMERICAN FOUNDRYMEN'S SOCIETY
WM. W. MALONEY
*General Manager*

# APPENDIX "C"

## Rules and Regulations for the Conduct and Servicing of Expositions of the Metal Casting Industry

This Society emphasizes strongly, to all Service Organizations considered for appointment, the following obligations and responsibilities of all parties involved in Expositions sponsored by this Society, in order of importance and precedence of obligations as indicated:

*First.* Service Organizations appointed by this Society have a primary obligation to the Society rather than to individual Exhibitors. This means that any Service Charge made against any Exhibitor at an AFS Exposition must first be satisfactory to and approved by the Society before such charges may be invoiced to Exhibitors. Once Service Charges have been approved by the Society for invoicing to Exhibitors, the Society will stand back of the invoiced amounts and make certain that payment is made to the Service Organization involved.

*Second.* The obligation of appointed Service Organizations to Exhibitors at Expositions of this Society is of a secondary nature. This means, in the event of any dispute over Service Charges between Service Organizations and individual Exhibitors, that the Society will consider that its own interests are primary and will at any and all times feel free to enter such disputes in order to negotiate settlements agreeable to the parties concerned.

Before executing any contract for holding a Convention and Exposition of this Society, the Society will require similar understanding and assurance, in writing, that the satisfaction and approval of the Society is accepted as the primary obligation of the following: the Convention and Exhibit Hall; the local Convention and Visitors Bureau; the local Hotels to which the Society's Convention and Exposition visitors are assigned; in the case of municipally-owned Convention and Exhibit Halls, the local City administration, through authorized City officials; and the responsible heads of the several Labor Unions whose memberships will be expected to service the Convention and Exposition through employment by the appointed Service Organizations.

This Society will consider it the direct obligation and re-

sponsibility of the local Convention and Visitors Bureau to obtain, from the City administration and Labor Unions involved, the acceptance of their primary obligations to the Society, which the Society requires.

*Wage Increases.* Service Charges quoted by Service Organizations for AFS Expositions must be based on current wage scales in the scheduled Convention and Exposition city, not on anticipated future wage increases. When it becomes known that wage increases for Service Organization labor will occur, between the date of Service agreement with the Society and the date of the scheduled Convention and Exposition, new schedules of Service Charges must be filed promptly with the Society so that adequate notice may be given by the Society to all Exhibitors.

Any new schedule of Service Charges so filed with the Society must be supported by full evidence showing old and new wage scales, date on which the new schedules will become effective, their effect on straight-time and overtime rates, and any changes in working conditions or "fringe" benefits that may have a bearing on pricing and servicing the scheduled event.

*Labor Markup.* This Society believes that any appointed Service Organization is entitled to a fair and reasonable profit on Services provided. To this end, the Society will accept from Service Organizations a markup not to exceed 56 per cent of actual Service Labor straight-time wage scales, and such markup will include ALL of the following:

Any and all "Fringe" Benefits applicable (such as Welfare Fund, Pension Fund, etc.).
Adequate Supervisory personnel.
Any and all Standby Time normally charged.
All General Overhead charges (such as Social Security, Workmen's Compensation, Unemployment Insurance, Public Liability Insurance, etc.).
Any and all Special Charges normally made.
Profit margin.

This Society will not accept any markup on straight-time Labor by Service Organizations exceeding the above-mentioned overall 56 per cent of actual Service Labor wage scales. In the event a Service Organization requests an overall markup exceeding 56 per cent, an itemized breakdown of intended Service Charges must be filed with the Society by the Service Organization in order to be considered for appointment.

*Overtime Markup.* This Society will not accept Service Charges for Overtime Labor which include more than 56 per cent overall markup on Straight-Time Labor wage scales, unless the Service Organization is willing and able to justify a higher markup with itemized cost breakdowns furnished by the Society.

*Examples:* (1) Carpenter wage scale is $5.00 per hour straight time . . . Service Charge may not exceed $7.80, including maximum overall markup of 56 per cent over wage scale. (2) Time-and-one-half wage scale is $7.50 per hour . . . Service Charge may not exceed $7.80 straight time (including 56 per cent overall markup), plus $2.50 additional wages actually paid, or total Service Charge $10.30. (3) Double-time wage scale is $10.00 . . . Service Charge may not exceed $7.80 straight time (including 56 per cent overall markup), plus $5.00 additional wages actually paid, or total Service Charge $12.80.

The Society emphasizes the above as examples only of the basis on which the Society will require that Service Charges on all Service Labor be developed and quoted. The Society is fully aware that proper cost accounting of a Service Organization may involve additional reasonable charges (beyond the 56 per cent maximum overall markup) because of special operating conditions. In such event the Society will require that the Service Organization, prior to appointment, inform the Society of intended additional charges and the basis on which determined.

*Service Charge Examples.* This Society will not accept Service Charges for Overtime Labor *based* on other than actual overtime wage scales in effect, as indicated in the following Examples:

*Example A:* Straight-time scale in the Community normally involves 8 hours consecutive work Monday-Friday, with 1½-time for the first 8 hours after 4:30 pm daily, double time thereafter. Service Charges will be based on actual work-pay conditions in effect for each man employed.

*Example B:* Rigger wage scale is $6.00 per hour straight time. $9.00 per hour 1½-time from 4:30 pm to midnight Monday-Friday, $12.00 per hour double time thereafter. Under such a local wage-scale agreement, Service Charges will be based on actual wages paid and will not include "automatic" double time for work between 4:30 pm and midnight when only 1½-time wages are paid.

*Example C:* Painter wage scale is $4.60 per hour straight time, $6.90 per hour 1½-time on Saturdays 8:00 am to 12:00 noon, $9.20 per hour double time thereafter. Service Charges will be based on actual wages paid and will not include "automatic" double time for Saturday work between 8:00 am and 12:00 noon when only 1½-time wages are paid.

*Example D:* Carpenter "D" works 8 hours (8:00 am-4:30 pm) at $5.00 per hour straight time. If "D" continues work beyond 8 hours in any 24-hour period, he is entitled to the overtime wage scale in effect, and Service Charges will be based on actual wages paid. If Carpenter "D" lays off after 8 hours and is replaced by Carpenter "E," Service Charges for "E" will be based on actual wage scale that "E" is paid, regardless of the hours that "E" works.

Where "E" commences work at 4:30 pm on straight time, plus a "bonus" or "nightwork" per cent, Service Charges may include such "bonus" per cent. The 56 per cent overall markup used in determining the proper Service Charge will be applied only to the straight-time wages paid "E," after which the "bonus" per cent that "E" actually receives may be added.

*NOTE:* It is obviously impossible for the Society to know or to specify exact Overhead and other charges that should be used by Service Organizations in developing Service Charges acceptable to this Society. Rather, the 56 per cent maximum overall markup on straight-time wage scales, and the examples presented herein, are intended solely to indicate a fair and reasonable principle on which the Society believes all Service Charges should be developed. Where abnormal, unusual or special conditions exist, which a Service Organization requests be considered in establishing acceptable Service Charges, the Society will require that itemized supporting data be furnished by the Service Organization in order to be considered for appointment.

*Service Invoices.* This Society requires all appointed Service Organizations, without exception, to furnish the Society promptly a copy of any and all invoices rendered to Exhibitors at AFS Expositions. The Society further requires that all such invoices be itemized as to all Labor, Materials and other Services rendered, and the charge for each itemization be shown.

*Examples:* It will be sufficient for a Service Organization, in itemizing charges on invoices, to show the following:

*Power Connections*—sizes and capacities; rates as published.

*Materials*—all Materials itemized; cost of each item.

*Installation Labor*—total hours worked; rates per hour as published; total labor charge.

*Dismantling & Removal Labor*—total hours worked; rates per hour as published; total labor charge.

*Handling Charges*—actual tonnage (weighed); cwt. rate as published; total charge. (Separately for Installation, and for Dismantling & Removal.)

*Equipment Rentals*—no. and kind of equipment; rental rates per item as published; total rental charges.

*Special Services*—itemized by clear description of the service performed; hours worked or materials required (as applicable); agreed or published price; total charge.

*Cartage Requirements.* Service Organizations appointed by this Society for the movement of Exhibit Materials and Equipment into and out of Exhibit Halls will be expected to agree to the following conditions in return for exclusive appointment:

(1) No Special charge shall be made against any AFS Exhibitor whose freight is inadvertently shipped to an AFS Exposition city without prepayment of shipping charges. Where prompt reimbursement for such shipping charges is required, the Service Organization may invoice the Society prior to the

Exposition opening, and the Society in turn will collect from the Exhibitor.

(2) No overtime rates of any kind shall be charged for unloading shipments arriving at the Exhibit Hall where an AFS Exposition is to be held, regardless of hour of arrival and unloading, and regardless of whose trucks such shipments arrive in. Installation of AFS Expositions require 10 to 21 days, depending on location and conditions, and Installation work during all except the last several days before Show opening will commence at 8:00 am and end at 4:30 pm daily. Where very large exhibits of equipment are delayed in arrival, necessitating late hours of delivery, unloading and setup, overtime unloading charges may be accepted at the sole discretion of the Society's Exhibit Manager.

(3) No service or service charge discrimination shall be made against any Exhibitor materials arriving at the Exhibit Hall by truck, regardless of whose trucks such shipments arrive in. The Service Organization will be expected to provide the same degree and quality of labor and equipment for prompt unloading of such shipments as for shipments arriving on the Service Organization's own trucks.

(4) All materials and equipment of Exhibitors handled from the Exhibit Hall unloading docks, to booths and return, shall be charged for at the same rate per ton or hundredweight agreed to and published by the Society, regardless of the type of materials and equipment moved. It is understood that agreed and published handling rates will cover crates, machinery crated, machinery on skids, or free machinery and parts, etc.

(5) The Service Organization will provide all necessary Labor and Handling Equipment designated by the Society, and such Labor and Equipment shall be available and usable between the hours of 8:00 am and 4:30 pm daily throughout the entire period of Installation and Removal. It will be the joint responsibility of the Service Organization Manager and the AFS Exhibit Manager to determine daily the requirements of each following day during both Installation and Removal.

(6) The Service Organization shall at no time charge Exhibitors overtime rates for the use of Handling Equipment (such as fork lifts, cranes, etc.) with Operator, regardless of the hour or day when used. It is understood that the agreed and published rates for such Equipment and Operator will cover all time contingencies. During the Installation period, prior to the 3 days preceding Opening Day, installation work will be performed only between 8:00 am and 4:30 pm, unless unusual circumstances arise. Failure of the Service Organization to unload waiting trucks by 4:30 pm shall not result in overtime charges against Exhibitors, unless such overtime work and charges are authorized by the AFS Exhibit Manager.

(7) Where shipments arrive at the Exhibit Hall on Exhibitors' own or contract trucks, under instructions of Exhibitors to place such shipments directly in assigned booths, the Service Organization shall provide such equipment and labor as may be required to assist in performing the work. Such labor and equipment may be charged against Exhibitors in accordance with agreed and published schedules of charges.

(8) The Service Organization shall provide the Society, upon notice by the AFS Exhibit Manager given one day for delivery not later than 8:00 am the next day, one or more fork-lift trucks together with competent Operator(s), for exclusively scheduled assignment and use by the Society for minimum periods of 8 hours. Such "pool" Equipment and Operator shall be charged direct to the Society for specified periods at agreed and published hourly rates, and the Society alone will be responsible for any charges made against Exhibitors serviced under such "pool" arrangement.

*Special Services.* Service Organizations appointed to provide such Exhibitor services as Floor Studding or Drilling, Excavating Pits, Concrete Anchors, Breakouts at close of Exposition, and re-Cementing Concrete floors as called for in Exhibit Hall contracts, shall furnish to the Society in advance a Price List covering all such work as the Society may indicate will be required. Such Price List, approved and published by the Society, will govern all charges made by the Service Organization against Exhibitors.

All such Price Lists shall include the furnishing and use of any tools or equipment necessary, and all Labor and supervision of same.

In the event the Service Organization is unable to complete the required installations in locations designated by the Society and or an Exhibitor, due to lack of accurate information on accessible locations provided by the Exhibit Hall, the Service Organization shall charge the Exhibitor only for the completed job as quoted, and shall look to the Exhibit Hall for reimbursement of any extra costs involved.

*Power Installations.* Any Service Organization appointed to install main power lines for the use of Exhibitors at an Exposition of this Society . . . such as electrical cable, air lines, gas lines, water lines, and the like . . . shall provide, in the back of each Exhibitor booth where power is required, and at a height specified by the AFS Exhibit Manager, suitable outlet connections in accordance with requirements stated by the Society in advance of the Exposition. The Service Organization shall charge the entire cost of such main lines and outlet connections direct to the Society, and shall make no charge whatsoever against Exhibitors for same.

All published rates agreed to by the Society for connecting Exhibitors' equipment to main power lines shall include all and only such Materials and Equipment as are actually installed and checked by the AFS Exhibit Manager, providing accurate and final Exhibitor requirements as to type and location of such connections are made known to the Service Organization in advance of the Exposition. Where accurate booth plans are provided in advance, the Service Organization shall furnish to the Society, on request, written estimates covering the total cost of installations, and such estimates shall govern actual charges against the Exhibitor unless special or unexpected conditions arise.

It is understood that any changes from Advance Booth Plans may entail additional costs to an Exhibitor.

*Service Tools.* All charges for Service Labor at an Exposition of this Society, as quoted by appointed Service Organizations, shall include any and all tools and equipment normally employed by such Labor in carrying on their trades, and no extra charge of any kind, not agreed to by the AFS Exhibit Manager, shall be made by the Service Organization to Exhibitors for the use of such tools and equipment.

It will be the responsibility of the Service Organization to make certain that all assigned Service Labor possesses or is provided with proper and sufficient tools and equipment to perform the work required when first checked in to an Exhibitor on a Work Order.

*Labor Supervision.* Any Service Organization appointed to provide Service Labor to an Exposition of this Society agrees to provide, at the sole expense of the Service Organization, such Labor Supervision as is necessary to assure proper Labor performance and work standards. Any deviation from this supervisory requirement will be at the sole discretion of the AFS Exhibit Manager.

*Service Work Orders.* All Work Orders for Service Labor at an Exposition of this Society must show Time-In and Time-Out for each assigned booth job, and Time-In shall be noted in the presence of the Exhibitor representative ordering such Labor. From that point to conclusion of the job, the presence of assigned Labor on the job shall be the sole responsibility of the Exhibitor.

On completion of the job specified, or at the end of the prescribed work period, Service Labor personnel shall have Work Orders signed by a qualified Exhibitor representative who shall also note on the Work Orders the correct Time-Out. For this purpose the Society will install in a prominent location a clock or time stamp which shall be accepted by all as standard equipment for determining Time-In and Time-Out.

*No Gratuities Permitted.* No Gratuity of any kind shall be accepted by any full or part-time employee of any appointed Service Organization, or by any employee of the Exhibit Hall, from any Exhibitor representative at any Exposition of this Society. All appointed Service Organizations and the Exhibit Hall management agree, in accepting this requirement, to: (1) report any violation at once to the AFS Exhibit Manager, and (2) immediately discipline or discharge the offending employee as may be requested by the Society. Any representation to an Exhibitor who offers a gratuity will be at the sole discretion of the AFS Exhibit Manager.

*Invoices to AFS.* Any Service Organization appointed to service an Exposition of this Society shall agree, if requested by the Society, to render any or all invoices for services to the Society instead of to individual Exhibitors. If so requested, the Service Organization shall render invoices, in triplicate, in the same manner and same detail as it would when invoicing Exhibitors direct. Invoices so rendered to the Society will be subject to 30 days payment and will include any cash or other discounts normally allowed, and it then becomes the sole responsibility of the Society to render its own invoices to Exhibitors and collect for same without further obligation on the part of the Service Organization.

# Instructions to AFS Exhibitors
## Essential — All Exhibitor Personnel Will Be Required to Abide by the Following Instructions!

*Arrival Date.* Each Exhibitor will be required by AFS to indicate and accept one definite installation commencement date, on which date the Exhibitor's personnel will arrive at the Exhibit Hall to commence Installation work. Selection of this date should be determined by each Exhibitor in relation to the exhibit planned, shipping dates and estimated completion time, if unnecessary overtime labor is to be avoided.

*Check-in Procedures.* Exhibitor installation personnel must check in not later than 8:00 am on assigned date for installation commencement. Personnel must check in only with the AFS Exhibit Management (NOT with Service Organizations). No service or labor will be furnished to any Exhibitor until checked in with AFS Exhibit Management.

> *NOTE*: This requirement is solely for the protection of Exhibitors. All service requirements and charges will be under control of the AFS Exhibit Management, in direct line of responsibility as follows: 1st, Service Organizations to AFS; 2d, AFS to all Exhibitors. Failure to follow this Check-in Procedure may require assignment of overtime labor, and unavoidable expense.

All Exhibitor personnel engaged in installation work will receive, at time of checking in, a Special Exhibitor Installation Badge to be worn at all times and required for entrance into the Exhibit Hall. The Badge will also provide identification for the assignment of services or labor by official Service Organizations. Badges must be surrendered prior to opening day of the Exposition.

*No "Preferred Treatment."* In accepting assignment of space by AFS, each Exhibitor agrees not to offer or give any form of gratuity (whether money, gifts, time added to work orders, etc.) to any organization or personnel associated with the Exposition. Since all services and service labor will be under direct control of the AFS Exhibit Management, for assignment as needed, no gratuities for "preferred treatment" will be permitted. Any exhibit service personnel accepting any form of gratuity will be subject to immediate discharge by the employing organization, without penalty, in line with AFS agreements with all Service Organizations.

*Advance Service Orders.* Each Exhibitor will be required to fill out and forward to AFS, not less than 45 days prior to opening day of the Exposition (or when specified in the Exhibitor Service Information Manual) all Service Order forms applicable to each Exhibitor's requirements. Failure to execute all applicable Service Order forms is a primary cause of overtime labor and high installation cost.

> *NOTE:* AFS and official Service Organizations agree to provide sufficient Labor and Equipment for all Exhibitors who cooperate in the matter of Advance Service Orders and prompt appearance of installation personnel on Assigned Dates of Arrival. In order for AFS to assure reasonable Service Charges, the full cooperation of ALL Exhibitors is essential.

*Power Requirements.* Each Exhibitor requiring Power of any kind (electricity, gas, compressed air, water and drainage), or any type and number of Power connections, will be required to furnish to AFS, not less than 45 days prior to opening day of the Exposition (or as specified in the Exhibitors Service Information Manual) an accurate scale drawing or sketch showing: (1) No. and Location of all pieces of Equipment intended for display; (2) approximate Weight of each piece of Equipment; (3) clear designation of any and all Power hookups required; (4) accurate location of each Power connection from at least the back and one side of the booth.

> *NOTE*: Accuracy of the required Booth Layout may largely determine total Service Charges for Power connections. Main Power lines, installed by AFS in advance at AFS expense, are brought to the back of each booth, with necessary outlets spotted in accordance with the required Booth Layout. All alterations of Advance Booth Layouts will be solely at Exhibitor expense.

*Approval of Charges.* Each Service Organization will be required to prepare all Work Orders, Time Slips, Delivery Orders and Invoices in triplicate. One copy will be turned over to AFS, and one given the Exhibitor . . . on the same day issued . . . and one retained by the Service Organization. These documents will constitute the only basis on which final invoices against Exhibitors will be accepted.

*Daily Cost Controls.* The AFS Exhibit Management will maintain accurate Time and Cost Sheets throughout the periods of Installation, Exposition and Removal . . . on a daily basis. All Service Costs and Charges will be recorded by an AFS Cost Department, and all time and other charges will be verified and approved daily by the AFS Exhibit Management.

*Service Complaints.* At the time completed Work Orders are signed by Exhibitor personnel, any question or complaint as to service, equipment or charges should be noted on the work order by the Exhibitor. All complaints should then be reported promptly to the AFS Exhibit Management, for immediate adjustment or settlement to the satisfaction of all concerned while all facts are clear and verifiable.

> *NOTE:* Signatures of Exhibitor representatives on work orders will be solely to determine check-in and check-out time, and will not be considered by AFS as acceptance per se of total labor hours performed, labor competence, or labor charges applicable.

*Final Invoices.* All official Service Organizations understand and have agreed that no Service Charges will be finally invoiced to Exhibitors, and no advance payments requested, without the written approval of the AFS Exhibit Management. Any dispute over services or service charges will be negotiated and settled by the AFS Exhibit Management prior to final removal of the Exhibitor involved.

AMERICAN FOUNDRYMEN'S SOCIETY
WM. W. MALONEY
*General Manager*

August 1, 1960

# APPENDIX "D"
## Operating Requirements for the Conduct and Servicing of AFS Expositions

### Basic Exhibit Operating Principle

This Society requires that all parties and persons involved in future Expositions of this Society . . . Hall Management, Convention Bureau, Service Organizations, Labor Unions and City Officials . . . shall accept and agree to the following basic Exhibit rule:

When any Exhibitor's display-booth has been completely installed in accordance with the published Rules & Regulations governing the Exposition, and is ready for the official opening of the Exposition, from that time on the right of the Exhibitor to perform any and all lawful activities within the

confines of his display-booth shall be absolute, as described herein.

This basic Exhibit principle has long been accepted without question in practically every Convention-Exhibit Hall in the United States. Construction of many new or enlarged Exhibit Halls in recent years, however, makes advisable a more precise definition of this basic rule and its practical application to the servicing of an industrial exposition.

This Society therefore will require from all parties involved in Expositions of this Society, acceptance of and adherence to the following points of application essential to the basic Exhibit principle above:

(1) The Exhibitor shall have the absolute right to arrange, rearrange and move any exhibited materials, literature, furniture, signs, supplies and any readily movable display pieces within his own booth as he may see fit, without interference of any kind.

(2) The Exhibitor shall have the absolute right to make any adjustments or readjustments of installed equipment that are normally required for operation of such equipment for purposes of demonstration and display during the Exposition.

(3) The Exhibitor shall not have the right to disconnect and reconnect any power connection after installation by an official Service Organization, without the express and advance approval of the AFS Exhibit Manager.

(4) The Exhibitor shall not have the right to move or relocate any piece of equipment after installation by an official Service Organization, without the express and advance approval of the AFS Exhibit Manager.

### Permissible Exhibitor Work

This Society will require of all official Service Organizations, prior to and as a condition of appointment, precise statements as to what Exhibitors will and will not be permitted to do with their own personnel during periods of installation and dismantling. Once such conditions have been published to Exhibitors by the Society, it becomes the full responsibility of appointed Service Organizations to maintain such conditions and prevent deviations without the express approval of the AFS Exhibit Manager.

### No Unexpected Changes

This Society will not approve, for invoicing to an Exhibitor, any Service Charge or basis of charges not advised to the Society by appointed Service Organizations in advance of the installation of Exhibits. Where Service Charges published by the Society are obviously intended to cover a complete specific service, no additional charges of any kind for such specific service will be permitted without the express approval of the AFS Exhibit Manager.

### "Minimum Hour" Applications

This Society will not permit application of "minimum hour" provisions by official Service Organization on service jobs which *obviously* will require less than the minimum time involved. Such jobs shall be called *immediately* to the attention of the AFS Exhibit Manager for "pool" service arrangement, or for exclusive handling by the AFS Exhibit Manager's personal staff. In any event, the method of handling, and of invoicing for such handling, shall be at the sole discretion of the AFS Exhibit Manager.

### Standby Labor

The Society will not permit the requirement, by any appointed Service Organization, that an Exhibitor must employ "Standby Labor" during the Exposition for constant maintenance of operating equipment, or to assure uninterrupted operation of displays, except with the express approval of the AFS Exhibit Manager. The maximum labor markup permitted by the Society is clearly intended to cover all Standby Labor necessary during installation, Exposition and dismantling.

### Adequate Service Labor

This Society will not permit official Service Organizations to require that Exhibitors employ Service Labor at a wage scale greater than that of the type of Service Labor actually necessary to perform the specific service job. For example, the Society will not permit the assignment of Carpenters, Millwrights, Riggers, etc., for service jobs normally performed by Common Labor. It will be the sole responsibility of appointed Service Organization to provide a variety of Service Labor during periods of installation and dismantling.

### Labor Union Supervision

This Society recognizes the right of service labor unions to protect, in exhibit halls utilized by the Society, the rights of their memberships with respect to types of service work performed. At the same time, the Society will not permit unnecessary or arbitrary interference by union stewards with the orderly, prompt and uninterrupted servicing of Expositions of the Society during the periods of installation and dismantling.

To this end, this Society will require that all parties and organizations, and in particular the appointed Service Organizations, accept and be prepared to enforce the following regulations governing the official activities of service labor unions and their representatives:

(1) Each separate, recognized local labor union involved in servicing an Exposition of the Society shall be entitled, on request, to one union steward on the Exposition premises during installation and dismantling periods.

(2) All union stewards on duty on the Exposition premises will be required to wear at all times an identification badge, obtainable from the AFS Exhibit Management, showing the steward's name and the union he represents.

(3) No union steward on duty on the Exposition premises will be permitted to wear the identification badge of any appointed Service Organization or any Exhibitor.

(4) Certified union stewards on duty on the Exposition premises will not be permitted to initiate work stoppages without first reporting grievances to the AFS Exhibit Manager for prompt resolution and settlement.

### Advance Payments

This Society will not permit any official Service Organization to invoice or require payment in advance, from any Exhibitor, any charges whatsoever for any Exhibit service to be performed. All invoices for Service Charges must be approved by the AFS Exhibit Manager before being rendered, and the Society will stand back of any invoice thus approved, for prompt payment by Exhibitors.

AMERICAN FOUNDRYMEN'S SOCIETY
WM. W. MALONEY
*General Manager*
December 1, 1960

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# Minutes
# Final Meeting of AFS Board of Directors 1960-61
# Sheraton Palace Hotel, San Francisco — May 9, 1961

ROLL CALL: President Norman J. Dunbeck, *presiding;* Vice-President Albert L. Hunt, Vice-President Elect John A. Wagner.

*Directors (Exp. 1961):* D. W. Boyd, J. Dee, W. L. Kammerer, C. E. Nelson, H. M. Patton, C. A. Sanders.

*Directors (Exp. 1962):* N. N. Amrhein, D. L. Colwell, E. C. Jeter, R. E. Mittlestead, A. J. Moore, W. H. Oliver, J. N. Wessel.

*Directors (Exp. 1963):* R. R. Ashley, A. E. Falk, W. C. Jeffery, T. T. Lloyd, J. T. Moore, W. E. Sicha.

*Directors Elect (Exp. 1964):* R. J. Hageboeck, D. Hall, C. J. Lonnee, M. E. Nevins, C. Ossenfort.

Staff Officers: Wm. W. Maloney, General Manager; S. C. Massari, Technical Director; A. B. Sinnett, Secretary; E. R. May, Treasurer.

Absent: Directors R. R. Deas, Jr.; D. E. Matthieu, D. E. Webster; Directors Elect D. E. Best, C. O. Schopp.

A quorum having been established, the President welcomed and introduced the newly elected Directors and invited all to participate in the Board discussions.

### Reading of Minutes

Minutes of the Board Meeting held February 20-21, 1961, having previously been approved by letter ballot of the Board, were briefed as to principal actions taken.

### Report of General Manager

The General Manager reported briefly on membership, finances, the 1961 Castings Congress, and developments for the 1962 Castings Congress & Exposition.

### Reports of Board Committees

*Chapter Contacts Committee*: Vice-President Hunt commended the Directors for an excellent job during the past year in contacting the Chapters. He emphasized the importance of this work as the prime liaison between the Board and the "grass roots" membership. He stated that the program should be continued with the same enthusiasm as during the past year.

*F. E. F. Trustees.* Directors Kammerer and Amrhein presented a written report on meeting of F. E. F. Trustees held April 18 in Cleveland. Highlights: increasing interest of companies to make an F. E. F. donation in lieu of customer Christmas gifts; 513 students enrolled in F. E. F. scholarships; 199 graduates from F. E. F. schools expected in 1961; 227 students available for summer jobs; technical institute programs progressing well; new Associate membership classification approved; 1961 College-Industry Conference excellent, featuring personal "success stories" by 14 F. E. F. graduates.

### Report on Training and Research Institute

The General Manager briefed discussions of the Trustees meeting held May 9 and reported on a recent letter from the Washington attorneys in regard to tax classification. He stated that, while the classification had not yet been approved, in the opinion of the attorneys action would be taken soon on the T&RI application and, in their opinion, it would be favorable.

The General Manager reported the resignation of Trustee L. R. Wagner and recommended that he be replaced by Joseph A. Gitzen, Chairman of the Board, Delta Oil Products Co., Milwaukee, as a Trustee for the unexpired term. On motion made, seconded and carried, J. A. Gitzen was elected T&RI Trustee.

### Society of Die Casting Engineers

The General Manager reported on continuing negotiations with the Society of Die Casting Engineers outlining the several points offered them in negotiation. Several Directors expressed interest in this endeavor and asked that they be informed by receiving a copy of proposals submitted to S.D.C.E.

It being reported that these negotiations have not been affective to date, it was suggested that effort be continued for a merger that would benefit greatly the die casting people and enhance the program of AFS.

### Recommendations of Retiring President

Before stepping down, President Dunbeck expressed his sincere appreciation for the cooperation given him by the Board of Directors and the Central Office Staff during his term of office. To the incoming 1961-62 Board he offered the following recommendations and suggestions:

(1) That the Directors maintain a vigilant attitude toward their position as a Director for AFS and fully realize their responsibilities and prestige in the Industry.

(2) That Directors should keep in mind at all times they are acting for the good of the Society and, at the same time, as liaison with the membership and that it should be their position to explain the actions of the Board to the membership as well as to receive requests and recommendations.

(3) That as Directors of the Society they should seek to establish a sound financial program for the Society, and live with it.

The President then welcomed the newly elected Directors to the AFS Board and asked their cooperation in continuing to build the Society under the purposes and policies for which it was organized.

There being no further business to be considered, the President declared the final meeting of the 1960-61 Board of Directors adjourned, and turned the gavel over to the new President, Albert L. Hunt.

Respectfully submitted,
ASHLEY B. SINNETT
*Secretary*

APPROVED:
NORMAN J. DUNBECK, *President*
July 25, 1961

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# Minutes
# Annual Meeting of
# AFS Training & Research Institute Trustees
# AFS Central Office, Des Plaines, Ill.—Dec. 14, 1960

ROLL CALL: H. Bornstein, *Chairman;* N. J. Dunbeck, L. H. Durdin, A. L. Hunt, C. E. Nelson, R. A. Oster, I. R. Wagner.

*Guests:* Jess Toth, Chairman AFS Education Division; J. A. Wagner, Vice-President Nominee.

*Staff:* Wm. W. Maloney, Secretary, S. C. Massari, Director, E. R. May, Treasurer, R. E. Betterley, Training Supervisor; A. B. Sinnett, Secretary AFS.

### Reading of Minutes

Minutes of the meeting of Trustees held May 10, 1960, were reviewed as to actions of the Trustees in regard to the T&RI program and proposed T&RI building.

### 1960 Course Review

A summary and analysis of T&RI courses to date was presented. The Director stated that cancellation of 5 courses in 1960 was due to insufficient advance registrations, in terms of course expense and caliber of instructors obtained. It was necessary to cancel one course because a proper teaching staff could not be obtained.

Regional courses, carrying the heaviest attendance to date, were conducted in Birmingham, Toronto, Indianapolis and Chattanooga. Four new courses were offered for the first time during 1960, giving a total of 15 courses for the year attended by 360 students, an average of 24 per class. During the past

four years, 1515 students from 613 separate companies have attended a total of 51 courses. Students included representation from 37 states, 4 Canadian provinces, Mexico, South America, Asia and Europe.

A list of nearly 600 companies whose employees attended courses up to April 1960 was presented and discussed. The suggestion was made that company representation be continually checked from year to year so as to maintain constant interest.

Trustee Oster emphasized strongly the current interest of Government and the labor unions in foundry training and personnel needs of the basic foundry industry. The point, he said, should be continually stressed to foundry management.

## Financial Report

Financial reports for T&RI were presented covering the fiscal year ended Dec. 20, 1960. Income & Expense statement showed the following:

| | | |
|---|---|---|
| Income from Course Fees | | $23,580.00 |
| Expense: | | |
| Course Expense – Direct | $ 8,135.51 | |
| Course Expense – Indirect | 3,485.44 | 11,620.95 |
| | | $11,959.05 |
| Research Expense | $ 9,347.68 | |
| Miscellaneous Expense | 131.63 | 9,479.31 |
| | | $ 2,479.74 |
| Other Expense: | | |
| Research – AFS Contributions | $25,000.00 | |
| Attorney Fees – AFS Contributions | 5,000.00 | $30,000.00 |
| Decrease in Fund Principal | | $27,520.26 |

Reference was made to balance sheet showing Asset of $25,000.00 for laboratory equipment donated by Northwestern University to T&RI. It was suggested and agreed that this equipment should be depreciated in the annual audit for 1959-60 because of continuing obsolescence. It was also pointed out that expense of storing this donated equipment totaled $1,209 in the past year, and that most if not all of the items might be obsolete by the time the proposed Training Center is constructed. Following discussion, no motion made, seconded and carried, the Director was authorized to dispose of any or all of this equipment for whatever can be realized by sale, so as to eliminate or appreciably reduce cost of further storage.

## Research Contributions

The AFS General Manager called attention to T&RI surplus funds totaling $29,687 on December 20, as a result of course operations and AFS donations for T&RI research. He stated that AFS anticipated substantial deficits in 1960-61 fiscal year, and indicated that if T&RI research could be financed during the six months Jan. 1-June 30 from the present T&RI surplus, rather than from AFS donations, it would be of great assistance to the Society. He emphasized, however, that the decision was solely up to the Trustees, and that AFS would fulfill its expressed obligations as mutually agreed.

Chairman Bornstein expressed the view that there seemed to be no reason why the AFS suggestion could not be accepted. Following discussion and on motion made, seconded and unanimously carried, the T&RI Trustees agreed that the current research program should be financed from existing T&RI funds until June 30, 1961. AFS officials expressed their sincere thanks for the understanding and cooperative attitude of the Trustees.

## 1961 T&RI Courses

The 1961 T&RI course program was discussed and an outline of proposed courses approved by the Course Advisory Committee was submitted to the Trustees. Suggestion was made that a survey be conducted to help establish the need for particular courses to which the Director commented that the Course Committee had given this considerable thought and felt that it represents a sufficient cross-section of the industry to enable them to set up courses without a survey.

Question was raised as to areas supplying the majority of the students. The Trustees were referred to the analysis of past courses, which pointed out the major foundry States carrying 41% of the student participation are Illinois, Wisconsin, Michigan, Ohio and Indiana, in that order . . . all States where foundry operations are heavily concentrated.

## Tax Exempt Status

The Secretary reported on T&RI tax exemption status, giving a brief history of efforts made and stating that no further action has been taken by the Internal Revenue Service on the T&RI application. He read a letter dated Nov. 15, 1960 from T&RI attorneys Lee, Toomey & Kent, stating that Treasury Dept. regulations issued Dec. 1, 1959 to govern exemption of "scientific" organizations had been withdrawn and new regulations proposed, including Sec.iv (a) stating that an organization will not qualify under Sec. 501 (c) (3) for exemption as a "scientific" organization if it . . . "performs research only for persons which are (directly or indirectly) its creators . . ."

The General Manager said that the attorneys had expressed concern over application of this section to T&RI and had discussed it with Internal Revenue, with inconclusive results. The attorneys, he said, felt that T&RI might wish to file comments or suggested revisions.

Following discussion, it was the consensus that T&RI should be willing to have its application considered on the basis of any legal regulations adopted, and should not file comments either pro or con on the new proposals. The Secretary was instructed to so notify Lee, Toomey & Kent. Once regulations are adopted, however, T&RI should then file promptly to show that such new regulations permit T&RI to fully qualify for tax exemption.

Actions of the AFS Board of Directors at the Society's Annual Board Meeting held August 11-12 were reviewed as affecting T&RI. The Director reported that the technical Divisions had requested funds totaling $65,300 for the 1960-61 fiscal year, but that this was reduced by the Board to $43,200 because of the Society's current financial situation. As of December 20, he said, approximately $13,000 of the year's budget has been committed.

## New Business

The Director opened a discussion concerning the possibility of discounts in course fees to any company sending three or more students to any one course. It was the consensus that such a policy would be detrimental in giving the impression of "wholesaling" education, contrary to all principles of a sound educational program.

## Announcement of Next Trustees Meeting

The Chairman announced the next meeting of T&RI Trustees would be held May 9 in San Francisco, during the Society's 65th Annual Castings Congress.

There being no further business to be considered, the meeting was declared adjourned.

Respectfully submitted,
WM. W. MALONEY
*Secretary*

APPROVED:
H. BORNSTEIN, *Chairman*
Dec. 24, 1960

# Minutes

## Meeting of AFS Training and Research Institute Trustees

## Sheraton-Palace Hotel, San Francisco — May 9, 1961

ROLL CALL: H. Bornstein, Chairman; N. J. Dunbeck, L. H. Durdin, A. L. Hunt, C. E. Nelson, J. A. Wagner, W. L. Kammerer, Jess Toth, Chairman, AFS Education Division.

*Absent:* I. R. Wagner, R. A. Oster.

*Staff:* W. W. Maloney, Secretary; S. C. Massari, Director: E. R. May, Treasurer; R. E. Betterley, Training Supervisor; A. B. Sinnett, Secretary AFS.

### Reading of Minutes

Minutes of the Trustees meeting held Dec. 14, 1960 were read and approved.

### Review of Training Courses

A detailed analysis of 1961 courses held to date was presented. Comparative course data 1957-1961 were presented in table form. It was pointed out that no courses had been canceled in 1961 and that recent registrations in all courses have been excellent and ahead of any previous comparable period. Average 1961 class attendance to date is 38.1 students, compared with a 5-year average of 30.7.

During 1961, Regional Chapter courses were conducted in Houston, Chattanooga, Hamilton, Ont., and Birmingham. The latter course, "Shell Molds and Cores," registered an attendance of 62.

The Director and Training Supervisor also presented graphs of 5-year course data covering National vs. Regional courses as related to distance traveled; cumulative Company and Student registration; student registrations and companies represented by years; and "repeating" companies and per cent of total registration.

### Course Promotion and Coverage

Question was raised as to how best to get the T&RI story across to top management. Discussion recognized this as a basic problem, the consensus being that it is extremely difficult to get management to read all promotional material coming through the mails. Trustee Wagner raised the question as to whether T&RI course material duplicates activities conducted in technical sessions by the foundry trade associations. It was the consensus that these sessions have in no way conflicted as yet with the T&RI course program.

### Finances

Financial reports were presented by the Treasurer covering the 3-month period ended March 20, 1961, showing Course Income from tuition fees $14,385, Course Expense $7,518, and Net Income from Courses $6,866; Research Expense totaled $13,641 and Miscellaneous Expense $363, representing a decrease of $7,138 in T&RI Fund Principal since Dec. 21, 1960. Balance Sheet showed $41,090 Cash on hand as of March 20.

In response to question, it was stated that foundry equipment donated by Northwestern University is still being carried on the books at $25,000 valuation, but that every effort is being made to sell those items not intended for use in the proposed building.

### New Business

The resignation of Trustee I. R. Wagner was presented and regretfully accepted, with unanimous vote of appreciation for his generous services to T&RI over a period of five years. Joseph A. Gitzen, Chairman of the Board, Delta Oil Products Co., Milwaukee, was nominated to succeed Mr. Wagner. On motion made, seconded and carried, Mr. Gitzen was unanimously approved, subject to election by the Board of Directors.

The following officers were elected for terms of one year, on motion made, seconded and carried unanimously: L. H. Durdin, Chairman of Trustees; S. C. Massari, Director; E. R. May, Treasurer, and W. W. Maloney, Secretary. A unanimous vote of thanks was extended to outgoing Chairman Hyman Bornstein for his conscientious and continuous effort on behalf of T&RI over the past five years.

### Building Program

The Secretary reported on the tax exemption status of T&RI, stating that the Internal Revenue Service had requested additional specific data in March, which were provided through the Institute attorneys in Washington (Lee, Toomey & Kent). A letter dated May 5 from the attorneys was read expressing the belief that a final and favorable ruling on tax exemption would be made in the near future. The Secretary suggested, in view of estimated imminence of a favorable ruling, that the Trustees review the building program toward the end of a possible recommendation to the AFS Board, contingent on favorable ruling by Internal Revenue.

The Secretary pointed out that no construction of the proposed building can proceed without approval by the full AFS Board, and that such Board consideration necessarily will hinge on T&RI tax exemption and Trustees' recommendations. To date, he said, AFS has expended approx. $77,000 on the building project: $45,000 for land essential for off-street parking and a building permit; $27,000 architect fees for plans and detailed specifications, now completed and approved; and $5,000 attorney fees (partial) in seeking tax exemption.

It was the consensus that no Trustee recommendations on the building should be developed until favorable tax exemption ruling has been received. Because such ruling now is anticipated, it was agreed to call a special meeting of the Trustees on July 19, just preceding the AFS Finance Committee meeting July 20-21, in the event favorable ruling by Internal Revenue is received in the interim. Retiring chairman Bornstein agreed to attend and give any assistance he can. It was also agreed that past Trustee I. R. Wagner should be invited because of his continuous service since T&RI was established.

It was agreed that the Secretary would meanwhile prepare, for information of new Trustees and new AFS Officers and Directors, an up-to-date digest of the entire T&RI program as originally approved, and progress to date. It was also the consensus that greater promotional effort is advisable on T&RI accomplishments to date, regardless of the building, especially with reference to T&RI courses and research. The Chairman of the Education Division urged that strong publicity be given the course instructors, both in direct mail literature and in MODERN CASTINGS.

There being no further business to be considered, the meeting was declared adjourned.

Respectfully submitted,
WM. W. MALONEY
*Secretary*

APPROVED:
H. BORNSTEIN, *Chairman*
May 24, 1961

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# Minutes

## AFS Technical Council

## Hotel Sherman, Chicago — June 6, 1961

ROLL CALL: AFS President A. L. Hunt, *presiding.*

*Divisions and Committees:* G. A. Colligan, R. A. Colton, R. P. Dunn, H. G. Haines, C. R. Howle, F. W. Jacobs, W. A. Koppi, J. M. Kreiner, H. E. Henderson, R. B. Fischer, D. L. LaMarche, Jr., J. L. Leach, R. LeMaster, D. Matter, J. G. Mezoff, R. L. Olson, A. H. Rauch, F. L. Riddell, V. Rowell, T. W. Seaton, M. T. Sell, N. Sheptak, J. Thomson, J. Toth, J. Zotos.

*Central Office Staff:* Wm. W. Maloney, S. C. Massari, H. J.

Weber, R. E. Betterley, H. E. Green, J. H. Schaum.
*Absent:* L. H. Durdin, M. C. Flemings, H. W. Gorman, L. R. Jenkins, S. Lipson, W. E. Mason, R. G. Megaw, K. M. Smith.

## Division and Committee Reports

*Sand Division.*

The Core Test Committee is developing reliable standardized techniques for evaluating materials used with the recently developed coremaking techniques. Particularly, they have been working on cores made by the $CO_2$ process, the oxygen-setting oil-type binders, and the non-baking chemically curing binders, primarily of the furan type. Certain effects of the behaviors of these binders necessitates several techniques and several methods. Considerable work has been done as well, on evaluating the synthetic liquid resins used in the hot box process.

Grading, Fineness and Distribution Committee has been investigating new methods of determining clay content of molding sands and studying problems involved in sampling sands, particularly dry unbonded sands, which have always proven difficult to sample reliably.

Green Sand Properties Committee has been studying precision obtainable in green-sand molding as far as casting dimension and weights are concerned. They have been identifying and classifying the factors of sand composition, properties and molding methods which influence casting accuracy. They have, so far, obtained excellent results in gray iron and are having a few problems in aluminum which are currently under study.

Mold Surface Commmittee has been pouring a variety of alloys into test molds with controlled variable mechanical surfaces and have been relating mold surface to casting smoothness as measured by the most modern and advanced methods.

Physical Properties of Iron Foundry Molding Materials at Elevated Temperatures Committee has found that the exact factors influencing veining characteristics are quite complex. Study on this problem continuing.

Physical Properties of Steel Foundry Sands at Elevated Temperatures Committee is engaged in research on casting inclusions particularly those originating from the mold. Slag inclusions are also being checked. Work includes the utilization of radioactive tracers.

Sand Handbook Revision Committee has been engaged in the project of revising the Sand Testing Handbook, now in final stages.

Materials Used in Malleable Foundries Committee has been studying various typical pin-hole defects which occur in malleable castings. The committee has found that no single variable is the cause; rather, a combination of factors have been identified which must be controlled to eliminate this particular defect. Present plans are to continue the attempts to identify possible variables still unknown that may have a bearing on pin holes.

Analysis of Castings Defects Handbook Revision Committee has a nearly complete selection of pictures for the new handbook. The outline, based on the 1947 edition, has been greatly expanded. The revised edition will contain chapters on each cause of foundry loss. Approximately two more years will be required to complete this book and have it ready for publication.

Molding Methods and Materials Committee is in process of editing the chapters completed. It is anticipated that the book will be completed in 1962, and will be a comprehensive manual covering the subject matter with great thoroughness.

Basic Concepts Committee has explored the mechanics and geometry of small particle packing. A study of the description, size and shape of the voids occurring between particles is now in progress.

*Steel Division.*

During the past year the Steel Division has pursued the following objectives: (1) conduct a practical research program, (2) improve the quality of the annual technical conference, (3) proceed with organizing a Steel Castings Design Handbook.

The Research Committee continued to sponsor research at the University of Michigan on nonmetallic inclusions in steel castings, and the mode of formation and identity of macro-inclusions have been studied. Gating systems were also investigated for their effectiveness in preventing macro-inclusions from being carried into the casting by the pouring stream. A resume of this work was presented at the 1961 Congress. The Electric Furnace Steel Committee of the AIME has received permission to have Professors Flinn and Van Vlack present an abbreviated report on this work at the 1961 Electric Furnace Steel Conference.

Four major highlights of the 1961 Steel program at San Francisco are worthy of mention: (1) 24 technical papers, (2) symposium on Vacuum Melting and Casting, (3) technical papers covering the aircraft industry; and (4) an Award certificate for the best technical paper, made by the Sub-committee on Awards.

The Steel Division has set forth the following objectives for the year 1961-62: (a) continue sponsored research work of practical value, (b) improve the quality of the annual technical conference, (c) resolve the present status of the Sub-committee for Handbook on Steel Castings Design, and (d) create new committees closely allied to the Steel Division.

To date, research on non-metallic inclusions in steel castings shows that the major cause of these inclusions is metal-ceramic reactions that occur under certain conditions of temperature, refractory chemistry and concentration of aluminum and manganese in the metal. Actual evaluation of the work, under a variety of plant conditions, has been started and already a variety of variables due to production conditions has emerged.

Membership in the Research Committee will be increased to include representation of the Investment Casting Industry. Should sufficient interest by this industry be shown in AFS activity, a Sub-committee for Investment Castings may be justified.

In view of the passive situation concerning the Handbook on Steel Castings Design, the Technical Council is asked to resolve the question: "Should the handbook be developed for steel castings or should it include all metals of general interest?"

Three new committees are of great interest to the Steel Division, and it is requested that the Technical Council give its immediate approval to the following proposals:

(1) The Titanium and Reactive Metals Committee is presently working under the Light Metals Division. It is submitted that this committee should be assigned to the Steel Division. It is commonly considered that the reactive metals include those metals that have a great affinity for oxygen, nitrogen, hydrogen and many other impurities, and therefore must be processed by special melting and casting techniques.

Some of these metals are titanium, zirconium, hafnium, thorium, vanadium, columbium, tantilum, molybdenum, tungsten, uranium, rhenium. With the exception of titanium, all of these metals have densities greater than iron or steel. According to the widely accepted definition for reactive metals, there is no terminology used which indicates that the reactive metals are related in any way to the Light Metals Division. The melting, casting, and heat treating techniques applied in producing reactive metal castings in many ways parallel the technology required for producing steel castings. The physical chemistry of melting and refining, the effects of temperature and interstitial elements on physcial metallurgy, metal processing and fabrication, and the properties and applications of the reactive metals cover a broad technological scope.

It is of interest to note that those working with reactive metals, such as, Allegheny Ludlum, Universal Cyclops, Mallory Sharron and Temescal, to name a few, are not necessarily concerned with the Light Metals. However, they are readily recognized by their close ties with the Steel Industry. It is requested that the Technical Council authorize transfer of the Titanium and Reactive Metals Committee to the Steel Division as a Sub-committee for Reactive Metals.

The Steel Division proposes to organize a Sub-committee for Stainless Steel Castings. A review of present-day steel foundry interests shows that more and more steel foundries are producing stainless steel. The Alloy Casting Institute has become prominent in this field, but many foundries are not in a position to justify the cost of membership. With the development of low-cost melting furnaces (low frequency) and the increasing number of steel foundries using the basic melting practice, it is the opinion of the Steel Division that organizing this sub-committee is justified.

The Steel Division feels that there is a great need for a Sub-committee for the Heat Treatment of Steel Castings. Where annealing and normalizing heat treatments were once popular, quench and temper heat treat cycles are now being used. The Technical Council is asked to give its immediate approval to the organizing of this sub-committee.

*Die Casting and Permanent Mold Division.*

Two policy meetings were held during the past year with representatives of the American Zinc Institute and the Die Casting Research Foundation of the American Die Casting Institute. The purpose of the meetings was to lay the groundwork for a Joint Research Committee on Metal Flow and Gating. It

was mutually agreed to waive rights to any patents resulting from the gating research; to publish results jointly; to submit periodic progress reports to the three sponsoring groups; and to utilize two technical members and one administrative member from each group for the Joint Research Committee.

To determine whether transparent plastic dies and high speed photography could be used to study the flow of water through various gating systems and thereby serve as a tool for a study of the fundamentals of gating, a preliminary program was initiated. The Aluminum Co. of America and the Dow Metal Products Co. agreed, at their own expense, to study the flow of water in a plastic die using three different gates and two injection speeds, and compare the photographic results with aluminum and zinc castings made under the identical injection conditions.

*Ductile Iron Division.*

After careful consideration the Executive Committee voted to put into effect a plan, similar to that of the Gray Iron Division, known as the Divisional Award for the Best Technical Paper. The Program & Papers Committee was instructed to select the best paper submitted to the Division for presentation at the annual Castings Congress of the AFS – the first award to be given at the 1961 Congress, but after careful consideration the Committee decided not to select any paper this year for special recognition.

Heat Treat Committee. Investigations are now under way on the following projects: quenching and drawing of ductile iron and its relation to the mechanical properties of ductile iron; selective hardening of ductile iron.

Basic Metallurgy Committee. The Basic Metallurgy Committee was consolidated with the Research Committee, since their project interests parallel each other. The merging to take effect with the start of the new fiscal year.

Research Committee. Due to the fact that the Board of Directors have been unable to appropriate funds for Research projects since the inception of the Research Committee in 1959 this committee has been following a "Do it yourself" project. With the completion of the Micro-Test-Lug project in early 1960 the Research Committee started a new project concerning the degree of reproducibility of spectral analysis for residual magnesium in ductile iron. Two different groups of samples were poured and sent to various commercial and foundry laboratories. Data were put in the form of a committee report describing the various problems concerning the magnesium determination in ductile iron.

The Research Committee has again decided to submit the same proposed research project for 1961-1962. This project concerns a study to determine the cause for carbon flotation such as cooling rate, silicon level and time of silicon additions. On the "do it yourself" project phase (no funds required) several projects are being considered: center-line carbide and its cause; Charpy impact properties of ductile iron on both notched and unnotched bars; a study designed to establish the relation of mechanical properties in ductile iron which had been quenched and drawn at various temperatures.

*Industrial Engineering and Cost Committee.*

The main 1960-61 activity of the Committee was the Casting Costing Survey in which various foundries, by metal groups, were asked to anonymously report their cost estimates for certain selected castings. The number of replies was not great but sufficient for a preliminary report. It would appear that cost estimating techniques could be improved by some standardization, based on the fact that a relatively wide divergence of cost per casting resulted.

*Light Metals Division.*

Research activities during the past year have been negligible, since the project to be accomplished was not funded. That project was intended to comprise a thorough review of the last 12 years' research conducted by AFS with the financial assistance of Battelle and the U.S. Government in the gating and feeding of light alloy castings. It is hoped to accomplish this project in the next year, to edit and combine the two existing films and present a coherent picture of this long-range effort for use by AFS membership.

Centrifugal Casting Committee. The Light Metals Division has abandoned its Centrifugal Casting Committee in anticipation of the formation of a general interest committee covering this subject area for all metals, rather than light metals only.

The new book on Recommended Practices for the Sand Casting of Light Metals is largely complete and scheduled for nearby completion.

*Pattern Division.*

Research Committee is developing laboratory techniques for evaluation of presently existing or newly developed materials for pattern and corebox equipment. The preliminary project plan is prepared.

*Malleable Division.*

The Research Committee completed work on the second part of their progress report on Casting and Annealing Heavy White Iron Sections, presented at the 1961 Convention. Work was continued on the effect of base metal composition and quantity of additions required to achieve compact graphite. Present results of this project show that compact graphite (as distinct from spheroidal graphite) can be produced as-cast. Future projects of the committee will include study of tensile properties and fluidity with regard to section size and mass as it will affect micro-structure. Response to heat treatment will also be considered.

The Heat Treatment Committee engaged in applying research data on the heavy-section castings to plant production. This included use of bismuth and tellurium additions to the melt and the study of the annealability of such composition. Tests were conducted in eight member foundries, the results of which are to be reported at the 1962 convention.

The Casting Design Committee has been evaluating studies in other societies and after much deliberation stands ready to form the nucleus of a future General Interest Committee on the writing of a Casting Design Handbook for ferrous metals.

The Finishing & Inspection Committee (6-G) has devoted considerable time to a symposium on standardization of magnetic particle inspection, report to be made at the 1962 Convention.

The Molding Materials & Methods Committee is gathering data preparatory to a study of special molding sand effect on the surface properties of malleable iron.

*Gray Iron Division.*

Research Committee. The outstanding activity of the Division during 1960-1961 was the research program at Case Institute of Technology. An investigation of factors affecting the gating and risering of iron castings was sponsored and supervised, resulting in two excellent papers at the San Francisco meeting.

The Gating and Risering Committee was formed to translate the results of the Case research into actual foundry practice. At a joint meeting of the Research and Gating & Risering committees, it was decided that by using the procedures described in the AFS publication "Gating and Risering Gray Iron Castings" the Committee would: select actual castings that have been troublesome because of gating or risering; compute gating and risering systems according to recommended procedures; obtain before-and-after photographs clearly showing details of gating and risering systems; determine soundness of castings obtained by each system; and include data as to molding and metal which might influence results obtained.

During the past year the Melting Methods Committee began work on the development of a session for the 1962 Congress dealing with melting of cast iron in furnaces other than the foundry cupola.

AWS-AFS Committee on Welding Iron Castings. This joint committee is continuing its work on the development of approved methods for structural welding of iron castings.

*Brass & Bronze Division.*

The research program of the past five years was continued at the University of Michigan, with the cooperation of the Brass and Bronze Ingot Institute and the Crucible Manufacturers Association, both of whom have supplied necessary materials. The project, which was originally set up to establish factors that contribute to the making of pressure tight castings in 85-5-5-5 alloy, has necessarily concerned itself with establishing test pieces and procedures. Only in the past year has it been possible to accumulate data on a representative group of alloys providing cooling curves, temperature gradients, gating and risering distance information. A report of progress was presented at the 1961 Congress.

It has been nine years since there has been a revision of the book on Recommended Practices put out by the Brass & Bronze Division. Accordingly, it is hoped to reactivate a committee to undertake the job of revising the present book, which will take approximately two years.

*SH & AP Control Program.*

Because of new problems arising from water pollution due to industrial wastes, a new committee on "Water Pollution

from Foundry Wastes" has been formed. This committee will investigate the extent of the problem, if any, and the remedies, and results of the investigation will be published.

### Award for Best Scientific Paper

It was the consensus of the Council that the Award should be optional with the divisions, and that AFS headquarters should draft rules for judging the best paper.

### AFS-T&RI Sponsored Research

The Technical Director requested all Divisions and Committees to state their needs for research project funds in 1961-62, for recommendations to the AFS Finance Committee and Board of Directors. All research projects, he pointed out, would be carried on under sponsorship of the Training and Research Institute, with funds contributed by AFS. When all requests were recorded, it was announced that a total of $59,870 would be recommended for the fiscal year ahead.

### 1962 International Congress Papers

It was pointed out that officially sponsored papers for the 1962 Congress must be cleared through the members of the International Committee, and never are reviewed by AFS Program & Papers Committees. It was agreed that the AFS Technical Council will receive copies of the International Exchange Papers Rules in three languages (English, French and German).

There being no further business, the meeting was declared adjourned.

Respectfully submitted,
HERBERT J. WEBER
*Asst. Technical Director*

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# Minutes
# Meeting of AFS Finance Committee
# Union League Club, Chicago — July 18-19, 1960

ROLL CALL: President N. J. Dunbeck, *presiding;* Vice-President A. L. Hunt, Past President C. E. Nelson, Gen. Mgr. Wm. W. Maloney, Treas. E. R. May; Secy. A. B. Sinnett, Tech. Dir. S. C. Massari.

The President announced that the purpose of the meeting was to develop and recommend to the Annual Board Meeting, budgets of Income & Expense for the fiscal year 1960-61, the second year of the 2-year budget established August 1959. He recalled that the 2-year budgets had been strongly influenced by heavy deficits of 1958-59.

The official audit of Society finances as of June 30, 1960, was studied, as well as Staff recommendations for the 1960-61 budgets in comparison with budgeted and actual Income & Expense for the four fiscal years 1956 through 1960.

At the outset, members of the Finance Committee reported that expected improvement of industrial conditions, and the foundry industry in particular, had not materialized and that 1960-61 budgets should again be set conservatively. It was agreed that the Committee would reconvene in January 1961 to consider, if business conditions and budget developments in the first six months warrant, reconsideration of Staff-recommended activities not undertakable at present.

The President also pointed out the continuing necessity to restore AFS finances to a sound, even keel . . . a program undertaken the previous year and admittedly extending at least through the anticipated large International Congress & Exhibit scheduled for 1962 in Detroit. Until this program has been successfully completed, he said, all expense-producing activities should be carefully evaluated before being budgeted.

### Recommended Income and Expense 1960-61

Following discussion of all items of Income & Expense, budgets for 1960-61 were recommended to the Board of Directors as follows:

| 1960-61 | Revised July 1960 | 2-Yr. Rev. Budget 1959-1961 |
|---|---|---|
| Total Income | $843,400. | $1,894,600. |
| Total Expense | 896,400. | 1,796,000. |
| Excess Expense | $ 53,000. | $ 98,600. |

Budgetary recommendations of the Finance Committee for 1960-61 are based on the following specific considerations.

*Member Dues.* Budget $295,000 conservative, based on no membership increase, but stronger solicitation activity.

MODERN CASTINGS. Due to excellent recovery of past year and record Net Income, Expense budget increased 12.9% to enable continued progress. Income budget for Advertising increased 10.3%. Total Income budget increased 9.0%.

*Special Publications.* Income budget $30,000, in belief major new publications not available early in year for distribution. Expense budget $23,000 with understanding that if certain book manuscripts develop earlier than expected the Staff may request additional funds in January.

*Convention & Exhibit Income.* While the Committee expressed uncertainty as to Income results from a first AFS Convention & Exhibit on the West Coast in 1961, it was felt advisable to budget two items: (1) Registration Fees $13,500 based on a Fee of $5.00 for all attending; (2) Space Rentals $160,000 based on 40,000 sq. ft., or approximately 80% of available. At the same time, the Committee agreed that Exhibit space rentals would be a strong factor in any budgetary revisions felt advisable next January.

*Convention & Exhibit Expense.* Staff recommendations totaling $75,000 (Convention $35,000 — Exhibit $40,000) were accepted . . . also subject to reconsideration next January.

BUYERS DIRECTORY. After considerable discussion, a majority of the Committee was disinclined to accept Staff recommendations to publish a 2d edition in Sept. 1961, as originally contemplated. The Committee pointed out that the 1st edition showed a 3-year Net Expense of $15,000, with 3,813 copies in Inventory and sale problematical. The Committee also questioned the actual usefulness of the Directory to foundries seeking supplier sources. The 1961 Buyers Directory therefore was eliminated from the budget, except for payment of such managerial fees as presently committed.

*Income from Investments.* The Committee authorized continued use, as operating revenues, of any Income from Investments available in 1960-61.

*Salaries.* The Committee expressed belief that Salary increases should be held at an absolute minimum until present business conditions improve. Budget $284,400 or 3.3% above 1959-60 Actual $275,300.

*Burden.* Total Burden budgeted $102,500 or 5.7% over 1959-60 Actual, mainly in two essential items . . . General Office Expense, and Board & Executive Committee Travel & Expense.

*Chapter Expense.* Budgeted $5,000 below 1959-60 Actual, due largely to lesser expense of Chapter Officers Conference under present policy of limiting delegates. The Committee recommended that the 1961 Conference be held on the same basis as in 1960 (one delegate af full AFS Expense, 2d delegate at AFS expense for hotel and Conference meals only, 3d delegate at AFS expense for Conference meals only; maximum of 3 delegates), and that major effort be made for full attendance at Regional Administration Meetings. The Committee agreed to consider, in setting budgets for 1961-62, a Staff suggestion that the Conference be held every two years with AFS accepting all expense of 2 delegates per Chapter.

*T&RI Contributions — Research.* Budget $43,200. The Committee expressed full willingness to provide substantially greater funds for Research when Society finances permit, but was unwilling to incur larger deficit spending until 1960-61 Income develops as currently budgeted. The Committee agreed to review the Research contribution budget in January 1961. The Committee also recommended that the entire $47,200 budgeted be contributed to T&RI as soon as possible, with the request that T&RI set up a Reserve for Research on a deferred accounting basis. Any unexpended Research funds at the close of

each fiscal year would then be considered by AFS in budgeting T&RI Research requirements annually.

*T&RI Contributions — Foundry Instructors Seminar.* The Committee declined to budget a 1961 Foundry Instructors Seminar, but agreed to consider the event further when reviewing the budget in January 1961, in terms of better business conditions and AFS Income developments.

*T&RI Contributions — Development & Promotion.* Budget $7,500 including partial payment of $5,000 to tax attorneys handling the T&RI tax exemption case, so as to avoid excessive fee expenditures in any one fiscal year when the case is concluded.

*Retirement Expense.* Budget $14,000, the net of total premiums less employee contributions to be made periodically throughout the fiscal year. The Committee expressed concern that no acceptable revision of the present Staff Retirement Plan, in effect unchanged since 1948, has yet been prepared and submitted. The General Manager was requested to pursue the matter for early submittal of a plan more in keeping with today's practices and requirements.

*Capital Expenditures.* The Committee recommended Capital Expenditures totaling $15,300 as requested by the Staff, to be paid from operating revenue surplus.

*Operating Cash Needs.* The Committee approved Staff requests for Bank Loans totaling $100,000 to provide operating funds during the period July-December 1960 until revenues from Dues, Exhibits and other sources become available. The Treasurer pointed out that because of Excess Expense of $53,000 budgeted for 1960-61 it undoubtedly would be necessary to again request bank loans totaling approximately $100,000 in July 1961, until 1961-62 Exhibit, Dues and other revenues become available.

These Minutes of the Finance Committee meeting were ordered to be made a part of the Minutes of the Annual Board Meeting.

There being no further business to be considered, the Finance Committee meeting was declared adjourned, to reconvene in January for review of budgets presently recommended.

Respectfully submitted
WM. W. MALONEY
*General Manager*

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# Minutes
# Meeting of AFS Finance Committee
# AFS Central Office, Des Plaines, Ill. — Dec. 14, 1960

ROLL CALL: *Presiding,* President N. J. Dunbeck; Vice-President A. L. Hunt, General Manager Wm. W. Maloney, Treasurer E. R. May; Vice-Pres. Nominee J. A. Wagner, Technical Director S. C. Massari, Secretary A. B. Sinnett.

## Purpose of Meeting

The President stated that, prior to the meeting, members of the Finance Committee had received specific Staff recommendations concerning the 1961 Castings Congress & Exposition scheduled to be held May 8-12 in San Francisco. These recommendations, he said, were made on the basis of financial statements as of November 30. Because of the importance of the recommendations to future AFS financial plans, the President stated that he had invited the new Vice-Presidential nominee, J. A. Wagner, to attend the meeting and enter the discussions.

Staff recommendations then were considered in detail and actions taken by the Committee as shown below.

## The 1961 Exposition

The Committee was unanimous in recommending cancellation of the 1961 Exposition, on the basis of Staff estimates of extensive selling efforts and space sales to date. The General Manager reported that selling effort had included 3 separate mailings to all Exhibitor prospects, a special mailing to West Coast sales representatives of Exhibitors, personal contacts with 160 of the 210 Exhibitors at the 1960 Show, and personal contacts with numerous West Coast sales representatives by the Staff and by prospective Exhibitors in determining the interest of their Western representatives.

In addition, West Coast Chapter officials had attempted over several months to develop Exhibit interest, and at least one AFS Director devoted much effort in the same cause. In spite of all efforts, in early December the Show was insufficiently developed, including only 3 West Coast applicants. Numerous Exhibitors stated their representatives would attend, but declined to exhibit. Major reasons for lack of participation were given as:

(1) Existing business conditions and curtailed 1961 budgets; (2) West Coast foundry market insufficient to justify exhibit expense; (3) almost universal apathy of West Coast sales representatives and sales agencies; (4) expense of shipping equipment and display materials to the West Coast, and possibly back home again afterward; (5) failure of some companies to realize that a 1961 Exposition was planned, hence no budget plans; (6) objections to AFS holding three consecutive Exhibits in 1960, 1961 and 1962.

It was felt that to put on a small 1961 Show would violate AFS obligations to all cooperating Exhibitors who had filed space applications, because of possibly smaller attendance. The Committee therefore concurred in 1961 cancellation, and in concentration on the 1962 Exposition in Detroit, commencing 1962 sales promotion months earlier than for any prior AFS Show. The Committee expressed the belief that many firms opposed to or unwilling to participate in the 1961 Exhibit may enlarge their 1962 exhibit plans, particularly when the effect of 1961 cancellation on AFS finances is more generally realized.

## The 1961 Castings Congress

The Committee also seriously considered cancellation of a West Coast Castings Congress in 1961 and moving it to a more central location, or cancellation of any 1961 Castings Congress. For the following valid reasons, the Committee recommended against either action:

(1) Advanced stage of the technical program, with over 80 papers already committed, over 100 now in sight, and many papers committed for specific presentation on the West Coast; (2) belief that AFS sooner or later must hold a technical Congress in the West, in view of 8 Far West Chapters with over 1100 members, a decision deferred since 1947; (3) the only AFS technical Convention ever cancelled in 64 years was in 1945, and then only because of government restrictions on travel; (4) cancellation of any 1961 technical Congress would seriously offend many members interested in AFS primarily as a technical society, and the effect of outright cancellation on AFS divisions and committees would probably be serious.

## The Post-Congress Hawaii Tour

Since the announced Post-Congress Hawaii Tour following the Castings Congress in San Francisco involves no financial obligations on the part of AFS, the Committee recommended that the tour be sponsored as intended.

## Effect of 1961 Actions on AFS Finances

The Committee fully realized that the Society's 1960-61 financial budgets would be seriously affected by the loss of $160,000 income anticipated from the 1961 Exposition. The Committee considered at great length a revised Budget for the balance of the fiscal year, incorporating drastic reductions of Expense to offset loss of Income.

The General Manager recounted steps already taken to reduce expenses of the fiscal year, and described other actions he proposed to take for further reductions. As a result of both, he proposed a decrease of $71,500 in Expense for the 2d half of the year, including a net decrease of $20,000 in Convention & Exhibit expense. As of Nov. 30, he said, committed Expense for the 1961 Congress and Exhibit totaled approximately $22,000.

The Finance Committee expressed the belief that the proposed Staff cuts, while extensive, would not sufficiently offset the much greater loss of Exhibit income. While it was realized that it would be impossible to cut Expenses to the full extent of the Income loss, the Committee felt it necessary to reduce

Expenses by a minimum of $90,000. The General Manager was therefore instructed to take the necessary steps, for Finance Committee approval.

## Final Budget Revisions

As instructed, the General Manager presented further Budget revisions which were approved by the Finance Committee, as follows:

| | Initial Budget | Approved Revision | Incr. or (Decr.) |
|---|---|---|---|
| Total Income | $851,200. | $686,700. | ($164,500.) |
| Total Expense | 928,700. | 835,000. | ( 93,700.) |
| Excess Expense | $ 77,500. | $148,300. | $ 70,800. |

The Finance Committee therefore approved the following *Revised Budget of Expense for 1960-61*, effective January 1, 1961:

| Expense | Initial Budget | Revised Budget | Incr. or (Decr.) |
|---|---|---|---|
| General Administrative Expense | $386,900. | $367,400. | $(19,500.) |
| Publications Expense | 54,000. | 39,300. | (14,700.) |
| MODERN CASTINGS Expense | 262,400. | 258,200. | ( 4,200.) |
| Convention and Exhibit Expense | 75,000. | 55,000. | (20,000.) |
| Chapter Expense | 70,000. | 68,000. | ( 2,000.) |
| Technical Activities Expense | 11,100. | 10,100. | ( 1,000.) |
| T&RI Contributions | 50,700. | 20,200. | (30,500.) |
| Other Expense | 18,600. | 16,800. | ( 1,800.) |
| TOTAL | $928,700. | $835,000. | $(93,700.) |

In approving Revised Budgets of Income & Expense for the fiscal year, some of the reasoning and far-reaching discussions of the Finance Committee, important to Board understanding and acceptance, are shown below.

## Immediate Financing Problems

The Finance Committee called immediate attention to the fact that the 1960-61 Revised Budget contemplated Net Expense of $170,500, even after cutting $93,000 from Expense of the 2d half. As a result, the 2-year budgetary period 1959-1961 could be expected to result in Net Expense of approximately $10,000 instead of Net Income $61,000, as shown below:

| | Initial Budget | Revised Budget |
|---|---|---|
| Excess Income 1959-60 | $138,500. | $138,500. |
| Excess Expense 1960-61, Initial Budget | 77,500. | |
| Excess Expense 1960-61, Revised Budget | | 148,300. |
| Excess Inc. or (Exp.), 2 Years | $ 61,000. | ($ 9,800.) |

Budgets for the 2-year period 1961-1963 will anticipate that Net Income earned in 1961-62 from the International Congress & Show in Detroit would exceed Net Expense of the non-Exhibit year 1962-63 and leave a Net Income balance. This would not be possible unless steps taken to liquidate existing and contemplated deficits: (1) current bank loan of $100,000 and (2) expected Net Expense of $23,400 to June 30, 1961. In addition, it was stated, the Society would require Working Capital of $40,000 to June 30, 1961, making a total of $163,400.

The Finance Committee pointed out that the current AFS financial situation was due largely to deficit financing, in effect since 1955. Prior to 1956 each 2-year budget period showed substantial Net Income, but since 1955 AFS has shown a Net Deficit for each two years, as follows:

| 2-Year Budget Periods | Net Inc. or (Net Exp.)* |
|---|---|
| 1949-1951 | $180,700. |
| 1951-1953 | $126,000. |
| 1953-1955 | $ 21,800. |
| 1955-1957 | ($ 5,500.) |
| 1957-1959 | ($104,000.) |

*Excluding Income from Investments, not available for Cash Operating purposes prior to 1959-60.

The Finance Committee was in unanimous agreement that deficit financing of AFS should end, and that the Society must adopt firm 2-year budgets on a balanced basis, commencing with budgets for the 2-year period 1961-63.

## Recommendations to the Board of Directors

The Finance Committee therefore unanimously recommended, for immediate approval by the Board of Directors, the following:

(a) Authorization to withdraw $153,000 from the AFS Investment Agent Trust Fund by sale of securities held, for the specific purposes of liquidating an existing $100,000 Bank Loan, providing Working Capital for the balance of fiscal year, and concluding fiscal year 1960-61 without outstanding deficits. It was pointed out that the first 4 months of 1961-62 could be expected to produce Net Expense of approximately $75,000, to be financed (if necessary) by a temporary bank loan July 1961, and to be liquidated from 1962 Exhibit revenues, leaving all 1961-62 Net Income to offset 1962-63 Net Expense. Since securities in the AFS Investment Agent Trust Fund (cost $453,043) showed market value of $564,100, it was proposed to reduce the Fund by $153,000 or 27.1%, to approximately $411,000.

(b) To provide immediate Working Capital for AFS operations, the Finance Committee authorized the Treasurer to renew the current $100,000 bank loan to March 31, 1961, and to negotiate a second bank loan of $10,000 (if necessary) at lowest possible rate of interest. The Committee requested approval by the Board of Directors of these interim finance actions.

(c) In view of the Society's experiences with non-Exhibit Castings Congresses in 1955, 1957, 1959 and 1961, the Finance Committee expressed belief that serious consideration must be given the method of selecting locations of future Congresses and Expositions, in order to insure maximum attendance and maximum income to the Society. To this end, the Committee unanimously recommended, for approval by the Board of Directors, the following:

> The location of future AFS Castings Congresses and Expositions should henceforth be decided by the Board of Directors in strict terms of Society financing and estimated attendance. While AFS should at all times welcome invitations from the Chapters and Membership to hold these events in specific localities, all such invitations should be referred to the Finance Committee for study and recommendations prior to action by the Board of Directors.

(d) The Finance Committee called attention to the considerable cost of sponsoring the annual Chapter Officers Conference ($16,300 in 1959 – $10,600 in 1960, 2-year cost $26,900). The Committee recommended that the 1961 Conference be cancelled, and that a 1962 Conference be announced on the same basis as in 1959, namely, two expense-paid delegates per chapter. It was emphasized that decision on the 1961 Conference required immediate action so that the Chapters could be informed well in advance.

(e) The Finance Committee expressed belief that serious consideration should be given the holding of future Castings Congresses with or without Expositions. A non-Exhibit Congress, it was stated, incurs a Net Expense of $30,000-$40,000 from the Convention alone (avg. for last 7 non-Exhibit years, $30,200), and attendance drops 60-80%. The Committee fully realized that AFS has held an annual Castings Congress since 1896, with or without an Exhibit, and that a decision to hold future Congresses only in Exhibit years would no doubt be opposed by many members. The Committee suggested, however, that as long as the current austerity program continues some less costly means of serving the membership might be considered in lieu of a non-Exhibit Congress in 1963, as currently scheduled.

(f) The Finance Committee gave serious consideration to the possibility of a Dues increase, and noted the following record over the past 16 years:

| | AFS Dues Increases | | | |
|---|---|---|---|---|
| Member Class | July 1 1944 | July 1 1945 | July 1 1948 | Dec. 1 1953 |
| Sustaining | $50. | $100. min. | $150. min. | $300. min. |
| Company | 25. | 50. | 65. | 100. |
| Personal | 15. | 15. | 20. | 20. |
| Affiliate | 8. | 8. | 10. | 10. |
| Associate | 8. | 8. | 10. | 10. |
| Junior | * | * | 4. | 4. |

*Student and Apprentice class estab. 1948.

Information provided by the Staff indicated the following changes between the Dues increase of 1953, and June 30, 1960:

| | June 30, 1953 | June 30, 1960 | Increase | %, Incr. |
|---|---|---|---|---|
| Total Members ..... | 11,033 | 12,790 | 1,757 | 15.9 |
| Total Dues Income .. | $241,800. | $293,900. | $52,100. | 21.5 |
| Total Net Expense Items ..... | $290,700. | $383,800. | $93,100. | 32.0 |
| Net Expense per Member ..... | $4.44 | $7.03 | $2.59 | 58.3 |

From the above the Committee noted that Dues Income in 1960-61 was lagging nearly $100,000 per year behind the cost of Net Expense Services to the membership. The Finance Committee therefore instructed the General Manager to prepare an analysis of the total membership and proposals for a Dues increase, for submission by the Committee at the February Board meeting for final approval, to become effective July 1, 1961.

Respectfully submitted,
WM. W. MALONEY
*General Manager*

# Minutes
## Meeting of 1960-61 Nominating Committee
## AFS Central Office, Des Plaines, Ill. — Dec. 12, 1960

ROLL CALL: Past President C. E. Nelson, *presiding;* Past President L. H. Durdin; Region 1 — H. Reitinger, Sr. (Philadelphia Chapter); Region 4 — R. A. Payne (Michiana), J. A. Barrett (Central Indiana); Region 5 — L. H. Brogley (Quad City); Region 6 — J. C. Henderson (Corn Belt), E. W. O'Brien (Tri-State).

### Nomination of President

On motion made, seconded and unanimously carried, incumbent AFS Vice-President ALBERT L. HUNT, Executive Vice-President, Superior Foundry, Inc., Cleveland, was nominated President of the Society for the year 1961-62.

### Nomination of Vice-President

Following submission of names of candidates and on motion made, seconded and unanimously carried, JOHN A. WAGNER, President of Wagner Castings Co., Decatur, Ill., was nominated Vice-President of the Society for the year 1961-62.

Chairman announced that nominations for President and Vice-President assumed continuation of the understanding that these officers would be able to function with minimum travel requirements.

### Nomination of Directors

AFS By-laws were read, relating to organization of the Board of Directors, and as to Nominating Committee procedures. The Chairman pointed out that the committee was expected to name six National Directors for terms of three years each, commencing May 13, 1961, and each from a Chapter Group which otherwise would have no representation on the National Board after next May. He also stated that Directors-at-Large elected by the Board of Directors, are not considered as providing, in themselves, "Chapter Group representation" on the Board.

The six Chapter Groups designated as requiring Nominating Committee consideration were named as follows: Chapter Group B (Region 1), Groups I and J (Region 4), Group M (Region 5), Groups N and P (Region 6).

Candidates submitted by the Chapters, in accordance with the by-laws, then were presented and considered. The Nominating Committee then nominated the following AFS members to serve as National Directors of the Society for the term 1960-1963.

*Chapter Group B (Region 1)* — Daniel E. Best, Superintendent Brass & Iron Foundries, Bethlehem Steel Co., Bethlehem, Pa. (representing Gray Iron, Steel, Philadelphia Chapter).

*Chapter Group I (Region 4)* — C. J. Lonnee, President, Alloyed Grairon Castings Corp., Ravenna, Mich. (representing Steel, Western Michigan Chapter).

*Chapter Group J (Region 4)* — Carl O. Schopp, Asst. General Superintendent, Link Belt Co., Indianapolis (representing Malleable, Central Indiana Chapter).

*Chapter Group M (Region 5)* — Roger J. Hageboeck, Vice-President, Frank Foundries Corp., Moline, Ill. (representing Gray Iron, Quad City Chapter).

*Chapter Group N (Region 6)* — Charles Ossenfort, Foundry Superintendent, Fairbanks Morse Co., Kansas City, Kans. (representing Gray Iron, Mo-Kan Chapter).

*Chapter Group P (Region 6)* — Dale Hall, Works Manager, Oklahoma Steel Castings Co., Inc., Div. American Steel & Pump Corp., Tulsa, Okla. (representing Steel, Tri-State Chapter).

All Officer and Director nominees selected by the Nominating Committee were contacted by telephone and their acceptances personally obtained before adjournment.

### General

The Chairman stated that all names of candidates submitted by the Chapters and not acted upon by the Nominating Committee would be referred to the special Board Nominating Committee, for consideration in electing one Director-at-Large by the Board of Directors, as provided in the by-laws.

It was the consensus that greater effort should be made in the future to impress upon Chapters the importance of naming candidates representing the various technical interests of the Society, as stated in the by-laws. Special reference was made to the fact that few candidates submitted represented Light Metals, Patternmaking, or Die Casting.

### Elections

The Chairman pointed out that, under the by-laws, additional nominations may be made by written petition signed by 200 members in good standing, at any time 45 days prior to the Society's Annual Business Meeting, to be held May 10 in San Francisco. Newly elected Officers and Directors assume office on the day following the close of the Society's Annual Meeting . . . in 1961, on May 13.

There being no further business to be considered, the Nominating Committee was declared adjourned.

Respectfully submitted,
C. E. NELSON, *Chairman*
1960-61 Nominating Committee

# Minutes
## Meeting of Board of Awards 1960-61
## AFS Central Office, Des Plaines, Ill. — Dec. 13, 1960

ROLL CALL: Past President C. L. Carter, *presiding;* Past Presidents F. J. Dost, B. L. Simpson, F. W. Shipley, H. W. Dietert, L. H. Durdin, C. E. Nelson; Wm. W. Maloney, Secretary.

The Chairman thanked all present for 100 per cent attendance, for the third year in a row. Minutes of the 1959 Board of Awards having been previously approved by letter ballot, reading was dispensed with.

The Chairman called attention to an action of the 1959

Board of Awards under which resolution was made that no candidate for an AFS award could receive an award so long as a member of his company was serving on the Board of Awards. The Chairman stated that this resolution had not been presented to the Board of Directors, due to a protest. Instead, the 1959 Board of Awards had decided that each successive Awards group should pass annually on whether such resolution would govern its deliberations. Following discussion, motion was made, seconded and carried, reconfirming the 1959 policy as governing the 1960 Board of Awards, but limiting its application to the award of Gold Medals only.

## Award Procedures

Before considering specific candidates for awards, the following procedures were approved:

(a) Following study of Award Fund accumulations and past frequency of medal awards, agreed that the Penton and Simpson medals would be considered for 1961 awards.

(b) Agreed that all medals would be considered interchangeable for purposes of 1961 awards.

(c) Agreed that any recipient of an Award of Scientific Merit or Service Citation would be eligible for future consideration for other awards.

(d) Agreed that all persons submitting nominations for award consideration, not acted upon by the Board of Awards, be thanked for interest and cooperation over the signature of the Chairman, and urged to continue such interest.

(e) Agreed that the Secretary would continue to prepare and maintain a single, cumulative and permanent book of all nominees submitted in prior years, kept up to date in terms of award decisions; and that the Board of Awards each year would receive full data on all new and continuing candidates in advance of each annual meeting.

## Gold Medal Awards

Following consideration of all candidates submitted, the Board of Awards recommended, on motions made, seconded and carried, that the following Gold Medals be awarded in 1961 for the reasons stated, subject to approval by the Board of Directors:

To MERTON C. FLEMINGS, JR., Asst. Professor Metallurgy, Massachusetts Institute of Technology, Cambridge, Mass. – the PETER L. SIMPSON GOLD MEDAL "for outstanding contributions to the Castings Industry in the practical application of basic research to production of quality castings, particularly in the Light Metals field."

To WILLIAM S. PELLINI, Head, Metallurgy Dept., Naval Research Laboratories, Washington, D.C. – the JOHN A. PENTON GOLD MEDAL "for exceptional contributions to the science of Metal Casting through leadership in fundamental foundry research at the U.S. Naval Research Laboratories, particularly in the field of heat transfer and flow of metals."

## Awards of Scientific Merit

After consideration of candidates, the following Awards of Scientific Merit for 1961 were recommended for approval by the Board of Directors, on motions made, seconded and carried:

To HARVEY E. HENDERSON, Technical Director, Research Dept., Lynchburg Foundry Co., Lynchburg, Va. – the AWARD OF SCIENTIFIC MERIT "for valuable contributions to Metallurgical Research in the casting of Gray Iron and Ductile Iron."

To THEODORE R. SCHROEDER, Foundry Superintendent, Pontiac Motor Div., General Motors Corp., Pontiac, Mich. – the AWARD OF SCIENTIFIC MERIT "for notable contributions to AFS and the Foundry Industry in the development of a revolutionary process for the production of cores, and new concepts of automated molding and aluminum processing."

To HERBERT J. WEBER, Director of SH&AP Control Program, American Foundrymen's Society, Des Plaines, Ill. – the AWARD OF SCIENTIFIC MERIT "for exceptional services to the Society and the Foundry industry in directing the Safety, Hygiene and Air Pollution Control Program for the betterment of the industrial community."

## AFS Service Citations

After consideration of candidates, the following Service Citations for 1961 were recommended for Board of Directors approval, on motions made, seconded and carried:

To ALEXANDER D. BARCZAK, Operations Vice-President, Superior Foundry, Inc., Cleveland, Ohio – the AFS SERVICE CITATION "for outstanding and inspirational services to AFS, its Chapters and the Ferrous Foundry Industry, for inspirational encouragement of young men to foundry careers, and for constant willingness to help the other fellow."

To KENNETH M. SMITH, Staff Engineer, Caterpillar Tractor Co., Peoria, Ill. – the AFS SERVICE CITATION "for noteworthy contributions to the Society and Metal Castings industry in the development of engineering principles to make the foundry 'a better place to work.'"

To JESS TOTH, Vice-President of Sales, Harry W. Dietert Co., Detroit, Mich. – the AFS SERVICE CITATION "for his dedication to the principles of AFS and long, continuous service to the Society and the Castings Industry, particularly in the field of Education."

## Honorary Life Membership

On motion made, seconded and carried, Honorary Life Membership was recommended for presentation to Norman J. Dunbeck as the outgoing President of the Society.

## C. E. Hoyt Memorial Lecturer 1961

It was pointed out that the 1961 Hoyt Lecturer, John B. Caine, had received Honorary Life Membership in 1955, as the recipient that year of the Society's Whiting Gold Medal. The Secretary therefore was instructed to arrange a suitable gift or emolument for presentation ot MR. Caine in appreciation of his lecture.

## Acceptances

It was agreed that all 1961 award recipients, following approval by the Board of Directors, would be notified of their selection by means of letter signed by Chairman Carter, and that first publicity on selectees would appear in MODERN CASTINGS in the earliest possible issue.

## Presentations 1961

The Board of Awards agreed that all 1961 awards would be presented as follows: Awards of Scientific Merit and Service Citations at the Annual Business Meeting; Gold Medal Awards at the Annual Banquet; all awards presented by the Chairman; all award recipients to receive leather-bound copy of programs for their respective meetings.

## General

(a) It was agreed that citations for all 1961 awards would be drafted by the Secretary for approval by Chairman Carter.

(b) It was agreed that the S. Obermayer Fund would henceforth be discontinued as a separate fund and would be consolidated with other Award funds. The Secretary informed the Board of Awards that officials of the S. Obermayer Co., when asked for approval of this procedure, had expressed willingness for AFS to make such use of the fund as the Board of Awards might determine.

(c) The Secretary stated that he was authorized to invest any accumulated cash in excess of $2,000, and that his only authorization for such investment was in U.S. Savings Bonds Treasury Notes. He pointed out that a number of bonds would be maturing in the next several years and that all current bond investments would mature by 1967. It was the consensus that the Secretary should discuss the investment of Award Funds with the Society's financial advisor at the Harris Trust & Savings Bank, Chicago, to determine the best types of investment for such funds.

(d) Attention was called to action of the 1959 Board of Awards, requesting development of a distinctive lapel pin designating Award status, for all past, present and future Award recipients. The Secretary stated that no design had yet been developed. It was the consensus that this should be done, all expense on pins for Award recipients to be a Board of Awards expense. The secretary stated that the Board of Directors had declined to develop similar pins. It was agreed that an Awards pin should be of the screw type, the same size as the current AFS member button, probably illustrated with the obverse side of the Gold Medals rather than with the AFS insignia, and of a different color and lettering than the AFS member pin.

### Resolution re Past President L. N. Shannon

The Secretary called attention to the tragic death of Past President and Mrs. L. N. Shannon in an automobile accident in October and suggested the desirability of a Board of Awards resolution for the family. By unanimous agreement, the Secretary was instructed to prepare an appropriate resolution for approval by the Chairman, such resolution to be spread on the minutes of the Board of Awards meeting and copies prepared to be signed by all members of the 1960 Board of Awards, for surviving members of the Shannon family.

There being no further business to be considered, the meeting was declared adjourned.

Respectfully submitted,
WM. W. MALONEY
*Secretary*, Board of Awards

APPROVED:
C. L. CARTER, *Chairman*
January 4, 1961.

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# 18th Annual Chapter Officers Conference

# AFS Headquarters and LaSalle Hotel, Chicago — June 15-16, 1961

## PROGRAM

*Thursday, June 15*

Group Buffet Breakfast — (LaSalle Hotel)
Chapter Workshops (AFS Headquarters) — 1st Series
Group 1 — Progressing AFS Technical Activities .....Massari
Group 2 — Behind the MODERN CASTINGS Scene .Green-Schaum
Group 3 — Education and T & RI .....Betterley
Group 4 — Organizing for 1961-62 .....Sinnett
Chapter Workshops — 2nd Series
Buffet Luncheon
Chapter Workshops — 3rd Series
Chapter Workshops — Final Series
Annual Conference Dinner — (LaSalle Hotel)
Program .....“The Mid-States Four”
Special Delegate Meetings

*Friday, June 16*

Group Buffet Breakfast — (LaSalle Hotel)
Regional Administration Meetings .....Chairman
Planning the Regional Conference .....Maloney
More and Better Chapter Programs .....*Panel*
KARL KOSTENBADER .....Philadelphia
JOHN LOWE .....Saginaw Valley
LARRY KRUEGER .....Wisconsin
BRUCE FARROW .....Southern California
Financing AFS Activities .....Maloney
Membership Activity and AFS Dues .....Sinnett
Luncheon
National Nominations and Elections .....Maloney
The 1962 International Foundry Congress .....Hewitt
What SH & AP Means to Foundries .....Weber
The Shakeout .....Chairman

## ATTENDANCE

*AFS President & Conference Chairman* — ALBERT L. HUNT.
*AFS Vice-President* — JOHN A. WAGNER.

### National Directors

Daniel E. Best, Bethlehem Steel Co., Bethlehem, Pa.
Roger J. Hageboeck, Frank Foundries Corp., Moline, Ill.
Dale Hall, Works Mgr., Oklahoma Steel Castings Co., Tulsa, Okla.
C. J. Lonnee, Alloyed Grairon Castings Corp., Ravenna, Mich.
Carl O. Schopp, Link Belt Co., Indianapolis, Ind.
*AFS General Manager* — WM. W. MALONEY.
*Guest* — ATHEL DENHAM, Denham & Co., Detroit, Mich.

### Chapter Officers

**Birmingham District**
Vice-Chairman & Program Chairman Wesley J. Estes, Prodn. Supt., U. S. Pipe & Foundry Co., Birmingham, Ala.
Secretary Ralph A. Petersen, Dist. Mgr., Pangborn Corp., Birmingham, Ala.

**British Columbia**
Chairman John F. Prior, Plant Mgr., Maple Leaf Pattern Works, Ltd., Vancouver, B. C.
Vice-Chairman & Program Chairman Arnold W. Greenius, Assoc. Rsch. Met., B. C. Research Council, Vancouver, B. C.

**Canton District**
Chairman Frank P. Tobakos, Dist. Sales Mgr., Eastern Clay Products Co., Wadsworth, Ohio.
Vice-Chairman Gerald N. Moorhead, Fdy. Met., United Engineering & Foundry Co., Canton, Ohio.

**Central Illinois**
Vice-Chairman & Program Chairman Lawrence Winings, Fdy. Supt., Wagner Casting Co., Decatur, Ill.

**Central Indiana**
Vice-Chairman George W. Dehn, Sales Engr., Wheelabrator Corp., Mishawaka, Ind.
Secretary Anthony B. Downey, Asst. Gen. Supt. Fdy., International Harvester Co., Indianapolis, Ind.
Treasurer Frank J. Sabla, Plant Mgr., Terre Haute Malleable & Mfg. Corp., Terre Haute, Ind.

**Central Michigan**
Vice-Chairman & Program Chairman Kenneth W. Rhoads, Secy.-Treas., Engineering Castings, Inc., Marshall, Mich.

**Central New York**
Chairman Donald C. Brainard, Sales & Serv. Repr., The Dayton Oil Co., Syracuse, N. Y.
Vice-Chairman & Program Chairman Robert F. Shea, Fdy. Engr., Crouse-Hinds Co., Syracuse, N. Y.

**Southern Tier Section**
Chairman William G. Parker, Fdy. Proc. Specialist, General Electric Co., Elmira, N. Y.

**Central Ohio**
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# ALUMINUM-URANIUM ALLOY CENTRIFUGAL CASTING

by N. E. Daniel, E. L. Foster and R. F. Dickerson


## ABSTRACT

**Centrifugal casting techniques were investigated as a method of producing hollow cylindrical extrusion billets of aluminum-35 w/o uranium. Among the variables evaluated were melt temperature, mold and pouring spout configuration, mold speed and method of pouring. With the equipment employed, it was found that the best castings were produced using a pouring temperature of 2400 F, a heavy-walled steel cylinder rotating between 700 and 900 rpm for the mold and a bottom pouring technique employing a retractable pouring spout.**

**Sound, nonporous billets 26 in. long and 5 in. in diameter were produced with a yield after machining of over 75 per cent of the original charge. The major losses occurred in the pouring spout and cup assembly. This loss is relatively unaffected by the casting length, and, therefore, castings of greater length than 26 in. should result in even greater recoveries.**


## INTRODUCTION

The most commonly used alloy fuel for low temperature water cooled water moderated reactors are those of the aluminum-uranium alloy system. One method of fuel element fabrication currently employed involves coextrusion of aluminum-uranium alloys and pure aluminum to form tubular elements clad inside and out with aluminum.[1,2] For this process a tubular casting is desirable. However, when alloys containing more than 20 w/o uranium are lip poured in static molds, porosity and segregation are encountered.

The porosity is attributed to shrinkage and to gas evolved during freezing. The segregation is due primarily to differences in the densities of the various phases present, and to the widely separated liquidus-solidus temperatures.[1,3-5] Figure 1 shows the phase diagram of the aluminum-uranium system with a table of density of the various phases.

Since horizontal centrifugal casting has been established as a method of obtaining dense castings, and has found acceptance in such diverse casting applications as the production of cast iron pipe and high integrity steel gun barrels,[6] and since the castings desired were essentially pipes of unusually high integrity, this technique appeared particularly attractive as a method of casting aluminum-uranium extrusion billets. In addition to improving the ingot soundness, horizontal centrifugal casting should eliminate end-to-end gravity segregation.

Permanent molds could be used, and the casting surfaces should require a minimum of machining prior to extrusion. No gating or riser scrap should be generated. Of equal importance to the above advantage is the fact that the process is adaptable to the production of long tubes. Thus, it was considered possible that the process would be amenable to the pro-

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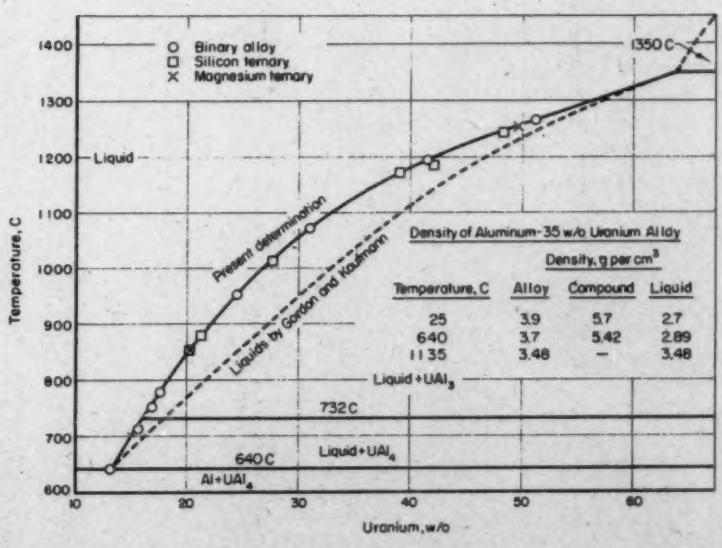



Fig. 1 — Partial aluminum-uranium constitutional diagram.




**Fig. 2 — Machined interior surface of an aluminum-35 w/o uranium 10 in. long extrusion billet exhibiting center shrinkage band.**

duction of multiple-length extrusion billets. The production of multiple-length billets and the reductions in scrap generation and in fabrication cost could offer a distinct economic advantage.

The ultimate objective of the study was to determine the feasibility of producing multiple-length extrusion billets. However, because of the numerous variables encountered in centrifugal casting it was desirable to conduct the investigation in two phases. The first phase was concerned with the application of the technique to the production of single-length billets, and the second phase was concerned with the production of multiple-length billets.

## PREPARATION OF SINGLE-LENGTH CASTINGS

### *Experimental Procedure and Equipment*

During the initial investigations, which were designed to obtain a more complete understanding of the problems that would be encountered in casting these particular alloys, single-length extrusion billets of the aluminum-25 w/o uranium alloy were prepared. The alloys were melted in either graphite or zirconia crucibles with the molten alloy being degassed by bubbling helium through it. The mold was made of tubular steel approximately 5 in. in I.D. with a wall thickness of ¼-in.

The back or closed end of the mold was sealed with a ½-in. steel plate attached to a turntable which was rotated at 550 rpm. The forward or charging end of the mold was closed with a one in. thick graphite plate having a 2 in. diameter hole for teeming. Through this hole a graphite pouring spout was inserted into the mold. The position of the pouring spout in relation to the forward end of the mold was changed with each succeeding melt.

Visual examination of the billets after sectioning revealed that the castings were essentially sound and had good exterior surfaces. The only evidence of porosity was noted in a band approximately one in. wide located about midway from the ends of the castings. The appearance of this band and its location, as shown in Fig. 2, indicated that the porosity was caused primarily by shrinkage.

In the next experiments, melts of both 25 and 35 w/o uranium were poured into a steel mold with a tapered wall. This mold, 10 in. in length with an I.D. of approximately 5 in., was mounted so that the rear of the mold, consisting of a ½-in. thick section, was in contact with the turntable. The front or charge end of the mold had a wall thickness of ⅛-in., and extended without support from the turntable. It was thought that such a design would impart directional cooling and displace the porous zone to the extreme front end of the casting. There it would be removed by cropping.

Since the effect of mold speed had not been established, castings were made using speeds up to 1200 rpm. The pouring temperatures investigated were 2100, 2200 and 2300 F for the 25 w/o uranium alloy, and 2300, 2400 and 2500 F for the 35 w/o uranium alloy. As expected, the increased cooling rate obtained at the rear of the castings displaced the porous shrinkage area toward the front of the mold;

however, the extent of this displacement was not as great as expected. Visual examination of these castings revealed that mold speeds of 700 to 1200 rpm yields the best interior surfaces.

The casting poured into the mold rotating at 1200 rpm exhibited exterior surface imperfections which appeared to be hot tears that had subsequently refilled with molten alloy. In castings poured at 2300 and 2400 F into the mold rotating at 700 or 900 rpm and heated to 400 or 500 F at the front end, it was possible to obtain ingots containing sound material for 75 per cent of their length.

In an effort to improve the quality of the castings in both casting integrity and homogeneity, various types of pouring spouts were investigated. These investigations showed that the best quality castings were obtained when a spout having only one large hole, which introduced the metal normal to the horizontal axis of the mold, was employed. This permitted the introduction of the metal at the exact location at which it was desired.

### *Evaluation of Melts and Techniques*

In most cases the as-cast shapes were evaluated by radiography and by macro- and micro-examination. In certain cases these tests were supplemented by chemical analyses. The aluminum-35 w/o uranium castings poured into steel molds were evaluated in detail. Examination of the surfaces revealed that acceptable inside surfaces were obtained at mold speeds of 700 and 900 rpm. All of the 35 w/o uranium castings poured at 2400 F, and with mold speeds of 550 to 900 rpm, exhibited good exterior surfaces that required little or no machining. Figure 3 shows an ingot cast into the tapered steel mold. It can be seen that the inside surface of the ingot is good.

Approximately three-fifths the distance from this end (the back end of the mold) there is a beginning of the roughness and porosity noted in these ingots. The band of extreme porosity noted in the earlier ingots is located about three-fourths of the distance from the back end of the castings. Radiographic examination of sections cut from the ingots revealed areas of high uranium concentration present in those sections of the ingots where the molten metal was introduced. The variation in uranium content in a number of castings was ±5 to 8 w/o from the nominal 35 w/o, with the lowest uranium concentrations occurring in the areas of the castings furthest removed from the pouring point.

### *Conclusions from the First Phase*

From the experiments it was observed that the following conditions were necessary for the production of satisfactory castings:

1. A mold speed at or near 700 rpm on a mold 5 in. in diameter. Since the effects of increasing the rotational speed of the mold is to increase the centrifugal force exerted in the molten metal, and thus increase the tendency toward radial segregation, it is desirable to use the slowest speed compatible with good interior surfaces.
2. A melt temperature of approximately 2400 F is optimum for the 35 w/o uranium alloys when centrifugal casting. This relatively high temperature is necessary to insure that the melt remain at least partly fluid until the entire mold cavity is filled.
3. A method of depositing the metal uniformly along the length of the mold is required. This elimi-

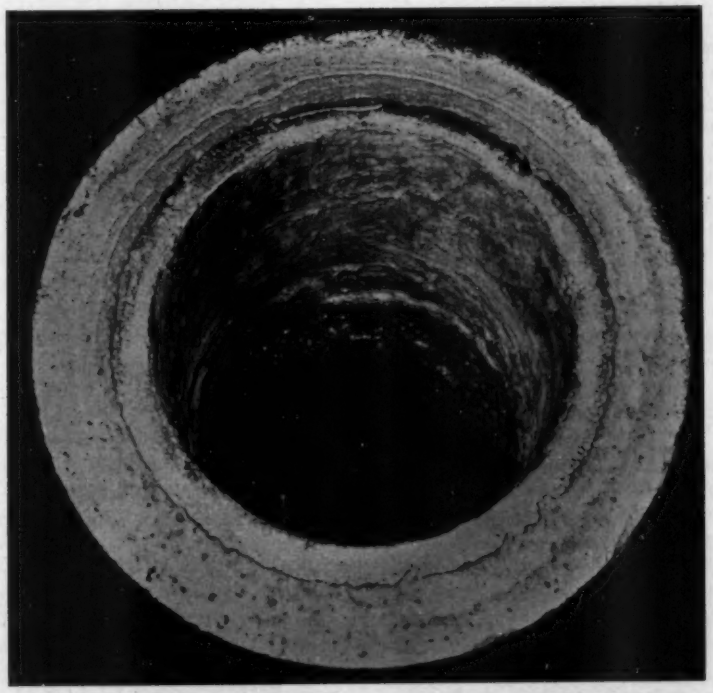


**Fig. 3 — 10 in. long aluminum-35 w/o uranium ingot cast into tapered steel mold. Note surface imperfections and porosity in far end of casting. 1 ×.**

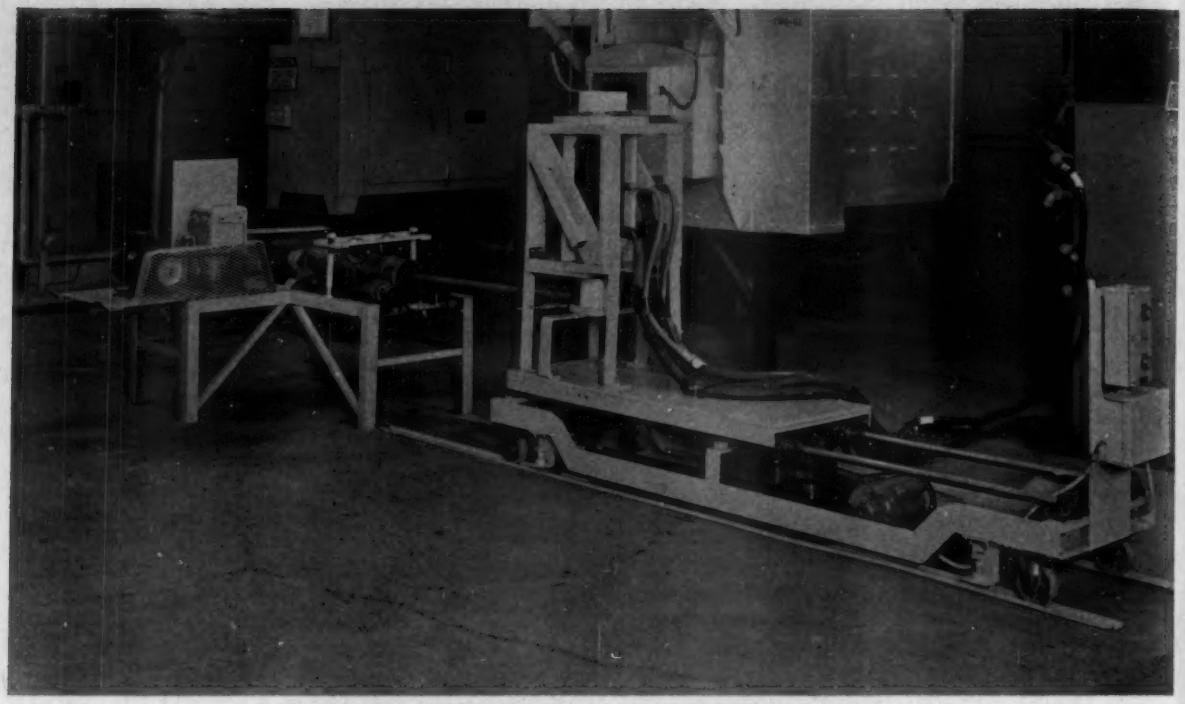


Fig. 4 — Centrifugal casting apparatus and retractable pouring spout assembly. The coil and crucible are in place for bottom pouring.

nates or appreciably lessens the end-to-end segregation. Areas of high uranium concentration are produced where the molten metal impinges upon the mold or upon previously solidified metal.

## PREPARATION OF MULTIPLE-LENGTH CASTINGS

### *Experimental Procedures and Equipment*

Upon completion of the first phase of the program in which single-length extrusion billets were prepared, emphasis was placed upon the production of aluminum-35 w/o uranium castings of sufficient length to yield two or more extrusion blanks up to 8.6 in. in length with an O.D. of 4.8 in. Mold speeds of 700 to 800 rpm and pouring temperatures of 2300 and 2400 F were used as the basis for the production of multiple-length castings. Further studies of spout configurations were required since the spout was to be moved along the axis of the mold as the metal was poured, whereas previously it had remained stationary.

The mechanism designed to produce movement of the spout is shown in Fig. 4. As shown in Fig. 4, the crucible and induction heating assembly was mounted on a movable platform which rested upon two one in. diameter steel rods which served as ways. The pouring cup and spout were also mounted on the movable platform. In Fig. 4, the cup and spout are in position for bottom pouring, a technique selected for use on the final sizes of the program.

Beneath the platform there was a reversible electric motor connected by a sprocket and chain to a speed reducer, which in turn was connected by a chain and sprocket assembly to the sprocket cluster shown directly beneath the right end of the movable platform. A continuous chain drive connected the sprocket cluster to a single sprocket mounted at the left or forward end of the platform. Contact with the movable platform was obtained by an extended link which engaged an extension welded onto the bottom of the platform. The chain drive permitted essentially slip-free operation, and by changing a sprocket in the sprocket cluster many spout withdrawal speeds were obtainable.

The electrical controls consisted of a three-station push-button control box and two limit switches. These permitted either automatic or manual control when it was desired to change the direction of travel of the pouring platform. The various speeds available, the ability to start, stop and to change the direction of travel when and where desired, made this a sufficiently versatile piece of equipment for the investigation of pouring techniques for centrifugal casting.

### *Preparation of Castings*

Since the centrifugal casting apparatus used to produce the single-length extrusion billets introduced eccentricities in the ingots when molds of 18 and 26 in. in length were used, it was necessary to design an apparatus in which the mold would be supported upon a roller bed. This apparatus is shown in Fig. 4. It consisted of a steel mold supported by 2 in. wide by 4 in. in diameter wheels mounted on an angle iron frame. The mold was driven by an electric motor through a speed selector connected by a belt drive to a shaft mounted in the back end of the mold.

During the casting of the initial 26 in. long billets

using this device the melts were lip poured. An evaluation of the ingots so produced revealed extensive porosity beneath the interior surfaces of the casting extending to a depth of approximately 50 per cent of the wall thickness. This porosity was attributed to the uneven feeding of the mold as a result of the variable pouring rate. Shrinkage was also noted, and manifested itself as linear depressions in the interior surface of the cast shapes.

These results indicated the need for a pouring rate that would deposit the proper amount of molten metal at or near the position in the mold where it was to solidify. It was thought that this could most easily be obtained by pouring through an orifice. It was also thought that less gas pickup would be encountered by bottom pouring.

A bottom pouring induction coil and melting crucible assembly was mounted in the movable platform. The use of bottom pouring permitted the operator to obtain the desired constant metal flow into the pouring cup by using a break-off plug with the required orifice. It also decreased the exposure of the molten metal to the atmosphere. Figure 4 shows the bottom pour coil-crucible assembly and the mold assembly used in these studies. The melting and pouring techniques finally developed for 26 in. long billets were:

1. The aluminum was melted to 1800 F.
2. Uranium was added intermittently over a period of 10 to 20 min.
3. The melt was stirred while adding uranium, and stirring was continued until the uranium was in solution.
4. Melt temperature was increased to 2400 F.
5. The graphite pouring cup and spout assembly preheated in an electric furnace to 1400 F was placed in position beneath the coil.
6. The coil and pouring mechanism was rolled forward to pouring position.
7. The melt was stirred vigorously and bottom poured into the mold rotating at 850 rpm.
8. At approximately the instant that the metal began filling the mold, the retraction mechanism of the pouring assembly was started.

It should be noted that this melting cycle differs from that often used in casting aluminum-uranium alloys in that no outgassing techniques were employed, and the melt was not skimmed prior to pouring. The improved protection from atmospheric contamination afforded the metal during pouring, and the easy path for gas removal through still molten metal, are responsible for eliminating the need for outgassing.

A charge of 15 kg of metal was employed in casting the 26 in. long billets. It has been determined that approximately 19 sec is required to pour a melt; therefore, the rate of retraction of the pouring spout was 26 in. in 19 sec. From surface temperature measurements it was determined that the pouring cup and spout temperatures were approximately 900 F at the time of pour.

### *Evaluation of Multiple-Length Castings*

Since one of the primary reasons for investigating centrifugal casting as a technique for the production of cylindrical extrusion billets was to determine if substantial savings could be affected in the amount of material that must be recycled or otherwise reclaimed, a material balance was obtained on all of the ingots produced by this technique. This material balance included the weights of all materials used in the charge and all losses during melting and casting. Also included in the original material balances were the weights of all materials removed from the ingots during machining and for chemical and radiographic analyses. For brevity, and to make the material balance more meaningful, the average material balance for 5 in. O.D. by 26 in. long extrusion billets is given in Table 1.

**TABLE 1 — MATERIAL BALANCE IN KILOGRAMS**

| | |
|---|---|
| Charge wt. | 15.00 |
| Material left in crucible and pouring cup and spout | 1.60 |
| Casting wt. | 13.40 |
| Wt. of material removed from ends and surface | 1.20 |
| Total wt. of machined castings | 12.20 |

The 26 in. long ingots were evaluated by various techniques—principally visual inspection, radiographic examination and chemical analyses. The visual inspection revealed extremely good exterior surfaces on all melts, including one 25 w/o uranium ingot. This ingot is shown in Fig. 5. In Fig. 6 are two aluminum-35 w/o uranium castings exhibiting the same fine surface. The interior surface of an aluminum-35 w/o uranium casting is shown in Fig. 7. Castings of this type required the removal of only approximately $\frac{1}{16}$-in. of material from the radius of the inside surface to obtain completely sound material. Radiographic examination of sections cut from the ingots revealed areas of low uranium concentration around the interior surface of all castings. Figure 8 shows, typical radiographs of ring sections. Although the areas depleted in uranium vary in thickness from end to end in the castings, chemical analyses of the heat castings made varied only from 32.1 w/o uranium to 35.3 w/o uranium. The $\frac{1}{4}$-in. wide bands which transverse the radiographed sections represent gaps where material was removed from chemical analyses. Some of the chemical analyses obtained are shown in Table 2, and indicate the range of uranium concentrations that were detected. These analyses show that the material removed from the inside surface during machining to produce a sound and uniform surface was extremely low in uranium containing 14.5 to 16.8 w/o.

An examination of the material balance shows that the greatest loss was due to material left in the pouring train and crucible. The material left in the crucible was almost negligible, and certainly no more than that which would have been removed from the melt if it had been skimmed. The major portion of this 1.60 kg was left in the pouring cup and spout. This material was massive and suitable for recharging.

It is believed that the results obtained during this investigation show that centrifugal casting techniques

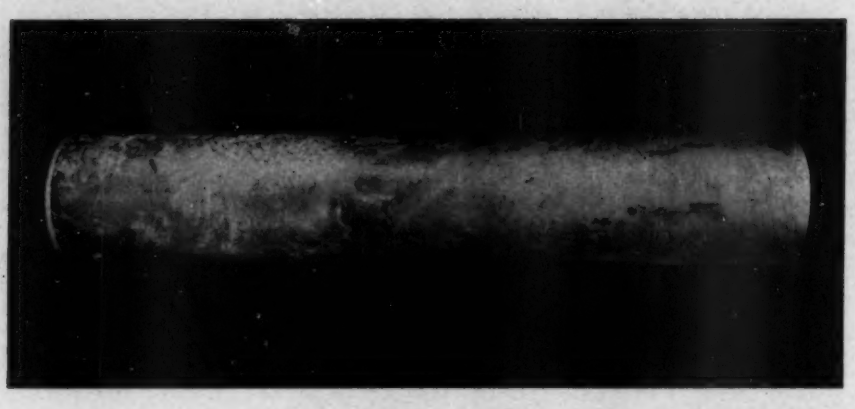


Fig. 5 — Exterior surface on a 26 in. long aluminum-25 w/o uranium extrusion billet. The surface finish is superior to that generally obtained by static casting techniques.

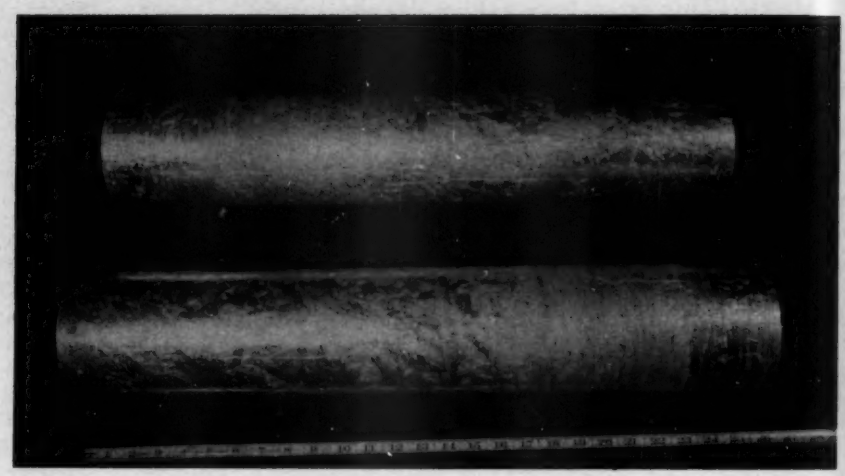


Fig. 6 — Exterior surfaces of two 26 in. long aluminum-35 w/o uranium extrusion billets. The fine surface finish was typical of that obtained on multiple-length 35 w/o uranium castings toward the end of the program.



Fig. 7 — Interior surface of 26 in. long aluminum-35 w/o uranium extrusion billet. Only minimum machining was necessary to produce a satisfactory interior surface.

TABLE 2 — CHEMICAL ANALYSES OF 26 IN. LONG ALUMINUM-35 w/o URANIUM BILLETS

| Ingot | Sample Location | Analyses Composition, w/o |
|---|---|---|
| 9 | Front | 34.7 |
| | Middle | 33.8 |
| | Inside Surface | 16.8 |
| 12 | Front | 32.1 |
| | Middle | 35.3 |
| | Rear | 32.6 |
| | Inside Surface | 15.5 |
| 13 | Front | 32.0 |
| | Middle | 33.0 |
| | Rear | 37.3 |
| | Inside Surface | 14.5 |

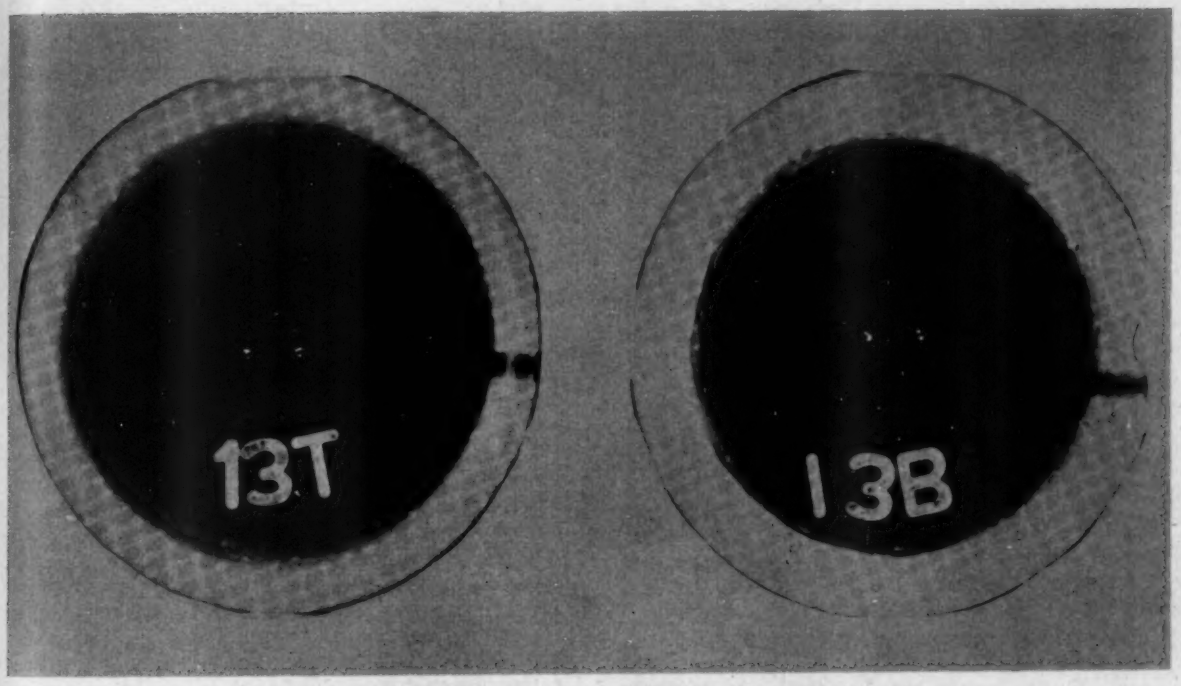



Fig. 8 — X-ray of front and rear sections of aluminum-35 w/o uranium extrusion billet. Some migration of the uranium toward the major diameter is apparent. The gaps in the edges were produced by sampling of chemical analysis.

are adaptable to the production of acceptable aluminum-uranium extrusion billets. Furthermore, the inherent economics of multiple-length castings, permanent molds and high yields that are associated with centrifugal casting are realized without sacrificing casting quality. Refinements in the pouring techniques and equipment might result in even better recoveries than those obtained during this study.

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# GRANULAR MOVEMENT DURING SQUEEZING

by D. C. Williams


## ABSTRACT

**Density of the sand adjacent to a squeeze board is greater than that found at the parting line of a flask. The author attempts to explain this using material from the discipline of soil mechanics. Granular movement during squeeze molding operations is divided between a movement toward a horizontal direction and a movement toward a vertical direction. The author suggests that better squeeze compaction would be accomplished if squeeze boards were decreased in size to take advantage of greater sand movement.**


## INTRODUCTION

The compaction of sand mixtures by the squeeze method has always been marked by the relatively short distance the rigid squeeze board would descend inside the perimeter of the flask. It is also well known that the density of the sand immediately adjacent to the squeeze board is considerably greater than that found at the parting line. This has usually been considered as the "nature" of squeezing, and so it is. This paper will make an attempt to explain this nature using material from the discipline of soil mechanics as provided by Terzaghi.[1]

## SQUEEZE BOARD DATA

Figure 1 is a reproduction of Terzaghi's[1] Fig. 15a. Let the superimposed load $q'c$ be equivalent to the weight of the squeeze board, and the load $q''c$ be equivalent to the load applied to the squeeze board. Point $a$ will indicate the side of a square or round squeeze board or the long side of a rectangular squeeze board.

Through point $a$ there are two lines, $aD_A$ and $aD_p$, which divide the mass of sand mixture into three zones, namely:

1) The active Rankine state wherein the sand mass is expanding in a horizontal direction.
2) Zone of radial shear.
3) The passive Rankine state wherein the sand mass is contracting in a horizontal direction.

It is important to note that above the line $aD_A$, and to the right under the squeeze board, the movement of the sand mass is along surfaces of sliding which descend from the surface at an angle $(45° + \phi/2)$. Also above the line $aD_p$ and to the left above which there is no supercharge load, the sand mass will move along surfaces of sliding at an angle of $(45° - \phi/2)$. Note also that in the zone of radial shear the surfaces of sliding are logarithmic in shape, and connect the angles of the surfaces of sliding in the active and passive Rankine zones.

In general squeeze molding practice the squeeze board is so designed that it fits just inside the perimeter of the flask (straight or tapered sides). Under such conditions one can consider that the inside surface of the flask can be thought of as a vertical surface through point $a$. Therefore, during squeezing sand grain movement along the surfaces of sliding will involve only the active Rankine state and a portion of the zone of radial shear.

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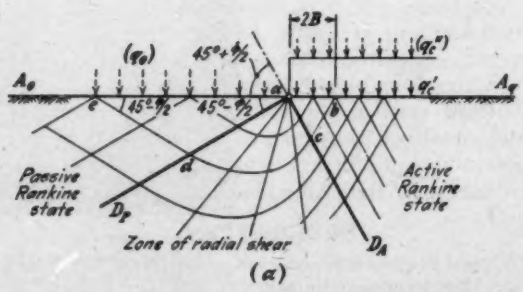



**Fig. 1 — Movement of sand grains under one side of a square or long side of a rectangular squeeze board (Terzaghi[1]).**

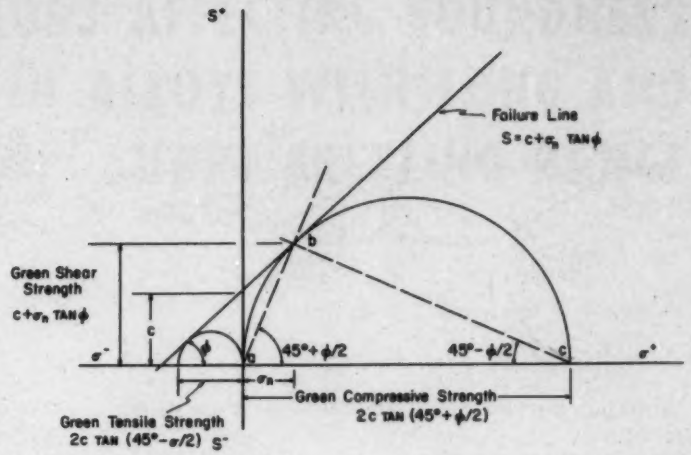



Fig. 2 — Graphical method to determine the surfaces of sliding (45°+Φ/2) and (45°—Φ/2) when sand testing produces values for green compressive and tensile strengths.

## SAND RESISTANCE

For the squeeze board to descend into the flask it is necessary for failure along the logarithmic shaped surfaces of sliding toward a horizontal direction. Only as the sand mixture moves outward to the vertical surface of the flask building up frictional resistance can the squeeze board descend. In this case then the descent of the squeeze board stops when the frictional forces against the flask become equal to the applied squeeze load.

It should be recalled that it is common practice to peen around the perimeter of the flask. However, this location is precisely that wherein the logarithmic surfaces of sliding take place which allows the squeeze board to descend. If peening has been done too near the top of the flask, this prior compaction in this zone will restrict the movement in the zone of radial shear thereby limiting the depth of squeezing.

In Fig. 1, it will be seen that the surfaces of sliding in the radial zone start from the line $aD_A$ at an angle of $(45° + \phi/2)$ and continue to the left. This means that the initial downward direction of the logarithmic surfaces of sliding is steeper as the angle of internal friction $\phi$ increases. The determination of the angle $\phi$ and $(45° + \phi/2)$ and $(45° - \phi/2)$ are readily made if either of the combinations green compressive and tensile strengths or green compression and shear strengths are available.

## MOHR'S CIRCLES

Figure 2 is a schematic representation of Mohr's circles of stress developed when the green compression and tensile strengths are available. Half circles whose radii are one half the values of the green compressive and tensile strengths are constructed, as shown about the $\sigma$ and $s$ axes. A line is drawn so that it is tangent to both half circles, and this line is known as the failure line. Where this line becomes tangent to the compression half circle, this location indicates the green shear strength of the sand mass. Constructing the lines *ab* and *bc* will provide the values for $(45° + \phi/2)$ and $(45° - \phi/2)$.

If the combination of test values for green compression and shear strengths are available, then by the representation in Fig. 3 the desired angles of the surfaces of sliding can be determined. Construct a half circle which represents the green compressive strength. Locate the shear strength value *b* on the half circle and draw the lines *ab* and *bc* by which

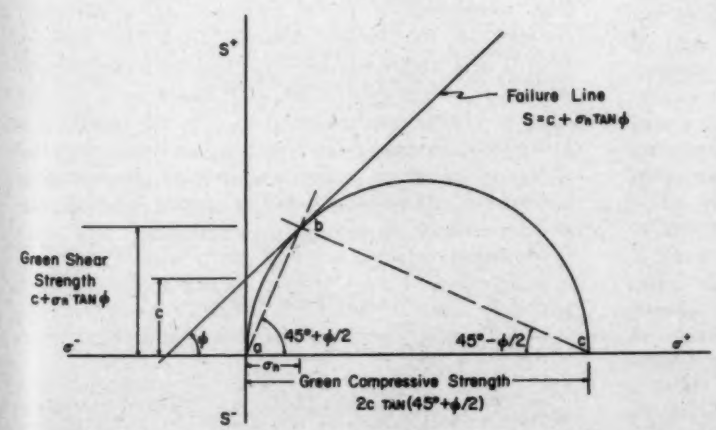



Fig. 3 — Graphical method to determine the surfaces of sliding (45°+Φ/2) and (45°—Φ/2) when sand testing produces values for green compressive and shear strengths.

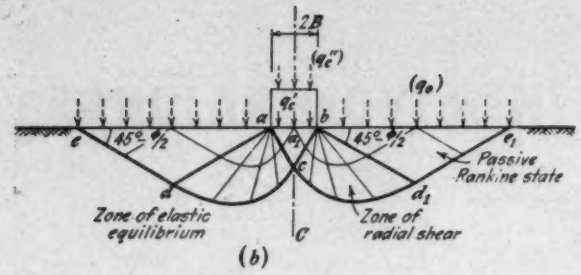



**Fig. 4 — Movement of sand grains under a rectangularly shaped squeeze board (Terzaghi[1]).**

the values for $(45° + \phi/2)$ and $(45° - \phi/2)$ are easily measured.

It would thus seem advisable that the dimensions of a square or round squeeze board should be changed from those that just fit inside the perimeter of the flask. Allowing a clearance of one in. or more between the squeeze board and the inside perimeter of the flask should permit increased horizontal movement of the sand grains in the radial shear zone before the frictional forces against the flask surface becomes equal to the applied load. Also, the squeeze pressure in psi will increase due to the smaller board area used. Consequently, the squeeze board will descend deeper into the flask tending to bring about a greater density toward the parting surface.

Figure 4 is a reproduction of Terzaghi's[1] Fig. 15b. Here will be seen the situation under a rectangularly shaped squeeze board. Note that points *a* and *b* indicate the sides of the squeeze board where the distance *ab* is the short dimension of the rectangular shape. For such a shape there are two zones of radial shear, two zones of the passive Rankine state and one zone *abc* directly under the squeeze board.

## RADIAL SHEAR ZONE

In order for the squeeze board to descend into the flask the sand masses in the radial shear zone must move somewhat horizontally in opposite directions from the center line toward the inside surface of the flask. The usual practice of fitting a rectangular squeeze board to the inside perimeter of the flask means that only a small portion of the zone of radial shear is available to accommodate compaction. It is important to remember that for rectangular squeeze boards there is no movement of grains in the direction of the flask surface of the shorter dimensions. This fact should be considered when laying out patterns on the patten plate and when making a diagnosis of casting defects attributable to molding practice.

Thus, it will be seen that for rectangular flasks wherein the compaction is accomplished by squeezing, it is useless to consider the problem of flowability in a direction parallel to the long dimension of a rectangular squeeze board because there is no movement in that direction. In addition, to compact the sand across the short dimensions of a rectangular flask peening is necessary at such locations.

It seems probable that better squeezing can be accomplished with rectangular boards, if the width of the board was decreased by one in. or more on a side to allow more of a zone of radial shear to be available to produce a greater descent of the squeeze board.

A relationship between soil mechanics and the compaction of a sand mixture can be made if one likens the squeeze board to a footing. Then from Trask and Skjei[2] we find that "considerable data exist to show that the only dimension of a footing having a large effect on the bearing capacity of the soil is the breadth of the footing."

## CONCLUSION

The information presented above shows that the movement of granular materials during the squeeze molding operation is divided between a movement toward a horizontal direction and a movement toward a vertical direction. For the squeeze board to descend the two movements occur simultaneously. Without a horizontal movement below the edges of the squeeze board there will be no descent of the board (or footing). A graphical method has been described which will provide the angles of the surfaces of sliding $(45° + \phi/2$, active Rankine state) and $(45° - \phi/2$, passive Rankine state).

For rectangularly shaped squeeze boards no granular movement is to be expected in a direction parallel to the longer dimension of the squeeze board. It is suggested that better squeeze compaction would be accomplished if the dimensions of the squeeze boards were reduced to take advantage of a greater movement of sand within the zones of radial shear. Since grain movement during squeezing is anisotropic, flowability studies must consider the geometry of the squeeze board (or footing).

## REFERENCES

1. Karl Terzaghi, *Theoretical Soil Mechanics*, Wiley and Sons (1943).
2. Parker D. Trask and Roger E. Skjei, "Pressure-Sinkage Tests on Different Types of Soil," Institute of Engineering Research, Wave Research Laboratory Technical Report, Series 116, Issue 1 (Astia Designation AD — 201449), p. 3 (1958).

# FACTORS AFFECTING SOUNDNESS IN ALLOYS WITH LONG AND SHORT FREEZING RANGE

by J. T. Berry and T. Watmough


## ABSTRACT

After a brief discussion of major mold material and casting alloy variables upon the production of sound castings in pure metals, eutectic composition and long freezing range alloys, a specific study is made of two areas of topical interest. The two areas concerned are:

1) The extent of the influence of mold material upon the feeding requirements of three short freezing range alloys.
2) The effect of applied pressure to the feeder head of castings made in various long freezing range alloys.

In the first area, simple-shaped castings of various configurations have been cast into green sand, oil, sodium silicate and cement bonded sands. Alloys considered were 13 per cent Si-Al alloy (no. 13), gray cast iron of eutectic carbon equivalent and 0.2 per cent carbon steel. The important interrelated effects of mold rigidity, geometric restraint and section thickness upon under riser shrinkage have been rationalized.

In the second area, the extent of interdendritic shrinkage porosity in degassed, long freezing range aluminum alloys cast into 2 in. square and 4 in. square bars has been investigated.

The particular alloys concerned were aluminum alloys no. 43, 220 and 355. Pressures up to 20 psi were applied to the feeder heads of the castings, and the effects of this pressure on the interdendritic shrinkage were recorded.


## INTRODUCTION

Riser dimensioning and positioning, as a means of procuring sound castings, has until recently been either a revered art or a matter of plain guesswork. The number of opinions on the solution of risering problems is probably only outnumbered by the number of publications on the subject. Any overall analysis of such publications would obviously be difficult as well as time consuming. However, it is generally acknowledged that application of the basic laws of heat and mass transfer, already limited by metallurgical variables, has been somewhat impeded by adherence to empirical information. It is the purpose of this paper to critically examine two particular aspects of casting soundness, namely:

1) The extent of the influence of mold material upon the feeding requirements of three short range feeding alloys.
2) The effect of applied pressure to the feeder head of castings made in three long freezing range aluminum alloys.

Although the first topic would appear initially to be connected with heat transfer alone, the families of variables shown in the tabulation are involved.

| Family | Variable | Effect |
|---|---|---|
| Metal | Pouring temperature<br>Metal or alloy type | Liquid contraction |
| | Solidification mechanism<br>Metal or alloy type<br>Gas content | Solidification contraction, or expansion |
| Mold | Interface temperature<br>Moisture or volatile content<br>Intergranular space<br>Material<br>Restraint | Mold contraction or expansion |

The second topic, although appearing to be one of mass transfer alone, is also interconnected with the above variables. However, the dispersed type of unsoundness, known variously as microshrinkage, layer porosity, interdendritic shrinkage and centerline shrinkage, depends particularly on:

1) Intrinsic properties of the liquid metal
   a) Inherent viscosity.
   b) The surface tension existing between the liquid and solid phases.
   c) Range of Freezing.
2) Local thermal conditions
   a) Longitudinal temperature gradient.
   b) Lateral temperature gradient.
3) Pressure differential involving
   a) Metallostatic head.
   b) Applied pressure, if any.
   c) Mode of solidification of the alloy.

Before describing the design and results of the current experiments, it is pertinent that a brief examination be made of the present state of knowledge.

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TABLE 1 — TABLE OF $K_g$ VALUES (CYLINDER-PLATE COMBINATIONS)

| Filling Temp. of Mold C | Filling Temp. of Mold F | Mold Material | $K_g = (M_C/M_R)$ |
|---|---|---|---|
| | | **Silumin Alloy** | |
| 700 | 1300 | Dry sand, medium grain, medium rammed | 1.17 |
| 700 | 1300 | Green sand, medium grain, medium rammed | 1.18 |
| | | **Cast Iron** | |
| 1290 | 2350 | Green or dry sand, coarse grain, hard rammed | 1.18 |
| 1290 | 2350 | Dry sand, fine grain, soft rammed | 1.25 |
| 1200 | 2190 | Green or dry sand, coarse grain, hard rammed | 1.14 |
| 1200 | 2190 | Dry sand, fine grain, soft rammed | 1.20 |
| | | **Low-Carbon Steel** | |
| 1585 | 2890 | Green or dry sand, coarse grain, hard rammed | 1.20 |
| 1585 | 2890 | Diatomaceous earth and fine sand | 1.35 |
| 1485 | 2700 | Green or dry sand, coarse grain, hard rammed | 1.16 |
| 1485 | 2700 | Diatomaceous earth and fine sand | 1.29 |

## UNDER RISER SHRINKAGE PROBLEM IN SHORT FREEZING RANGE ALLOYS

### *Present State of Knowledge*

The attack on the above problem has often left the practical foundry man in such confusion, that a critical evaluation of existing methods of dimensioning risers is required. Examination of previous work indicates that two basic methods have been employed:

a) Those based partly or wholly upon theoretical considerations, i.e., theoretical heat transfer methods.
b) Those based wholly upon empirical findings, i.e., measurement on actual castings to determine minimum riser sizes for particular castings.

It is the considered opinion of the writers that there has been insufficient integration of the two schools.

Referring to the fundamental rules as put forward some years ago by Chvorinov,[1] for risers and castings and relating total solidification time, $t_s$ to volume-to-surface-area ratio V/A, the following difficulties are encountered:

a) General insensitivity of the original criterion:

$$\frac{V_R}{A_R} \geq \frac{V_C}{A_C}$$

where suffixes *R* and *C* represent riser and casting, respectively.

b) Neglect of the effect of differences in shape combination upon solidification times involved in riser-casting combinations.
c) Neglect of the many variables concerned in the expansion-contraction phenomena, within the metal or the mold.

Tentatively, to describe the deviation from a 1:1 ratio of $V_R/A_R$ to $V_C/A_C$, intelligent application of two correction factors could be made. These factors could be designated as $K_g$ (a geometric factor) and $K_s$ (an expansion-contraction sensitive or shrinkage factor).

The mold geometry factor $K_g$ can be predicted. This prediction involves the comparison of quantities known as the mold constant M for particular shapes. This constant is defined by the relationship

$$t_s = M\left(\frac{V}{A}\right)^2$$

$K_g$ is thus, in turn, defined by the square root term in in the equation

$$\frac{V_R}{A_R} = \left[\sqrt{\frac{M_C}{M_R}}\right] \cdot \frac{V_C}{A_C}$$

Values of the square root term, i.e., $K_g$ have been calculated using theoretical data from previous work.[2]

In Table 1 are exhibited typical values of $K_g$ for cylinder-plate combination (The value of $K_g$ for cylinder-sphere combinations is approximately unity).

It may be anticipated that by application of the tabulated $K_g$ values to existing information[3,4] upon riser requirements, some data upon the extent of the $K_s$ value may be determined. The experiments referred to [3,4] were designed to cover a range of shape combinations which deviate from the ideal cylinder-wide plate and cylinder-long bar combinations. Consequently, it is necessary to describe the casting shape in some way—for example, a degree of ranginess, i.e., maximum length divided by thickness.

Collected in Table 2 are observed critical ratios of $V_R/A_R$ to $V_C/A_C$ for the production of sound castings of varying "degree of ranginess." In addition, an apparent $K_s$ factor has been extracted using an approximate $K_g$ value, the above ratio and the equation:

$$\frac{V_R}{A_R} = K_s \cdot K_g \cdot \frac{V_C}{A_C}$$

It will be seen that the apparent $K_s$ value varies with degree of ranginess.

It is suggested that the changing $K_s$ values are to some extent dependent upon the variation of mold wall movement.[5] Quite obviously, certain mold cavity shapes—such as spheres, cubes and thick plates—should possess inherent rigidity characteristics which differ from those of wide plates and long cylinders. Consequently, further information upon the extent and interrelation of the effects of such variables as shape combination, mold cavity shape, mold material, section thickness and metal cast upon feeding requirements is required.

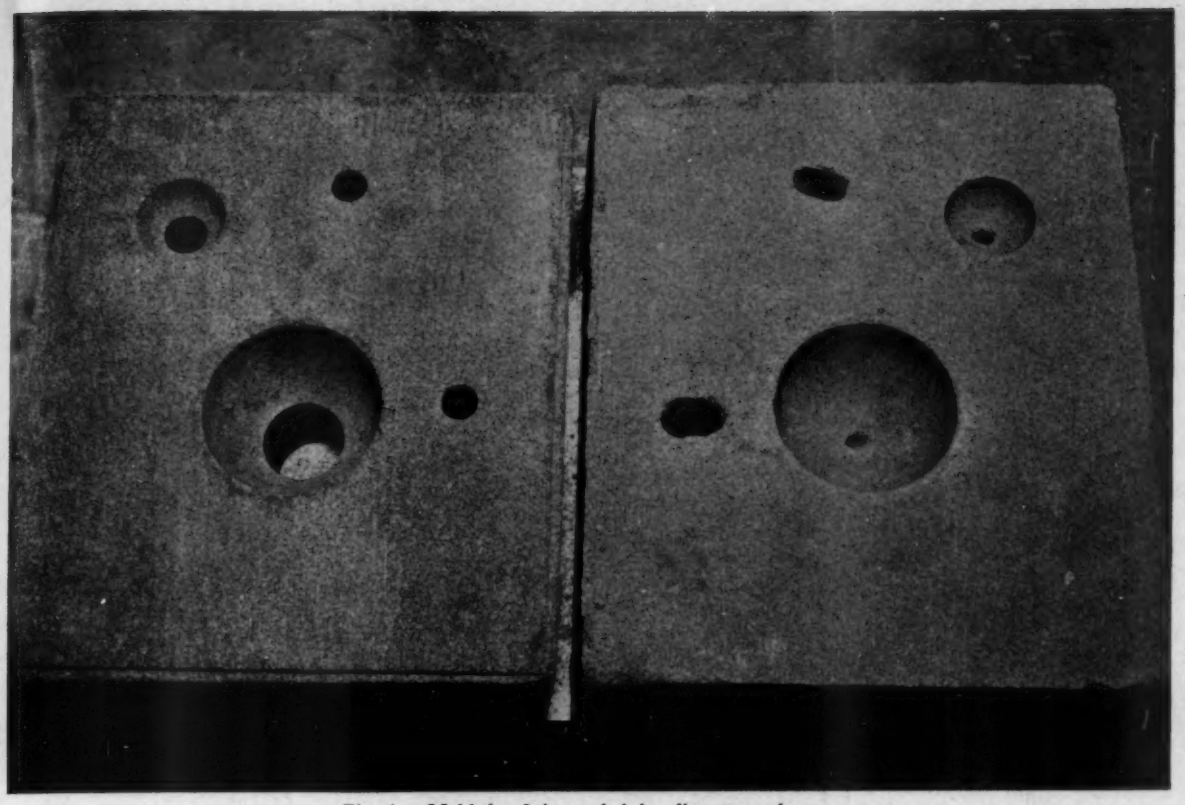


Fig. 1 — Mold for 2 in. and 4 in. diameter spheres.

*Experimental Techniques*

The experiments were initially confined to two commercial eutectic alloys, namely:

1) Aluminum-silicon alloy, no. 13 (13 per cent silicon).
2) Gray cast iron of eutectic carbon equivalent.

In addition, a small number of supplementary experiments were carried out utilizing a 0.2 per cent carbon steel.

Particular combinations of alloy, mold material, mold design and section thickness were covered. The mold materials concerned were:

1) Green sand.
2) Core sand.
3) Sodium silicate bonded sand.
4) Cement bonded sand.

The above represent an increasing order of rigidity. Details of the mold materials are given in Table 3.

Two casting designs were involved in the experiments, plate and sphere, both fed by cylindrical risers. The plate had major dimensions of 10 in. by 13 in. The section thicknesses concerned were 2 in. for the plates and 2 in. and 4 in. for the spheres.

Both shapes were bottom gated, the spheres being horn gated. Figure 1 shows a typical mold for a pair of spheres. The plate was made in the cope, and a horseshoe type gate was employed with the correct ingate-sprue ratio, to ensure even entry of metal into the mold cavity.

TABLE 2 — SHAPE COMBINATION EFFECT ON APPARENT $K_s$ VALUES[4] FOR CYLINDRICAL RISERS

| Casting Shape | Degree of Ranginess† (casting) | Observed Critical Ratio* of $V_R/A_R$ to $V_C/A_C$ | Calculated $K_s$‡ | Apparent $K_s$ |
|---|---|---|---|---|
| Plate | 5 | 1.04 | 1.20 | 0.89 |
| | 6 | 1.10 | 1.20 | 0.92 |
| | 12 | 1.09 | 1.20 | 0.91 |
| | 18 | 1.33 | 1.20 | 1.11 |
| Bar | 2 | 1.01 | 1.20 | 0.84 |
| | 3 | 1.09 | 1.20 | 0.91 |
| | 7.5 | 1.10 | 1.20 | 0.92 |
| | 10.0 | 1.37 | 1.20 | 1.14 |

† Maximum length/thickness.
* Minimum for production of sound castings.
‡ Assuming no corner effects.

TABLE 3 — DETAILS OF MOLD MATERIALS

| Components | Type of Sand: Green sand | Core sand | Sodium silicate bonded | Cement bonded |
|---|---|---|---|---|
| Wedron 5010 sand (W & D), lb | 200 | 300 | 300 | 300 |
| Western bentonite, lb | 12 | 1 | | |
| Cereal binder, lb | | 4 | 4 | |
| Core oil, qt. | | 2 | | |
| Kerosene, pt | 1 | 1 | | |
| Sodium silicate, lb | | | 36 | |
| Cement, lb | | | | 30 |
| Wood flour, lb | 2 | | | |
| Water, % | 4 | | | 5 |
| Mold treatment | None | Baked 400 F 12 hr | Baked 350 F, 12 hr | Air dried 3 days |

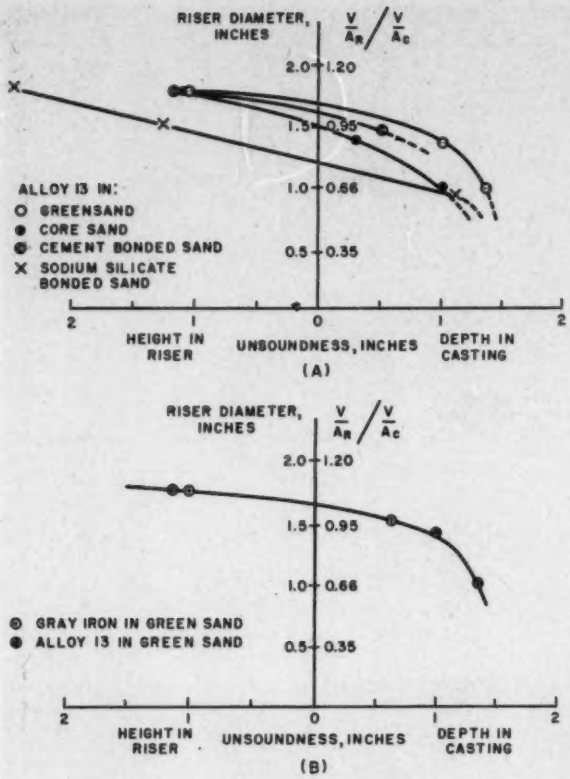



**Fig. 2 — Shrinkage extent into castings made from aluminum alloy 13 and eutectic equivalent gray iron. The influence of various sand conditions on 2 in. diameter spheres is shown. All risers 4 in. high.**

The metal was melted in a high frequency induction furnace under carefully controlled conditions.

The aluminum alloy was melted directly from high grade ingot stock. Vigorous chlorine degassification was employed to remove all possible traces of hydrogen.

The gray cast iron was produced from a malleable pig, additions being made to the charge to give a composition equivalent to that of the eutectic.

Both the cast iron and the steel, made from virgin stock, were melted under an argon cover to minimize oxidation.

Pouring temperatures were maintained as follows:

| | | |
|---|---|---|
| Aluminum alloy | — | 1300 F ( 700 C) |
| Gray cast iron | — | 2475 F (1360 C) |
| 0.2 per cent Carbon steel | — | 2950 F (1620 C) |

The castings were subsequently sectioned in several planes, to determine the extent of under riser shrinkage. Measurements were also made of casting thickness to gage the mold dilation. In addition, supplementary measurements of the effect of mold dilation upon unfed aluminum bronze plates involved in previous work were made.[2] The critical ratio of $V_R/A_R$ to $V_C/A_C$ was determined by plotting this measured ratio against depth of unsoundness for particular riser combinations, e.g., cylinder feeding sphere. Figure 2 exhibits typical results for gray iron and aluminum alloy no. 13 cast into various molds. This method has been proposed previously.[6]

### *Results of Experiments*

Given in Tables 4 to 8 are inferred $K_s$ values for the experiments performed. These values were obtained by dividing the experimentally determined product ($K_s \cdot K_g$) by the approximate value of $K_g$, as calculated for plate-cylinder and sphere-cylinder combinations, without corner corrections. In the instances of the few 4 in. cubes cast it was deemed necessary to attempt to account for the corner effects using the Ruddle-Skinner correction.[7]

In Tables 4 and 5 the effects of mold shape are shown for 13 per cent silicon-aluminum alloy (no. 13) and the gray iron in green sand molds.

**TABLE 4 — MOLD SHAPE EFFECT ON $K_s$ VALUE FOR ALUMINUM ALLOY NO. 13 IN GREEN SAND MOLDS**

| Shape | Section Thickness, in. | Observed Critical Ratio* of $V_R/A_R$ to $V_C/A_C$ | Calculated $K_g$ | Inferred $K_s$ |
|---|---|---|---|---|
| Sphere | 2 | 1.05 | 1.00 | 1.05 |
| Sphere | 4 | 0.82 | 1.00 | 0.82 |
| Cube | 4 | 0.92 | 1.00 | 0.92 |
| Plate | 2 | 1.04 | 1.17 | 0.89 |

* Equal to $K_g \cdot K_s$.

The 2 in. spheres in both alloys require the most feeding (as reflected by the large $K_s$) while the 4 in. spheres require the least. Intermediate values of $K_s$ are held by the 2 in. plates and also the 4 in. cubes (The latter were determined for silumin alloy only).

Table 6 indicates the interdependence of $K_s$ on both mold material and mold shape. The rigidity increase brings the $K_s$ value down to 0.71 with the 2 in. diameter spheres in the silicon-aluminum alloy no. 13.

In plate castings the sodium silicate bonded sand deviated from this trend. Examination of the casting surface and the thickness profile indicated that the stress configuration for this mold shape, in combination with this particular mold material, had caused

a general mold cracking and irregular wall movement.

TABLE 5 — MOLD SHAPE EFFECT ON $K_s$ VALUE FOR CAST IRON* IN GREEN SAND MOLDS

| Shape | Section Thickness, in. | Observed Critical Ratio** of $V_R/A_R$ to $V_C/A_C$ | Calculated $K_g$ | Inferred $K_s$ |
|---|---|---|---|---|
| Sphere | 2 | 1.05 | 1.00 | 1.05 |
| Sphere | 4 | 0.79 | 1.00 | 0.79 |
| Plate | 2 | 1.20 | 1.18 | 1.02 |

* Eutectic carbon equivalent.
** Equal to $K_g \cdot K_s$.

TABLE 6 — PARTIAL DEPENDENCE OF $K_s$ ON MOLD SHAPE AND MATERIAL FOR ALUMINUM ALLOY NO. 13

| Shape | Mold Material (Table 3), sand type | Section Thickness, in. | Observed Critical Ratio* $V_R/A_R$ to $V_C/A_C$ | Calculated $K_g$ | Inferred $K_s$ |
|---|---|---|---|---|---|
| Sphere | Green | 2 | 1.05 | 1.00 | 1.05 |
| Sphere | Cement bonded | 2 | 1.00 | 1.00 | 1.00 |
| Sphere | Core | 2 | 0.96 | 1.00 | 0.96 |
| Sphere | Sodium Silicate bonded | 2 | 0.71 | 1.00 | 0.71 |
| Plate | Sodium Silicate bonded | 2 | 1.09 | 1.18 | 0.93 |
| Plate | Green | 2 | 1.08 | 1.17 | 0.89 |
| Plate | Cement bonded | 2 | 0.94 | 1.18 | 0.80 |

* Equal to $K_g \cdot K_s$.

With the 2 in. plates in gray iron (Table 7), a change in rigidity due to drying out of the mold occurs during the early part of the solidification in green sand. This has probably led to the feeding requirements being similar to those in cement bonded sand.

TABLE 7 — MOLD MATERIAL EFFECT ON $K_s$ VALUE FOR GRAY CAST IRON CAST INTO 2 IN. PLATES

| Mold Material, sand type | Observed* Critical Ratio $V_R/A_R$ to $V_C/A_C$ | Calculated $K_g$ | Inferred $K_s$ |
|---|---|---|---|
| Green | 1.20 | 1.18 | 1.02 |
| Cement bonded | 1.20 | 1.18 | 1.02 |

* Equal to $K_g \cdot K_s$.

Table 8 gives comparison between 0.2 per cent steel and the 13 per cent silicon-aluminum alloy no. 13, cast as 2 in. diameter spheres in molds of various rigidities once again. It would appear that with the higher interface temperature material, early rigidity changes occur. These effects will be considered more fully in the discussion.

TABLE 8 — $K_s$ VALUES FOR 0.2 PER CENT CARBON STEEL AND ALUMINUM ALLOY NO. 13 IN VARIOUS MOLDS FOR 2 IN. DIAMETER SPHERES

| Alloy | Mold Material, sand type | Observed Critical Ratio* $V_R/A_R$ to $V_C/A_C$ | Calculated $K_g$ | Inferred $K_s$ |
|---|---|---|---|---|
| Steel | Green | 1.08 | 1.00 | 1.08 |
| Steel | Cement bonded | 1.08 | 1.00 | 1.08 |
| Aluminum | Green | 1.05 | 1.00 | 1.05 |
| Aluminum | Cement bonded | 1.00 | 1.00 | 1.00 |
| Aluminum | Core | 0.96 | 1.00 | 0.96 |
| Aluminum | Sodium silicate bonded | 0.71 | 1.00 | 0.71 |
| Cast Iron | Green | 1.05 | 1.00 | 1.05 |

* Equal to $K_g \cdot K_s$.

### *Discussion and Conclusions*

The wide range of values for the $K_s$ shrinkage factor, contained in Tables 4 to 8, obtained for the short freezing range alloys indicates a strong dependence of feed metal requirements upon mold material, casting shape, section thickness and solidification temperature.

In general terms, provided no cracking of the mold material occurs, the more rigid the mold, the less the feeding requirements. If drying out occurs there is, for a particular mold cavity shape, little or no difference between the feeding requirements of castings produced in what initially were green sand molds as compared to those made in what were initially more rigid molds.

For a constant mold material, when mold shape is varied from a sphere to a plate, a change in feed metal requirement occurs. For casting solidification times of similar order (compare 2 in. plate, 4 in. sphere in green sand molds) more feed metal is demanded by the castings produced in a mold with the least restrained walls, i.e., plates.

Although only the bonding material of one particular sand has been varied in the present work, the effect of changes in the basic mold material will also be important. Figure 3 shows an indication of thickness profiles of unfed eutectic aluminum bronze plates poured into molds made with sands of varying coarseness and ramming density.

Finally, from the data currently presented, it would also appear that the sensitivity of feed metal requirement to differences in mold materials falls off with increasing solidification temperature and/or increasing section thickness of the alloy poured.

Before any further claims are made regarding a universal theory for the feeding of short freezing range alloys, full acknowledgment must be made of the aforementioned effects. These effects have not been taken into account by previous workers. Somewhat unsuitable graphical representations of data have therefore appeared. Ideally, there should exist a straight line relationship between the volume-to-surface-area ration $V_R/A_R$ of the riser to that of the casting $V_C/A_C$ which passes through the origin; the slope will depend upon the shape combination.

In practice, because of the limitations imposed by

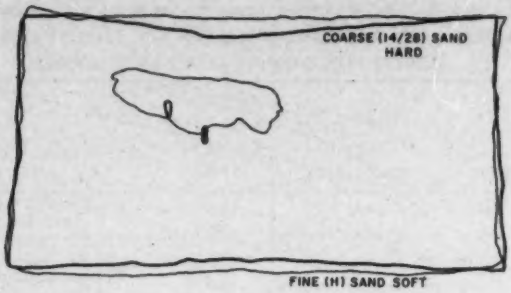
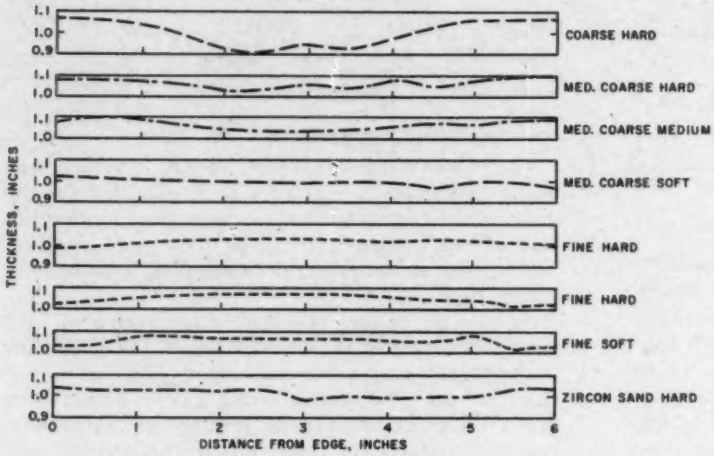



Fig. 3 — Mold material effect on casting shape (aluminum bronze eutectic).




the mold dilation effects, a ready, simple relationship cannot exist. However, it is reasonable to expect that a collection of calculated $K_g$ and inferred $K_s$ values can be steadily built up from past work as well as future data. This would be, in effect, determining the parts of particular straight lines adhered to for the feeding requirements of given families of castings. Recourse to the literature[4] indicates that this treatment need not be limited to simple-shape combinations but that the effects of appendages, etc., can also be accounted for (Table 9).

TABLE 9 — APPENDAGES EFFECT ON SHRINKAGE FACTORS[4]

| Appendages | Appendage thickness, in. | Critical ratio, $V_E/A_E/V_C/A_C$ (parent casting) | $K_g$ | Apparent $K_s$ |
|---|---|---|---|---|
| | | **4 x 4 x 30 in. Bar** | | |
| Plate | ½ | 1.16 | 1.2 | 0.97 |
| | 1 | 1.13 | 1.2 | 0.94 |
| | 2(longitudinal) | 1.38 | 1.2 | 1.15 |
| | 2(transverse) | 1.32 | 1.2 | 1.10 |
| Bar | 1 | 1.05 | 1.2 | 0.87 |
| | 2 | 1.06 | 1.2 | 0.88 |
| | 3 | 1.10 | 1.2 | 0.92 |
| | | **4 x 20 x 20 in. Plate** | | |
| Plate | ½ | 1.15 | 1.2 | 0.96 |
| | 2 | 1.28 | 1.2 | 1.07 |
| | 3 | 1.35 | 1.2 | 1.12 |
| Bar | 2 | 1.15 | 1.2 | 0.96 |
| | 4 | 1.26 | 1.2 | 1.05 |
| | 6 | 1.30 | 1.2 | 1.08 |

NOTE: $K_g$ as calculated does not take into account appendages or corner effects.

## DISPERSED SHRINKAGE PROBLEM

### *Present State of Knowledge*

According to Baker[8] long freezing range alloys solidify in sand molds in a pasty fashion and are fed in two distinct phases:

1) Mass feeding. This involves the sinking and transport of semisolid material within the casting. Transport is clearly influenced by the metallostatic and atmospheric or applied pressure.

2) True liquid metal feeding through the tortuous paths that occur as a result of the solidification mode and mass feeding.

It has been further suggested by Cibula and Jay[9] that feeding persists until at least 60 per cent of the metal has solidified.

From the variables treated previously an impression of the methods open to attack may be gained:

1) Inherent properties of the metal during solidification are obviously fixed by choice of alloy system employed for the casting.

2) Local thermal conditions have been manipulated by Johnson, Pellini and Bishop[10] by the use of chilling techniques.

3) Effective pressure differential is dependent only, in normal casting techniques in a particular alloy at a particular temperature, on metallostatic head.

An essential feature of all the feeding methods adopted is the use of the metallostatic and atmospheric pressure to force feed the liquid metal into the casting. The only commercial methods at present where pressures greater than atmospheric are used are centrifugal casting and pressure die casting. The soundness and integrity of centrifugal castings is well known and, in view of the success of this technique, it should be possible to produce sounder castings by normal foundry techniques by the application of feeding pressures much in excess of those presently employed.

Both phases of feeding will clearly be affected by increased feed pressure.

The use of additional agents to procure relatively high pressures in feeder heads of steel castings has been reported by a few investigators. This recently reported work has been performed principally in Eastern Europe or the Soviet Union. Desnizki[11] employed compressed air at pressures up to 70 psi on the feeder head of steel castings approximately 10 ft long by 15 in. wide by 2½-in. thick. Jazwinski[12] has adopted a sophisticated self-pressurizing system involving a check valve incorporated into the down sprue. The linseed oil based pressurizing charge is closely similar in concept to Pearson's[13] refractory-coated lead nail. The application of steam feeding to large gray iron castings has, of course, been employed in certain iron foundries for a number of years.

### *Long Freezing Range Alloys*

The effects of the application of pressure to long freezing range aluminum alloy castings during solidification has been studied by a number of workers. Hanson and Slater,[14] Parker[15] and Uram,[16] with their respective co-workers, have been prominent in this field. Although the work of Parker, etal, and Uram, etal, was performed in autoclaves under approximately equivalent conditions, the results obtained from physical tests of the castings were not in agreement. Specifically, Parker reported a 10 per cent increase in ultimate tensile strength and an increase of 30 per cent in elongation in high purity 220 aluminum alloy, well degassed with chlorine. However, Uram, working with aluminum alloys 356 and 195, reported no increases in either property concurrent with the application of pressure. Neither of the techniques adopted appear to take full advantage of the effect of applied pressure upon mass feeding. Ideally, as previously stated, a pressure differential should be promoted to effect markedly the mass transport of the semisolid feed metal.

In addition, the long freezing range alloys, normally assumed to be nonskin formers, under certain metallurgical conditions, can effectively behave partially at least as skin forming alloys. Cibula[17] observed that a transition from skin forming to pasty solidification is strongly dependent on pouring temperature when gun metal and bronzes are cast into the sand molds—high pouring temperatures lead to skin formation—presumably reflecting the reduction in number of nuclei present in the pouring stream.

### *Experimental Techniques*

Three aluminum alloys selected for the investigation were:

| Alloy No. | Composition, % | | | | |
|---|---|---|---|---|---|
| | Cu | Fe | Si | Mg | Ti |
| 43 | 0.00 | 0.36 | 4.89 | 0.01 | 0.12 |
| 220 | 0.13 | 0.07 | 0.09 | 10.14 | 0.09 |
| 355 | 1.17 | 0.25 | 4.90 | 0.45 | 0.10 |

The respective freezing ranges of these three alloys[18] are:

| Alloy No. | Liquidus, F (C) | Solidus, F (C) |
|---|---|---|
| 43 | 1165 (629) | 1070 (577) |
| 220 | 1150 (621) | 840 (499) |
| 355 | 1160 (627) | 1075 (579) |

Two different castings were chosen for the work, one to be 2 in. square in cross-section by 14 in. long and the other to be 4 in. square by 26 in. long. If castings of these dimensions are run and fed from one end, then shrinkage porosity will certainly occur when cast in the above aluminum alloys.

Figure 4 shows the pattern for the 2 in. square cross-section casting. The molds were made in oil-bonded core sand. Figure 5 shows a schematic view of the casting. The plaster-of-paris cap which is used for retaining gas pressure on the feeder can be seen. Shown in Fig. 6 is an aluminum alloy casting produced without pressure application to the head.

The aluminum alloys were melted in a 70 lb aluminum capacity carbon/clay crucible surrounded by a high frequency drop coil. The absorption of hydrogen from water vapor or other sources is normally negligible in this type of melting; nevertheless, the melts were vigorously degassed with dry chlorine gas admitted to the melt by means of a submerged refractory tube. The casting temperature for the 4 in. square cross-section test bars was 1290 to 1310 F and for the 2 in. square cross-section test bar, 1310 to 1330 F.

Standard vacuum degassification tests were conducted on samples of the liquid metal from selected melts to check the efficiency of the chlorine degassing operation.

Dry argon gas was utilized for pressurizing the feeder head. In the initial experiments, considerable difficulty was experienced because of extreme metal penetration into the oil-bonded core sand molds. Figure 7 illustrates the severity of this metal penetration. To overcome this effect a colloidal silica mold wash was applied to the mold faces; the application of gas pressure was delayed for 30 sec in the case of the 2 in. square test bar and 2 min in the case of the 4 in. square test bar. By employing these techniques cast-

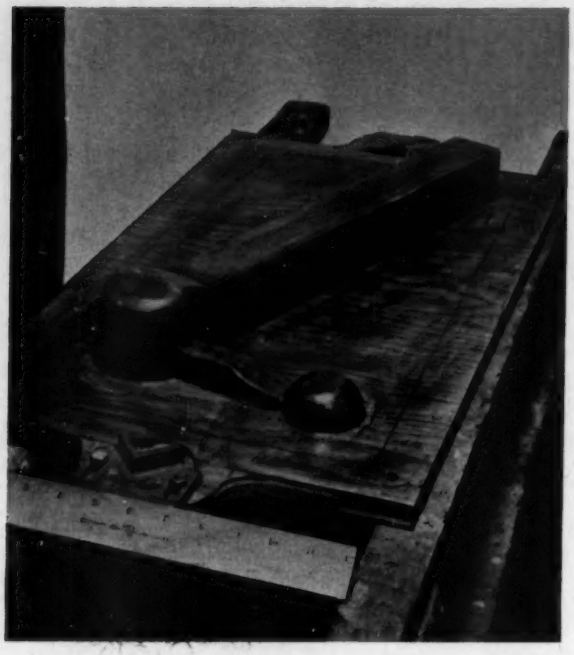


Fig. 4 — Pattern for 2 in. square cross-section test casting.

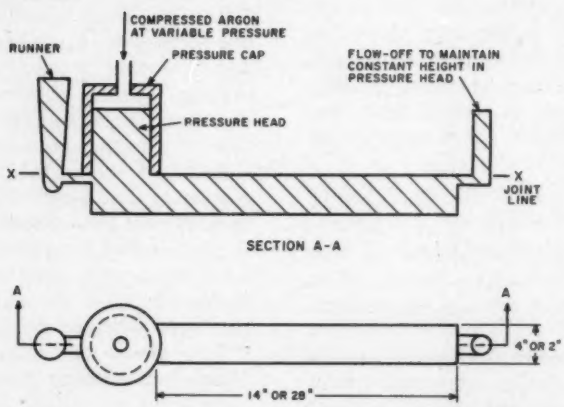



Fig. 5 — Schematic view of test casting with pressure cap.

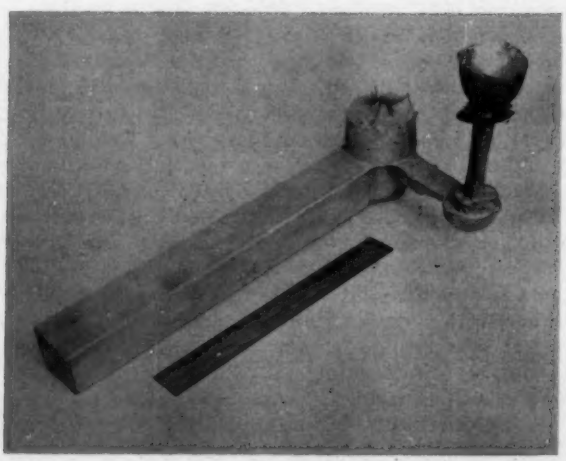


Fig. 6 — Trial 2 in. square cross-section test casting (no pressure applied).

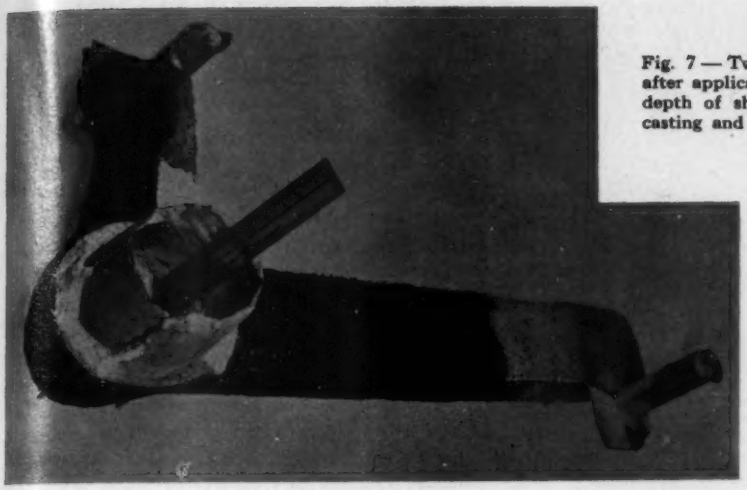


Fig. 7 — Two in. square cross-section aluminum casting after application of 20 psi pressure to feed head. Note depth of shrinkage cavity in feeder and swelling of casting and metal penetration of sand.

ings were produced which had satisfactory surfaces and shapes.

To evaluate the performance of the technique, 1/8-in. cross-sectional slices were machined from the bars at 2 in. intervals along their length. These slices were subjected to careful radiographic examination.

### *Results of Experiments*

Illustrated in Fig. 8 are two views of sectioned feeder heads. The heads were taken from 2 in. square castings produced in aluminum alloy no. 43. The feeder head shown in Fig. 8a was taken from a casting which had an indicated pressure of 10 psi dry argon gas applied to it. The feeder head shown in Fig. 8b was taken from a casting which did not have any supplementary pressure applied.

Both of the feeder heads or risers were surrounded by a plaster-of-paris sleeve to retard solidification and ensure that the metal in the risers did not solidify before the castings proper.

The effect of the application of pressure can readily be observed. The inner surfaces of the shrinkage cavity in the pressure head had an appearance similar to that normally associated with hot tear faces, indicating that semisolid or pasty feed metal had been forced by the applied pressure in the casting.

### *Aluminum Alloy No. 43*

As previously stated radiographic examination of cross-sectional slices taken from the 2 in. and 4 in.

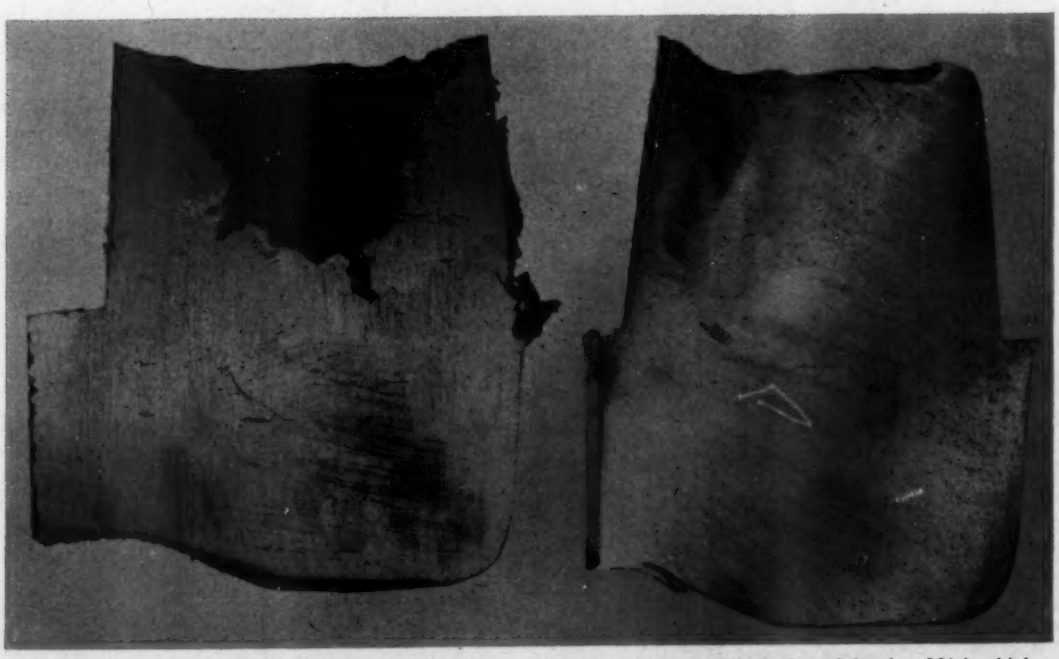


Fig. 8 — Section feeder heads from two aluminum alloy 43 castings (original size of head — 3¼-in. high above cope face of casting, 3½-in. diameter). *Left* — 10 psi applied pressure. *Right* — no applied pressure.

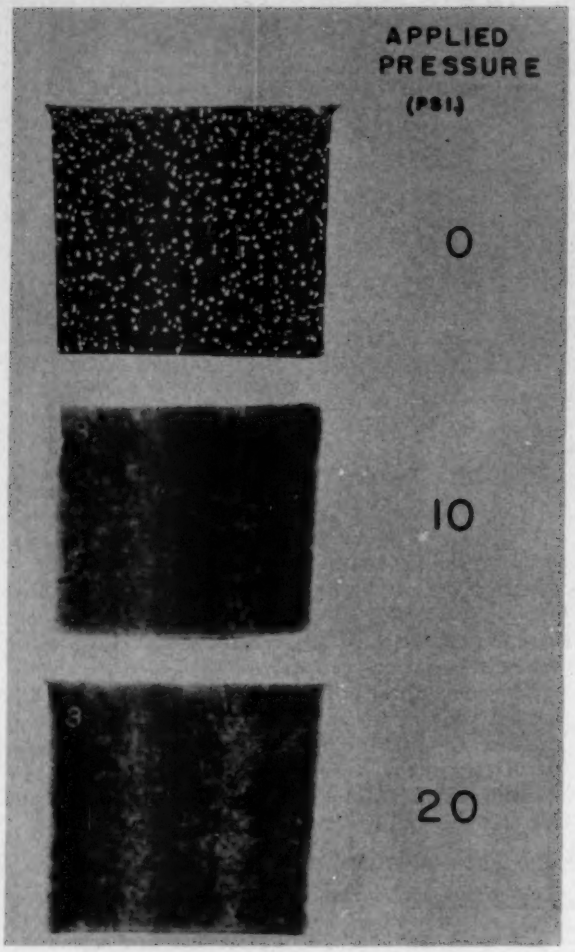



Fig. 9 — Radiographs of slices taken from the center section of 2 in. square cross-section casting in aluminum 43.

square bars was used as the criterion for the success or failure of the pressure application to the feeder head. It was considered that the area or zone which would be most prone to dispersion shrinkage would be approximately in the center of the bar. Accordingly, exhibited in Fig. 9 are the radiographs of ⅛-in. slices taken at sites 8 in. from the end of three bars. The upper radiograph relates to the slice from the casting produced without any applied pressure.

The center radiograph was taken from a casting which has 10 psi indicated argon gas pressure applied to it, and the lower radiograph taken from a slice of the casting with 20 psi indicated argon gas pressure. The effect of the gas pressure can readily be observed. All traces of interdendritic porosity have been eliminated by the application of 10 psi argon gas to the feeder head. Shown in Fig. 10 are radiographs taken from slices which originated at a position 12 in. from the end of the 4 in. square test bar castings.

The upper radiograph is taken from a casting with no supplementary pressure applied; the center radiograph is with 10 psi applied pressure and the lower radiograph is with 20 psi applied argon pressure. The results obtained here are not so spectacular as those obtained from the 2 in. square bar, but the reduction in shrinkage is readily apparent.

*Aluminum Alloy No. 220*

Figure 11 shows a similar series of radiographs to those exhibited in Fig. 9, except that the aluminum alloy investigated was no. 220, which has an extremely long freezing range.

The reduction and elimination of interdendritic shrinkage can be seen from these radiographs.

Figure 12 shows radiographs taken from the central portion of two 4 in. square cross-section castings. The effect of 20 psi indicated gas pressure upon the incidence of dispersed shrinkage is apparent. The upper radiograph relates to a casting produced without benefit of supplementary pressure, and the lower one after the application of 20 psi argon gas during solidification.

*Aluminum Alloy No. 355*

Exhibited in Fig. 13 are radiographs of ¼-in. slices taken from 2 in. square cross-section castings made from Alloy No. 355. The upper radiograph relates to a casting produced by ordinary techniques, the center radiograph with 10 psi applied argon pressure and the lower one with 20 psi applied argon pressure. Again, the reduction in shrinkage is apparent when supplementary gas pressure is applied to the feeder head.

Figure 14 shows the effect of applying up to 20 psi indicated argon pressure to the 4 in. square cross-section castings. Although the interdendritic shrinkage is still present in the casting which had the benefit of pressure application, the severity of this defect is considerably reduced.

*Conclusion and Discussion*

The success of the pressurizing technique on feeder heads of the selected long freezing range aluminum alloys, indicates that this method could be used for the production of castings with complete structural integrity.

Techniques which involve the deliberate partial gassification of aluminum alloys to reduce feeding problems cannot make this claim. This type of defensive approach is hardly desirable.

Similarly, the approach adopted in some bronze and gun metal foundries, involving high pouring temperatures of fully degassed metal, leading to casting sections, possessing only two or three grains per cross-section, should be deplored.

While the application of this technique does involve certain difficulties connected with breakout and metal penetration affects, it is considered that, for the production of the best quality castings in long freezing range copper and aluminum based alloy, this approach could profitably be investigated. Further, the possibility does exist that a self-pressurizing feeder head could be made wherein a plug of gas producing material, such as hexachlorethane, is located.

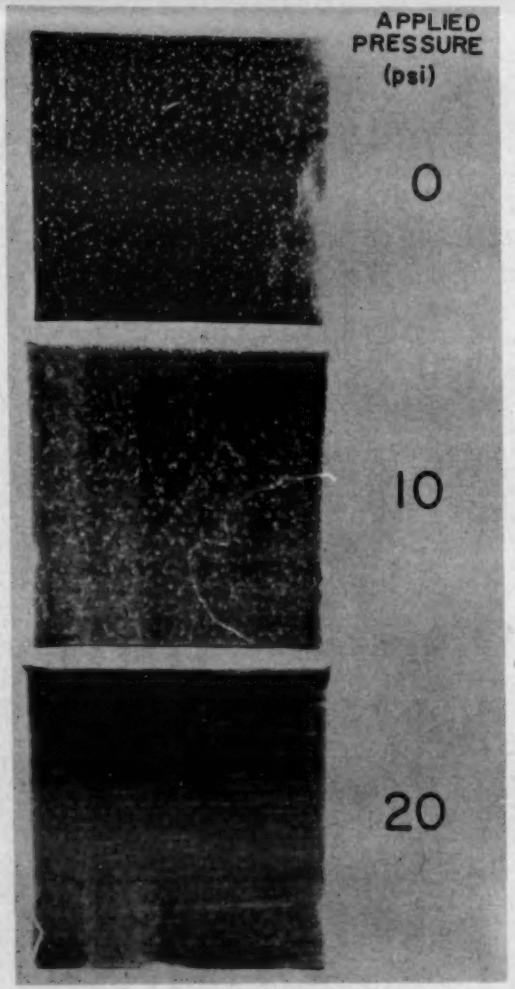



Fig. 10 — Radiographs of slices taken from the center of 4 in. square cross-section bar casting in aluminum 43.

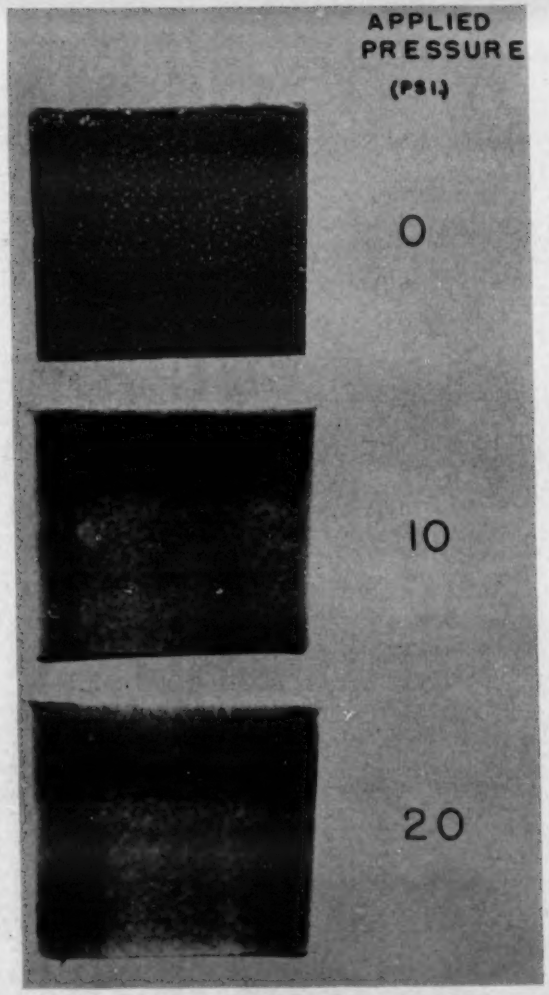



Fig. 11 — Radiographs of slices taken from center of 2 in. square cross-section bar casting in aluminum 220.

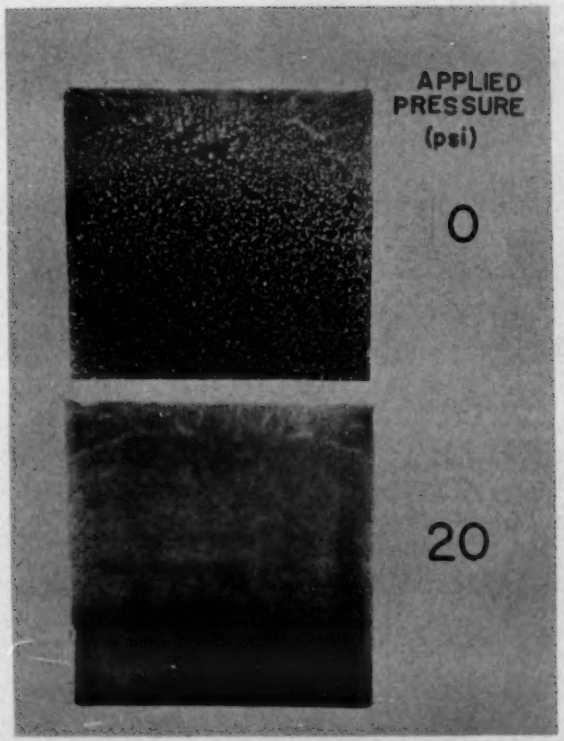



Fig. 12 — Radiographs of slices taken from center section of 4 in. square cross-section bar casting in aluminum 220.

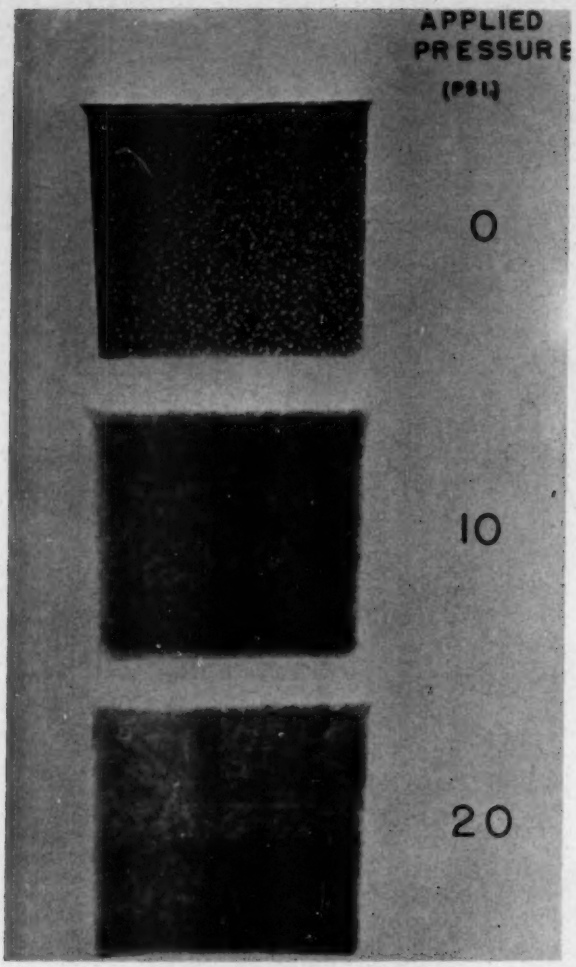



Fig. 13 — Radiographs of slices taken from center section of 2 in. square cross-section bar casting in aluminum 355.

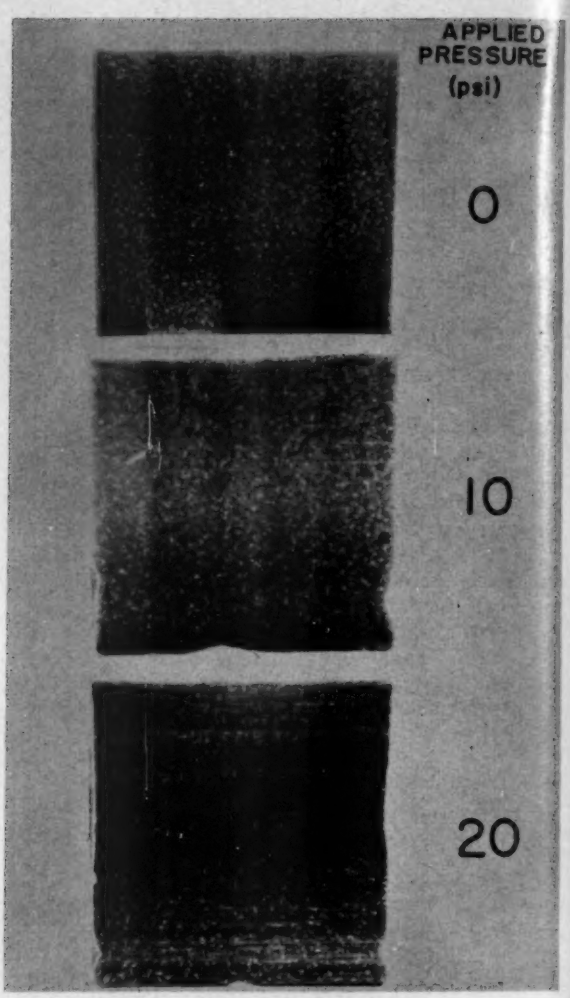



Fig. 14 — Radiographs of slices taken from center section of 4 in. square cross-section bar casting in aluminum 355.

Finally the application of the technique to skin forming alloys such as steels, aluminum bronzes, etc., represents a highly desirable means of minimizing centerline shrinkage in thin section castings where conventional feeding ranges are exceeded.

## SUMMARY

Any rationalized approach to the problem of precise feeding of short freezing range alloys will have to account for the mold, as well as heat transfer, variables described in the present work such as mold cavity shape, mold materials, section thickness and metal cast. Previous methods of dimensioning have oversimplified this problem.

In the production of castings of thick sections and high integrity, due consideration to the application of supplementary feed pressure to the risers should be given. Although experimental difficulties may be encountered, these should not prove insurmountable.

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# CHILLING BEHAVIOR OF A ROLL-TYPE CAST IRON

## *Some factors affecting*

by K. E. Pinnow and R. W. Lindsay


### ABSTRACT

**Superheating temperature, pouring temperature and charge materials effects on the chill characteristics of a roll-type cast iron were evaluated by making chill castings from melts of constant chemical composition. Superheating temperatures and pouring temperatures were precisely measured by immersion thermocouples.**

**An increase in superheating temperature from 2600 to 2900 F (1427 to 1593 C) increased the chill capacity of melts poured at 2450 F (1343 C). A decrease in pouring temperature from these superheating temperatures to 2450 F reduced chill depth. Differences in melting stock had a marked effect on the chilling behavior of melts heated to low superheating temperatures (2600 F), but had less influence on those superheated to high temperatures (2900 F).**

**Superheating temperature did not markedly affect the structure of the white iron and gray iron portions of the chill test castings. Lowering the pouring temperature from the superheating temperature to 2450 F produced noticeable changes in the white iron structure.**

**The exact causes for the changes in chill behavior produced by these variables were not determined. However, they were not a direct result of differences in customary chemical composition nor perhaps in gas content between melts superheated to different temperatures.**


### INTRODUCTION

The chilling behavior of cast iron is of prime concern to producers of chilled cast iron rolls, as well as to all foundrymen involved in the manufacture of chilled castings. Chemical composition has a pronounced effect on this behavior, and the influences of the elements commonly found in cast iron have been well established. Elements such as carbon, silicon and nickel retard the formation of carbide during solidification; and those such as chromium, manganese and sulfur promote it.

Hydrogen and nitrogen are considered to behave like the latter elements, whereas oxygen has been reported to have a carbide-stabilizing influence, a graphitizing influence, or perhaps no effect at all.[1-4]

Other important, but less obvious factors affecting chilling behavior are superheating temperature, pouring temperature and the nature of the charge materials. Increased superheating temperature has been observed to increase the depth of white iron formed in chilled castings poured at identical temperatures, while reducing the casting temperature below the superheating temperature has been found to decrease it.

Melts produced from different charges often exhibit differences in chilling behavior even though their customary chemical composition is similar. The exact explanations for these effects have not been determined, but they have been attributed variously to differences in the amount of graphite in the charge (graphite nuclei theory), variations in the hydrogen and oxygen contents of the melt (gas content theory) and the presence or absence of certain nonmetallic nuclei (silicate-slime theory).[5-7]

Much of the work in which these aforementioned results were obtained is subject to question on the basis that the chemical composition of the experimental melts was not kept constant, nor were the superheating and pouring temperatures accurately controlled or measured. The purpose of this investigation was to determine the influence of superheating temperature, pouring temperature and charge materials on the chilling behavior of a roll-type cast iron under conditions of constant chemical composition and exact temperature control.

### MATERIALS AND EXPERIMENTAL PROCEDURE

Roll iron, low-sulfur pig iron, graphite, ferrosilicon (47 to 51 per cent silicon) and ferromanganese (70 to 80 per cent manganese) were used to prepare the experimental melts. The roll iron was obtained from castings poured from two different roll heats. The castings were made in green sand molds with a diameter of 10 in. and a depth of 40 in., and thus corresponded closely to actual roll castings. The chemical

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compositions of the roll iron and the pig iron are given in Table 1.

TABLE 1— CHEMICAL COMPOSITIONS OF CHARGE MATERIALS

| | Weight, % | | | | | | |
|---|---|---|---|---|---|---|---|
| | TC* | GC** | Si | Mn | P | S | Mo |
| Roll Iron, Lot No. 1 | 3.30 | 1.67 | 0.62 | 0.24 | 0.450 | 0.110 | 0.37 |
| Roll Iron, Lot No. 2 | 3.30 | n.a. | 0.74 | 0.25 | 0.444 | 0.107 | 0.34 |
| Pig Iron | 4.12 | 3.08 | 0.97 | 0.27 | 0.62 | 0.058 | n.a. |

*TC — total carbon.
**GC — graphitic carbon.
n.a. — not analyzed.

The basic charge for the melts used to study the effects of superheating temperature and pouring temperature was 40 lb of roll iron from lot No. 1 and 60 lb of roll iron from lot No. 2. Depending on the superheating temperature involved, different amounts of ferromanganese, graphite and ferrosilicon were added to such a charge to adjust the manganese content and to compensate for the losses of carbon and silicon which occurred during melting. These additions were made early during meltdown to avoid inoculation effects.

The heats produced to study the influence of charge materials were made from different proportions of roll iron, pig iron, ferrosilicon and ferromanganese. In these melts the carbon and silicon contents were controlled by varying the amounts of pig iron and ferrosilicon. The manganese content was adjusted by making small additions of ferromanganese.

The experimental melts were made in a 100 lb capacity induction furnace, lined with a commercial magnesia crucible. Magnesia crucibles were used in order to minimize metal-crucible reactions. Subsequent chemical analyses of two melts, one superheated to 2600 F (1427 C) and the other to 2900 F (1593 C), indicated that the magnesium contents of both melts were quite low (less than 0.001 per cent). Thus, it was assumed that magnesium contamination was not a factor in these experiments.

### *Melting Procedure*

The melting procedure for each melt consisted of placing one-half of the charge into the crucible. The unit was then energized with an input that was kept constant until the desired superheating temperature was reached. After the first portion of the charge began to melt, more roll iron or pig iron was added until about 20 lb of material remained uncharged.

At this point the metal was allowed to become completely molten, and then, in the following order, additions of ferrosilicon, ferromanganese and graphite were made (Graphite additions were not made to the melts prepared with pig iron). When these materials had dissolved, the remainder of the roll iron or pig iron was added. The melt was then heated to the superheating temperature and held at this temperature for 15 min.

The temperature of the melt was measured frequently during the holding period, as well as throughout the entire melting cycle, by an immersion thermocouple. Changes were made in the power input if necessary to maintain the temperature. It was difficult to keep the temperature constant by this technique, but temperature fluctuations were kept within ± 20 F of the superheating temperature.

At the end of the holding period the furnace was shut off, and in the following order a sample for chemical analysis and two chill tests were poured. After this, the metal which remained in the crucible was allowed to cool to 2450 F (1343 C) and another sample for chemical analysis and two more chill tests were cast. The chill test castings poured at the superheating temperature were designated as *A* and *B*; those cast at 2450 F (1343 C) as *C* and *D*.

The chill behavior of any melt was determined by pouring the metal directly from the furnace into a vertical, dry sand mold fitted with a horizontal, ingot iron chiller plate. The casting was shaken out 30 min after pouring, cooled to room temperature and fractured for visual examination. The depth of white iron and the character of the mottle in the castings were then used to evaluate chill behavior. The design of the chill test casting was the same as that used by the British Cast Iron Research Association.[8] Its shape and dimensions are shown in Fig. 1.

## EXPERIMENTAL RESULTS

The chemical compositions of the chill tests, and their pouring temperature together with the superheating temperature of the melt from which they were cast, are given in Table 2. It is evident from these data that the chemical compositions of the chill castings are quite similar, despite differences in the super-

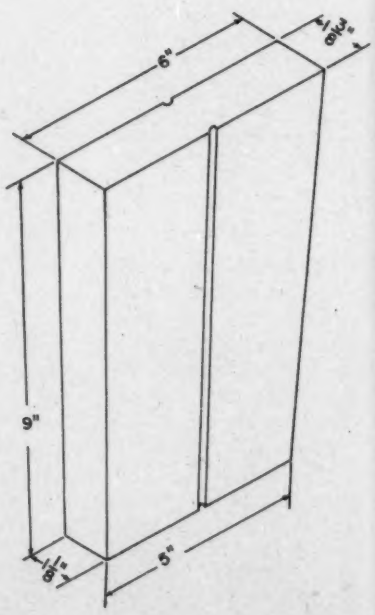



Fig. 1 — Chill test casting.

heating temperature, pouring temperature and the charge makeup of the melts from which they were produced.

TABLE 2 — CHEMICAL COMPOSITIONS OF CHILL TEST CASTINGS

| Chill Test No. | Superheating Temp., F | Pouring Temp., F | Weight, % C | Si | Mn | P | S |
|---|---|---|---|---|---|---|---|
| 1A, 1B | 2600 | 2600 | 3.29 | 0.77 | 0.27 | 0.418 | 0.108 |
| 1C, 1D | 2600 | 2450 | 3.28 | 0.76 | 0.28 | 0.406 | 0.107 |
| 2A, 2B | 2700 | 2700 | 3.31 | 0.75 | 0.37 | 0.456 | 0.148 |
| 2C, 2D | 2700 | 2450 | 3.31 | 0.74 | 0.34 | 0.434 | 0.114 |
| 3A, 3B | 2800 | 2800 | 3.31 | 0.75 | 0.37 | 0.456 | 0.143 |
| 3C, 3D | 2800 | 2450 | 3.30 | 0.74 | 0.36 | 0.402 | 0.138 |
| 4A, 4B | 2900 | 2900 | 3.35 | 0.77 | 0.35 | 0.454 | 0.132 |
| 4C, 4D | 2900 | 2450 | 3.31 | 0.75 | 0.31 | 0.450 | 0.140 |
| 5A, 5B | 2600 | 2600 | 3.27 | 0.74 | 0.35 | 0.478 | 0.092 |
| 5C, 5D | 2600 | 2450 | 3.30 | 0.74 | 0.32 | 0.462 | 0.115 |
| 6A, 6B | 2900 | 2900 | 3.35 | 0.74 | 0.34 | 0.436 | 0.079 |
| 6C, 6D | 2900 | 2450 | 3.29 | 0.75 | 0.35 | 0.416 | 0.090 |

Melts 1-4 produced from roll iron, graphite, ferrosilicon and ferromanganese; melts 5 and 6 from roll iron, pig iron, ferromanganese and ferrosilicon.

The results of vacuum fusion gas analyses, which were made on samples cast from two melts prepared principally from roll iron, are given in Table 3. The samples were taken from the white iron castings made for chemical analysis. Because the samples were not properly refrigerated before analysis, the values for hydrogen may not be representative of the hydrogen content of the melts before casting. It is interesting to note, however, that the gas contents of all the samples were about the same.

TABLE 3 — VACUUM FUSION GAS ANALYSES

| Chill Test No. | Superheating Temp., F | Pouring Temp., F | Weight, % $O_2$ | $N_2$ | $H_2$ |
|---|---|---|---|---|---|
| 1A, 1B | 2600 | 2600 | 0.0073 | 0.0040 | 0.00033 |
| 1C, 1D | 2600 | 2450 | 0.0068 | 0.0040 | 0.00027 |
| 4A, 4B | 2900 | 2900 | 0.0067 | 0.0050 | 0.00016 |
| 4C, 4D | 2900 | 2450 | 0.0072 | 0.0040 | 0.00015 |

### *Superheating Temperature, Pouring Temperature and Melting Stock Effect on Chill Formation*

The combined influence of superheating temperature and pouring temperature on chill behavior is shown in Fig. 2. Only the fractures of the *B* and *D* tests are shown, since the fractures of the matching tests were essentially the same. Reading from left to right, the melting and pouring temperature for each test was 2600 F (1427 C), 2700 F (1482 C), 2800 F (1538 C) and 2900 F (1593 C). It is evident that the depth of chill and the dispersion of the mottle increase as the superheating temperature and pouring temperature are increased from 2600 to 2900 F (1427 to 1593 C).

The effect of superheating temperature on chill capacity is shown in Fig. 3, where the chill tests were taken from melts superheated to 2600 F, 2700 F, 2800 F and 2900 F, as shown from left to right. The casting temperature in each case was 2450 F. There is an increase in chill depth and in the dispersion of the mottle as the melting temperature increases from 2600 to 2900 F. The effect of raising the temperature from 2800 to 2900 F, however, was not as pronounced as that obtained by increasing it from 2700 to 2800 F.

The influence of pouring temperature on chill behavior can be seen by comparing the chill tests shown in Figs. 2 and 3. It is clear that for a given melt, the chill depth is reduced by lowering the pouring temperature from the superheating temperature to 2450 F.

The differences in chill behavior produced by changes in the nature of the charge at low superheating temperatures are shown in Fig. 4. From left to right, in Fig. 4, the first two castings were taken from irons melted and cast at 2600 F and the second two from irons melted at 2600 F but poured at 2450 F. The first and third castings were prepared principally from roll iron and graphite, whereas the second and fourth were produced from roll iron and pig iron. It is quite evident that the chill depth in the castings is markedly different despite similar composition.

The changes in chill characteristics, produced by differences in the nature of the charge at high superheating temperatures, are shown in Fig. 5. Reading left to right, the first pair of tests was cast from metal melted and cast at 2900 F and the second pair of tests from metal melted at 2900 F and poured at 2450 F. The first and third tests were prepared primarily from roll iron and graphite, while the second and fourth were produced essentially from roll iron and pig iron. Here again, the castings produced from roll iron exhibited less chill than the ones made from roll iron and pig iron. The differences, however, are not as pronounced as those in the case of melts superheated to lower temperatures.

### *Melting and Pouring Temperature Influence on the Microstructure of the Chill Test Castings*

Samples for metallographic examination were taken from the gray and white iron portions of the chill test castings used in the investigation of superheating temperature and pouring temperature. The structure of the white iron was examined at the chilled surface of each casting; at ¼-in. from the chilled surface; and at ½-in. from the chilled surface. Specimens used to examine the gray iron portion of the chill test castings were taken 5 in. from the chilled surface.

The microstructures in the white iron locations in castings, poured at 2600 F (1427 C) and 2450 F (1343 C) from a melt superheated at 2600 F, are shown in Fig. 6. Those for the same locations in castings poured at 2900 F (1593 C) and 2450 F (1343 C) from a melt superheated to 2900 F, are presented in Fig. 7. It is evident that superheating temperature, in itself, has no noticeable effect on the structure of the chill. However, pouring temperature does have an influence.

The structure in the castings poured at the superheating temperature is in general columnar and strongly oriented toward the chilled surface, whereas the structure in those poured at 2450 F is more



**Fig. 2 — Combined influence of superheating and pouring temperature. Chill test — left to right — 1B, 2B, 3B, 4B.**



Fig. 3 — Superheating temperature effect. Chill tests — left to right — 1D, 2D, 3D, 4D.



Fig. 4 — Charge materials influence on chill behavior at a superheating temperature of 2600 F. Chill tests — left to right — 1B, 5B, 1D, 5D.

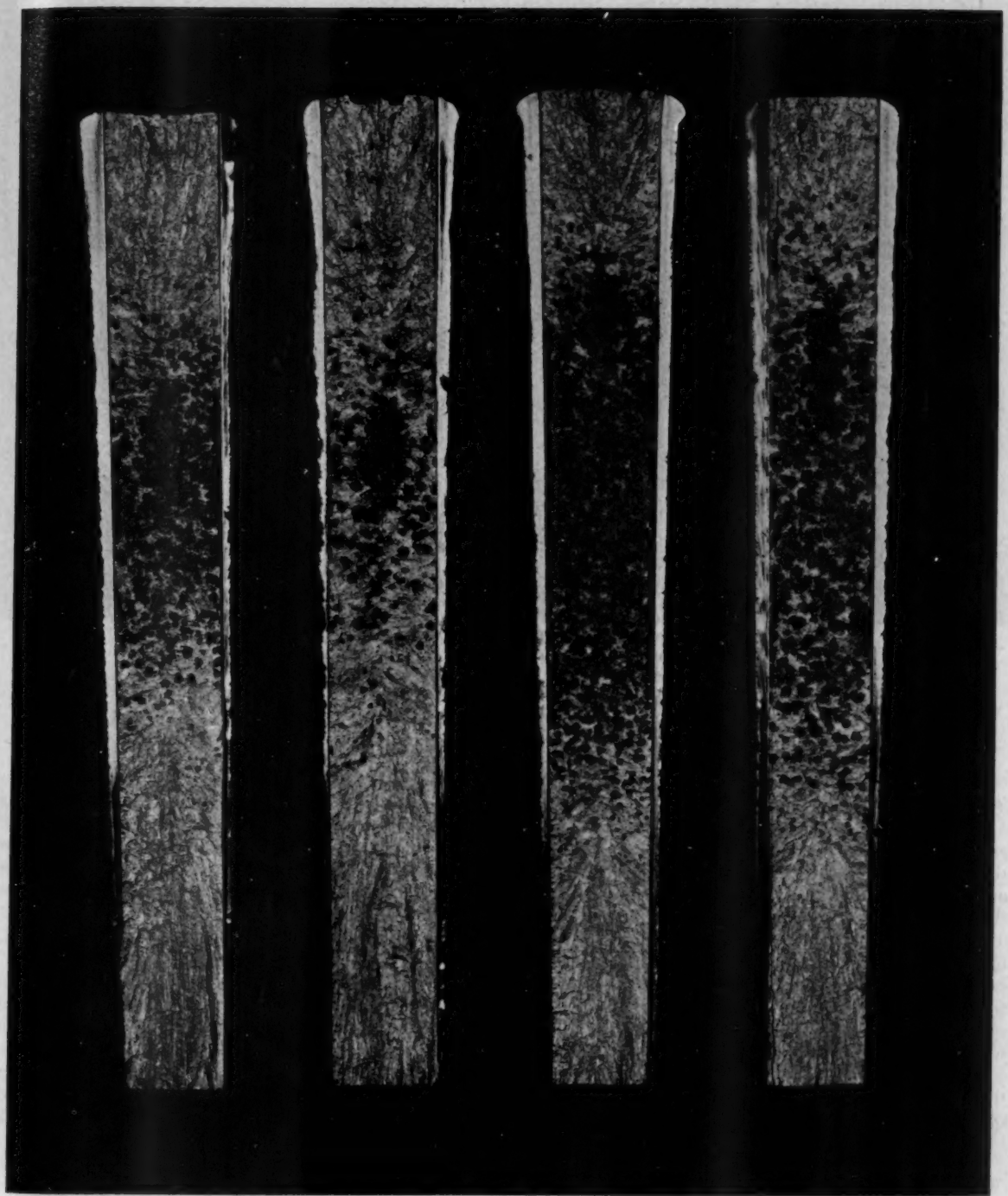


Fig. 5 — Charge materials influence at superheating temperature of 2900 F. Chill tests — left to right — 4B, 6B, 4D, 6D.

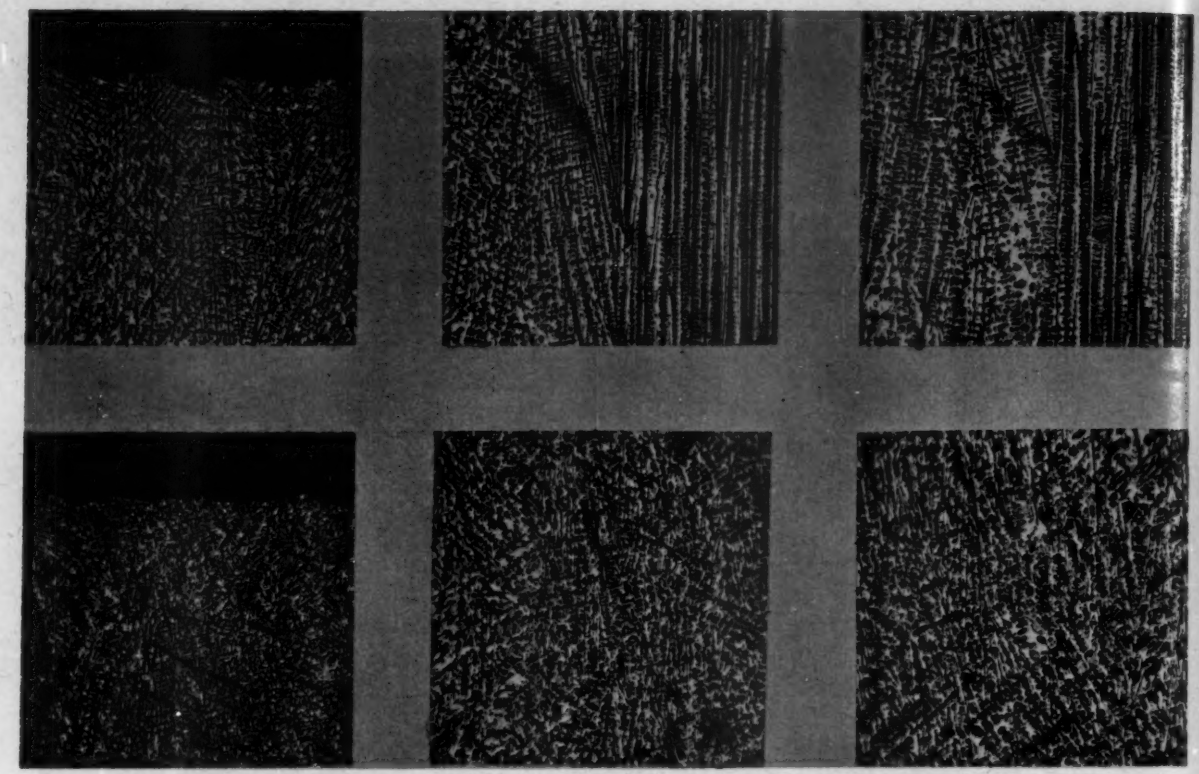


Fig. 6 — Microstructure of chill test castings at specified distances from surface of chill. *Top row* — superheating temperature — 2600 F; pouring temperature 2600 F. *Bottom row* — superheating temperature — 2600 F; pouring temperature 2450 F. Left to right — adjacent to chill, ¼-in. from chill and ½-in. from chill. 4 per cent nital etch. 25 ×.

random and does not exhibit any marked orientation toward the chill surface.

Metallographic examinations were made of the graphite flake distribution, matrix structure and cell size in the samples taken from the gray iron portion of the chill test castings. In general, it was found that neither superheating temperature nor pouring temperature had any marked effect on these features of the microstructure. Somewhat larger amounts of excess carbide were present in the castings poured from melts superheated to 2800 F (1538 C) and especially 2900 F (1593 C), but no other differences were evident. Photomicrographs illustrating the typical distribution and size of the graphite flakes, and the structure of the matrix in the gray iron section of the castings are presented in Fig. 8.

## DISCUSSION

The experiments performed in this investigation have shown that differences in superheating temperature, pouring temperature and melting stock definitely affect the chill behavior of melts with almost identical chemical composition. Furthermore, if the results of the vacuum fusion analyses are considered to be valid, the changes in chill behavior produced by superheating temperature and pouring temperature do not appear to be related to changes in total oxygen, nitrogen or even perhaps hydrogen content.

This evidence would thus seem to refute the gas content theory which links the effects of superheating to the influences of different gases on graphitization. However, it must be considered that a customary gas analysis does not differentiate between the amount of gas present in uncombined form and that present in the form of compounds (oxides, etc.); and perhaps it is premature to discredit entirely the gas content theory.

In the investigation dealing with the influences of charge materials, an interesting comparison resulted when the amount of carbon present as graphite in each charge was calculated. These determinations indicated that the melts produced from the charges with the smaller amount of graphite (those prepared from roll scrap and graphite) exhibited less chill capacity than those produced from charges with a greater amount of graphite (melts produced from roll iron and pig iron).

This conclusion is not in agreement with the results of previous investigations, which indicate that greater amounts of graphite in the melting stock should decrease the chilling tendency of the subsequent melts.[8-9] At first consideration, the latter observations would seem to substantiate the graphite nuclei theory, while those of the current investigation appear to disprove it.

Research work carried out at The Pennsylvania

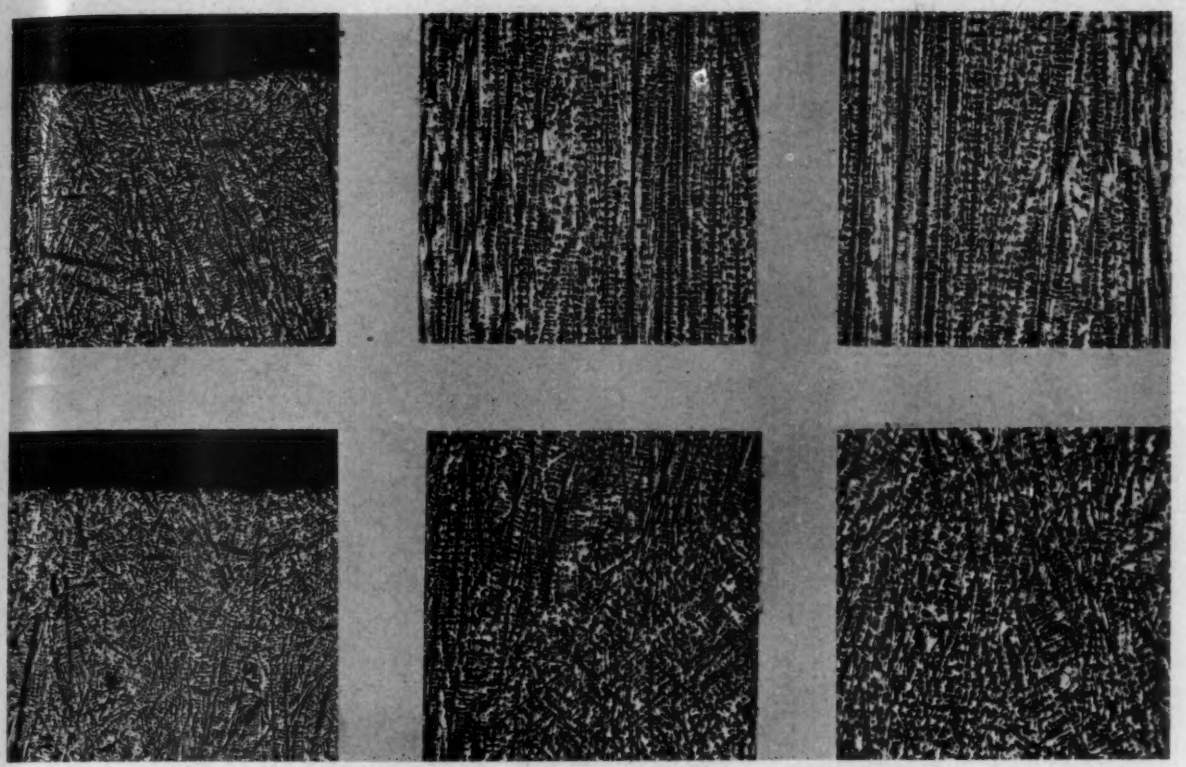


Fig. 7 — Microstructure of chill test castings at specified distances from surface of chill. *Top row* — superheating temperature — 2900 F; pouring temperature — 2900 F. *Bottom row* — superheating temperature — 2900 F; pouring temperature — 2450 F. Left to right — adjacent to chill, ¼-in. from chill and ½-in. from chill. 4 per cent nital etch. 25 ×.

State University, however, has shown that cast iron can graphitize to a considerable extent upon heating to its melting temperature.[10] It would appear necessary, therefore, to ascertain the amount of graphite present in the charge not at room temperature, but at the temperature of melting. Unfortunately, this was not done in the current experiments, and no firm conclusions regarding the validity of the graphite nuclei theory can be made.

### *Superheating Temperature Effect*

Superheating temperature was shown to have no marked effect on the character of microconstituents in the white iron structure or in the gray iron structure of the chill test castings, providing pouring temperature was held constant. It should be pointed out, however, that superheating temperature does have an effect if a specific location in a casting, and hence a specific cooling rate, is considered.

Increased superheating temperature can, for instance, cause the structure of a casting at a certain location to become white or mottled depending on the temperature to which the melt is heated. It follows then, that increased superheating temperature can influence the structure of a casting by reducing the rate of cooling required to change the eutectic solidification from gray to white.

## CONCLUSIONS

An increase in superheating temperature from 2600 to 2900 F (1427 to 1593 C) increases the carbide stability of roll-type cast iron.

The carbide stabilizing effects of superheating temperature are not a direct result of differences in customary chemical composition between melts superheated to different temperatures. Furthermore, they do not appear to be caused by differences in oxygen, nitrogen and perhaps hydrogen content.

Superheating temperature did not markedly affect the character of the microconstituents in the white iron and gray iron structures of the chill test castings. This observation should not be construed as evidence, however, that superheating has no effect on the overall structure of a casting.

The chill capacity of roll-type cast iron decreases as pouring temperature is reduced from superheating temperatures above 2600 F (1427 C).

The influence of pouring temperature on chill behavior is not the direct result of differences in customary chemical composition. This may also be true of differences in oxygen, nitrogen and hydrogen content.

Pouring temperature had a marked effect on the structure of the white iron in chill test castings produced from roll-type cast iron. The chill structure in

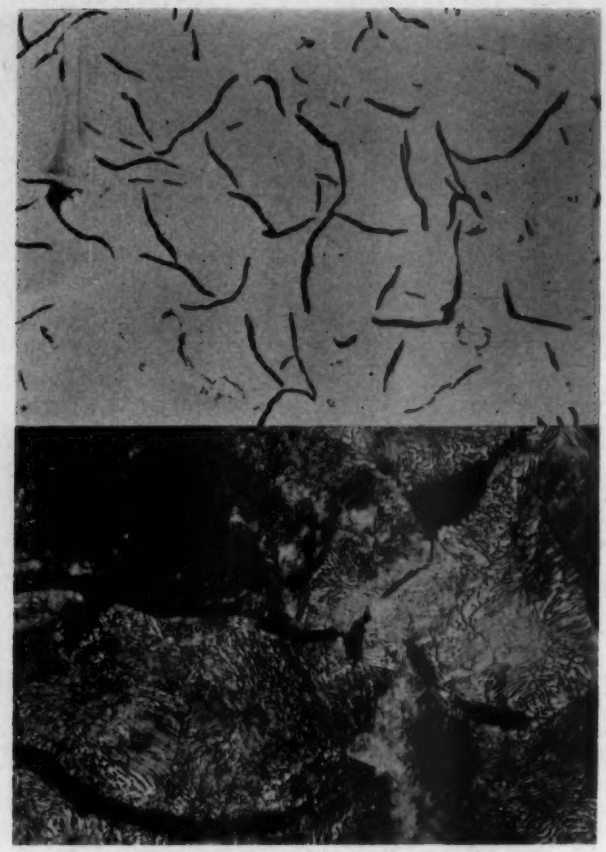


**Fig. 8 — Typical microstructures of gray iron portion of chill test castings at a location 5 in. from chilled face. *Top* — unetched. 100 ×. *Bottom* — 4 per cent nital etch. 500 ×.**

castings poured at 2450 F (1343 C) was considerably less columnar than that in castings poured from the same melt at the superheating temperature.

Charge materials noticeably affect the chill behavior of the subsequent melts. The changes become smaller, however, as the superheating temperature is increased from 2600 to 2900 F (1427 to 1593 C). Differences produced by charge materials evidently are not the result of differences in customary chemical composition.

## ACKNOWLEDGMENTS

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# SYSTEMATIC APPROACH TO SAND DESIGN AND CONTROL

## *Progress report 2— clay bonded sand dry properties*

by A. H. Zrimsek and G. J. Vingas


### ABSTRACT

**The second in a series of reports presenting basic data which describe the effect of a number of variables on foundry sands, this report deals with the dry properties of clay-sand-water combinations. The data presented compliment that given in the previous progress report. The systems covered in this report are those of western and southern bentonite and fireclay.**


### EXPERIMENT

The base sand used in the initial phases of this work is a 51.3 fineness Portage silica, a description of which is given in the table. As in the previous progress report,* data collected were obtained from sands mulled in an 18 in. laboratory muller. With this muller a batch size of 4000 grams of sand was used. To the dry sand was added the clay additions of 4.75, 7.45 or 10 per cent for the bentonites and 10 or 15 per cent for fireclay. A 15 sec dry mull preceded water additions.

To study the effect of mulling time on the development of dry properties, 2, 4 and 6 min wet mull cycles were used. The mulled sand was kept in airtight polyethylene bags while dry strength specimens were made at 1, 3, 5, 7 and 10 rams. Specimens were oven dried at 220-230 F for periods not less than 5 hr. At least five different water contents were used for each clay per cent at each mulling time.

### RESULTS

Because the volume of data collected in this series of reports are large, results are shown in graphical rather than tabular form. The graphical presentation (Figs. 1-12) also allows a clear picture of trends which develop as variables are changed.

**SCREEN ANALYSIS OF BASE SAND USED**

| U.S. Standard Sieve No. | Retained, % |
|---|---|
| 20 | 0.0 |
| 30 | 2.6 |
| 40 | 19.3 |
| 50 | 30.1 |
| 70 | 24.4 |
| 100 | 15.1 |
| 140 | 5.6 |
| 200 | 2.4 |
| Pan | 0.6 |

### DISCUSSION

As in the previous progress report, mulling time emerges as one of the major variables encountered. The effect of mulling time on physical properties in some instances overshadows the effects of clay type and per cent. The supposed low dry strength of southern bentonite versus the high dry strength of western does not appear in any of the data gathered from sands prepared in the 18 in. muller at mulling periods of 6 min or less. Dry strengths obtained with southern are in fact higher than those obtained with western in most instances.

The data so contradict the accepted precept, that western bentonite imparts higher dry strengths than southern, that checks were made on several other commercially available western and southern bentonites with similar results encountered. Western bentonites as a group followed closely the characteristics shown in Figs. 4 and 5, and southerns as a group followed the trends shown in Figs. 7 through 9. In no instance did the western bentonites impart significantly higher dry strengths than southern when mulled in the 18 in. laboratory muller.

Only when extensive mulling periods were used did the western surpass the southern. Use of the more efficient 12 in. laboratory muller allowed develop-

*A. H. Zrimsek and G. J. Vingas, "Systematic Approach to Sand Design and Control," AFS TRANSACTIONS, *vol. 68*, p. 440 (1960).

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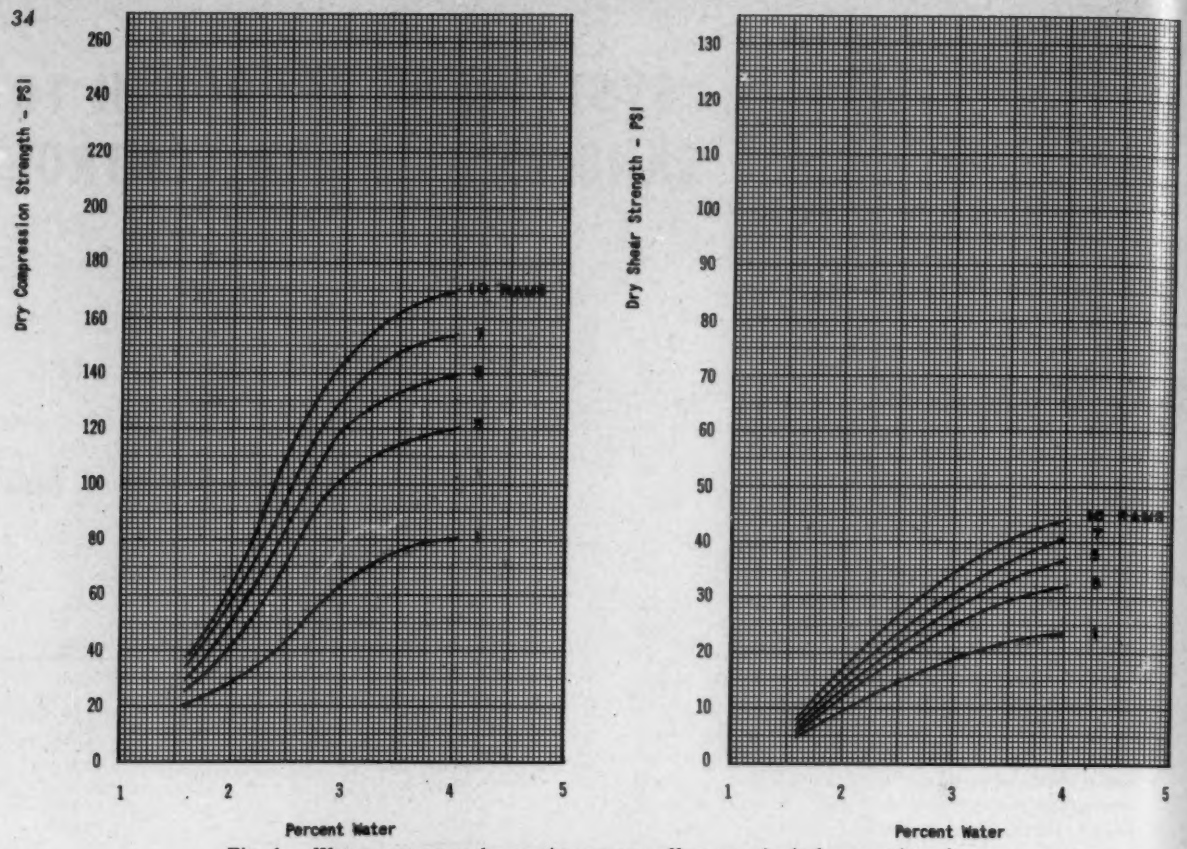



Fig. 1 — Water content and ramming energy effect on physical properties of 4.75 per cent western bentonite bonded Portage silica sand. Mulled 6 min.

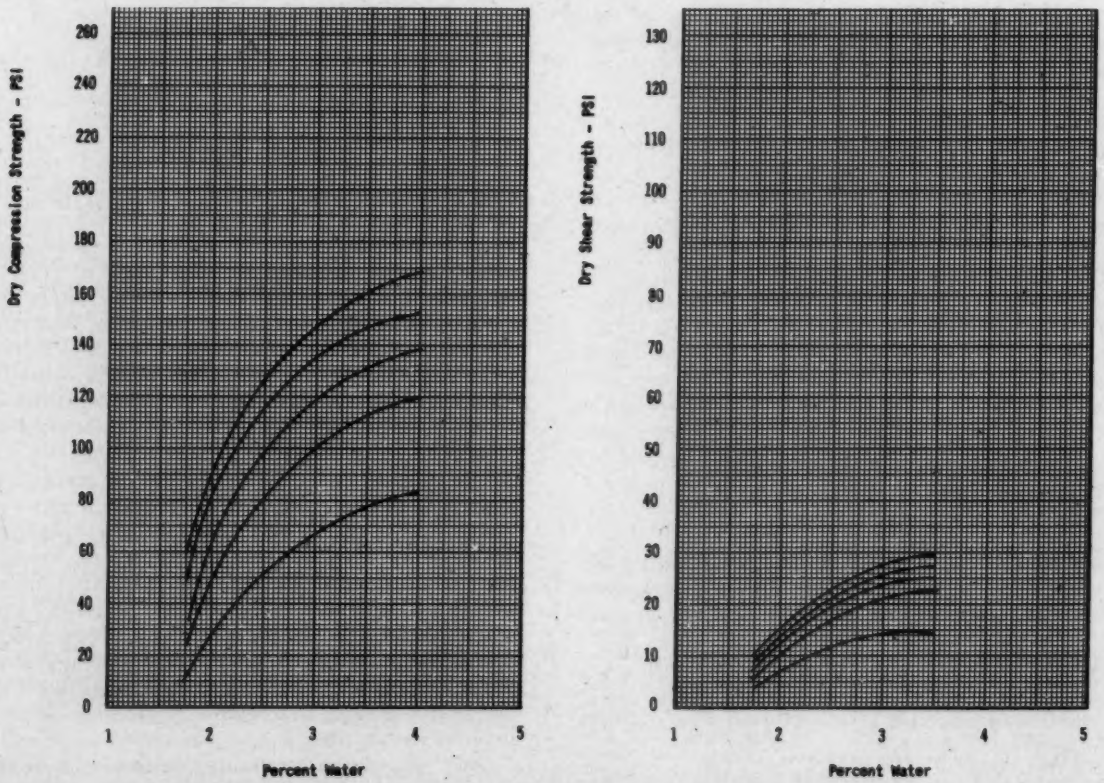



Fig. 2 — Water content and ramming energy effect on physical properties of 4.75 per cent southern bentonite bonded Portage silica sand. Mulled 6 min.

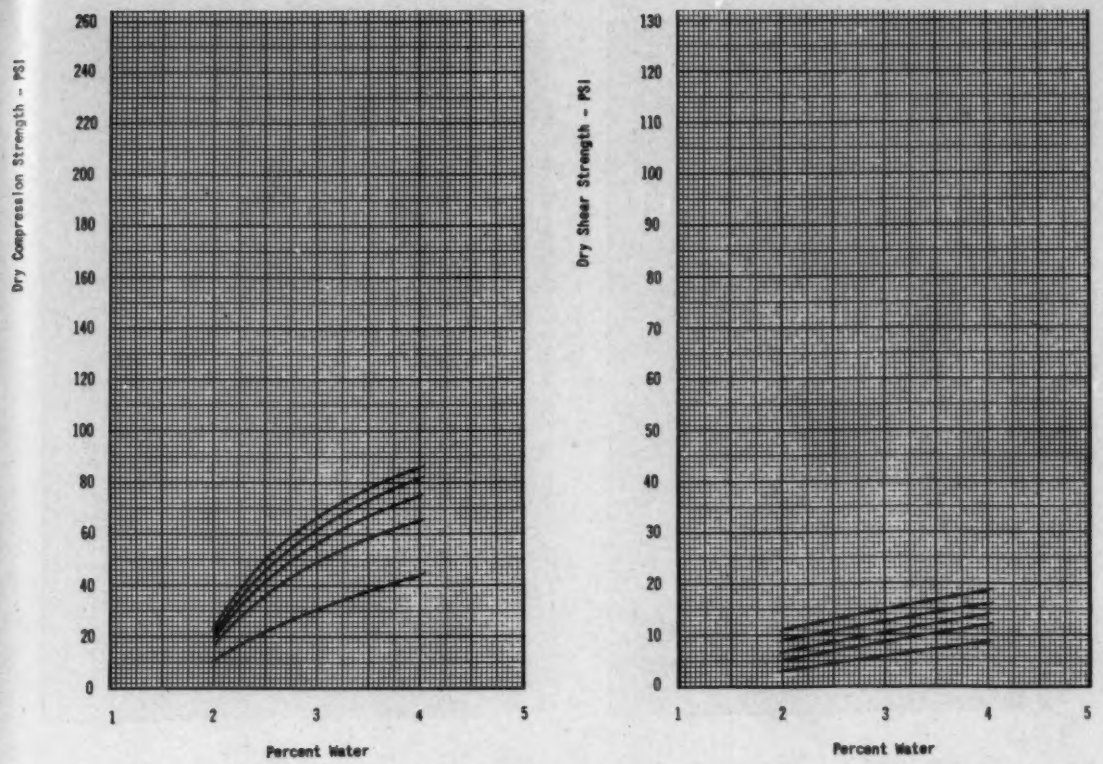



Fig. 3 — Water content and ramming energy effect on physical properties of 10 per cent fireclay bonded Portage silica sand. Mulled 6 min.

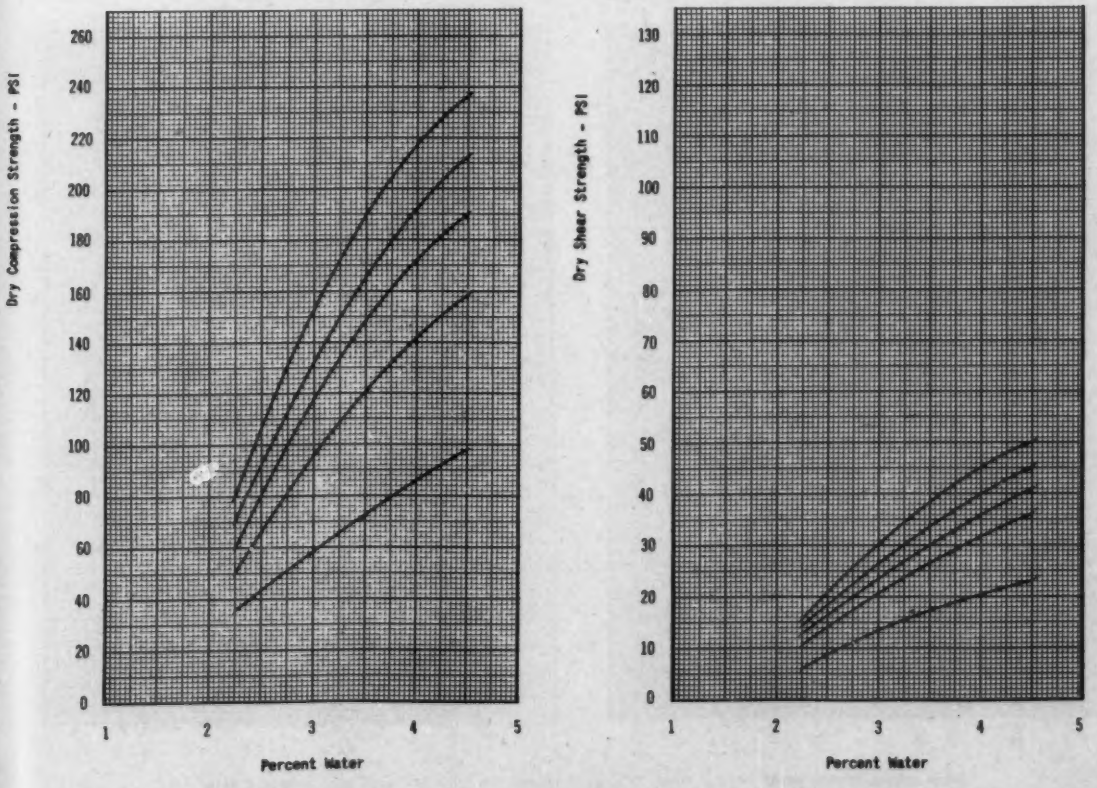



Fig. 4 — Water content and ramming energy effect on physical properties of 7.45 per cent western bentonite bonded Portage silica sand. Mulled 6 min.

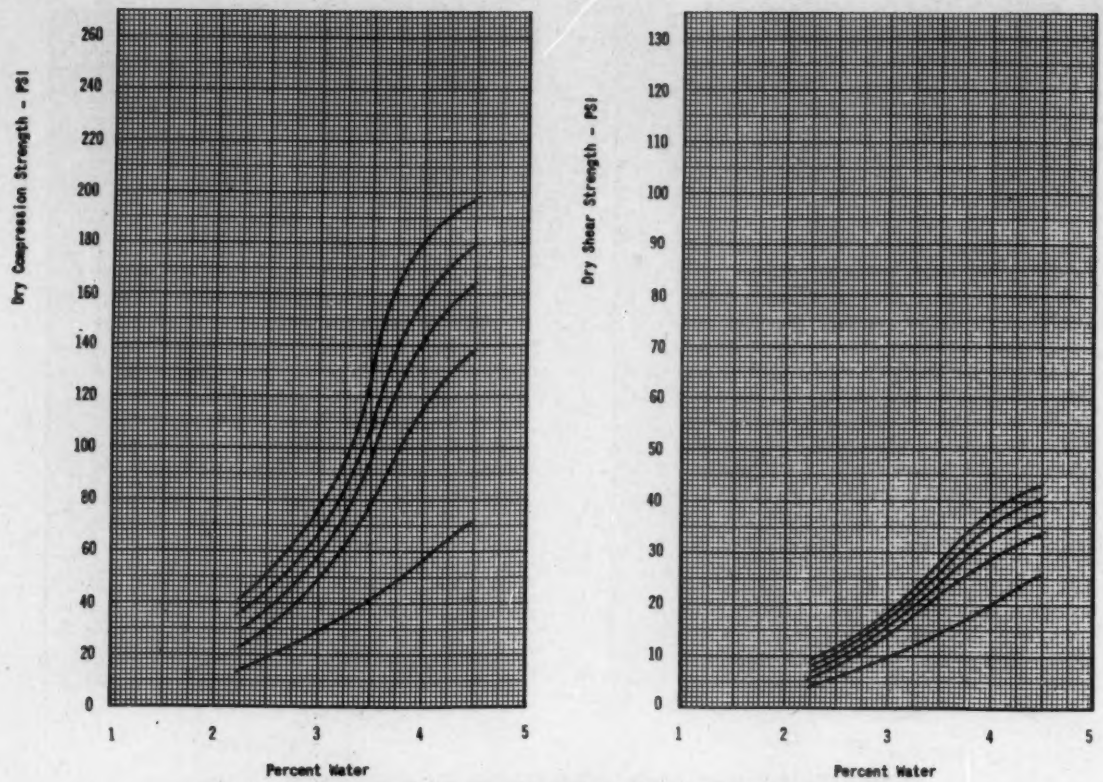



Fig. 5 — Water content and ramming energy effect on physical properties of 7.45 per cent western bentonite bonded Portage silica sand. Mulled 4 min.

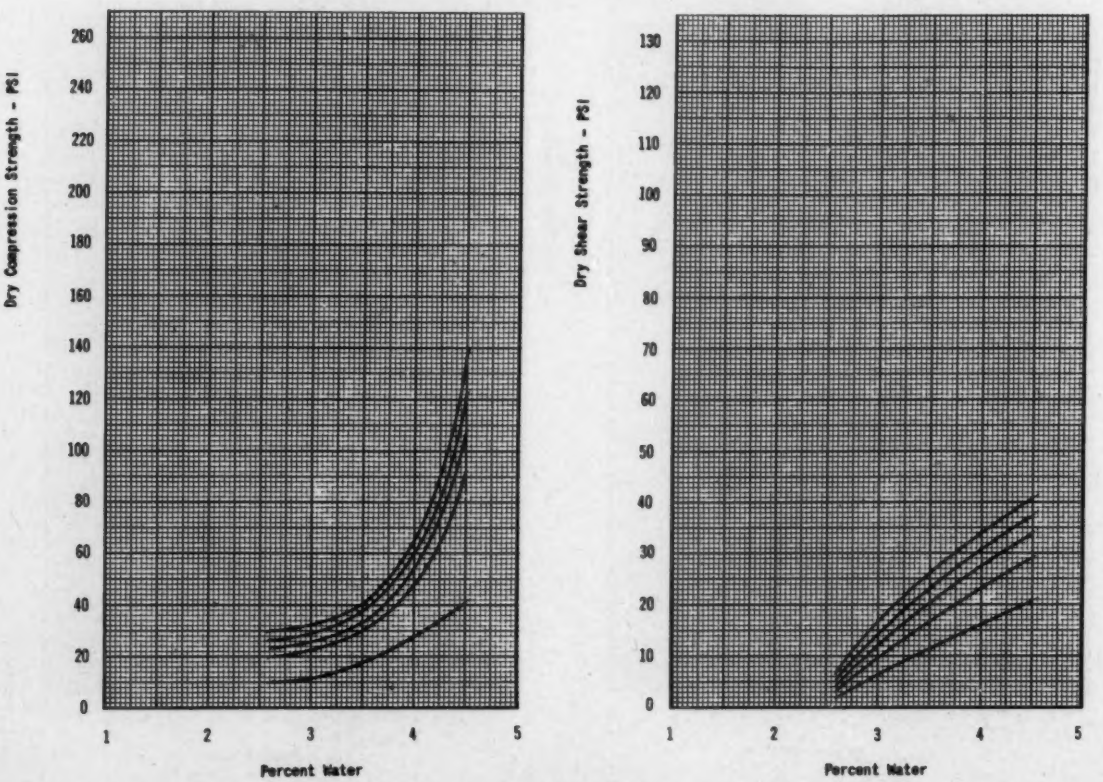



Fig. 6 — Water content and ramming energy effect on physical properties of 7.45 per cent western bentonite bonded Portage silica sand. Mulled 2 min.

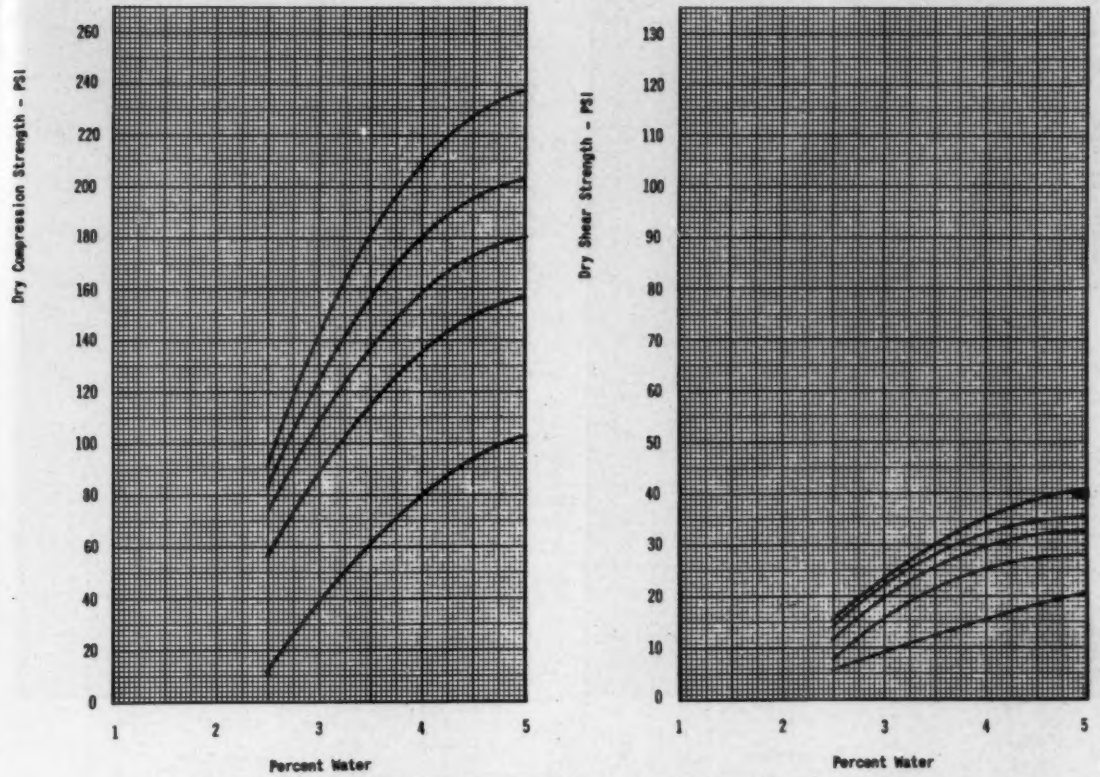



Fig. 7 — Water content and ramming energy effect on physical properties of 7.45 per cent southern bentonite bonded Portage silica sand. Mulled 6 min.

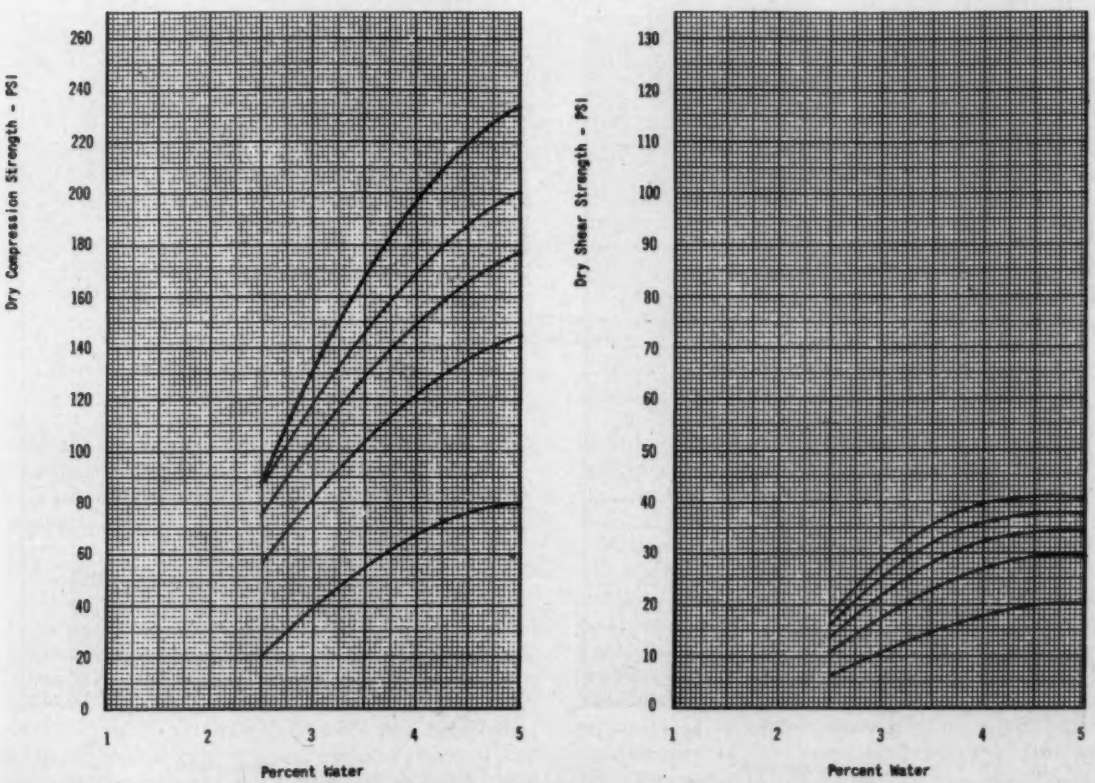



Fig. 8 — Water content and ramming energy effect on physical properties of 7.45 per cent southern bentonite bonded Portage silica sand. Mulled 4 min.

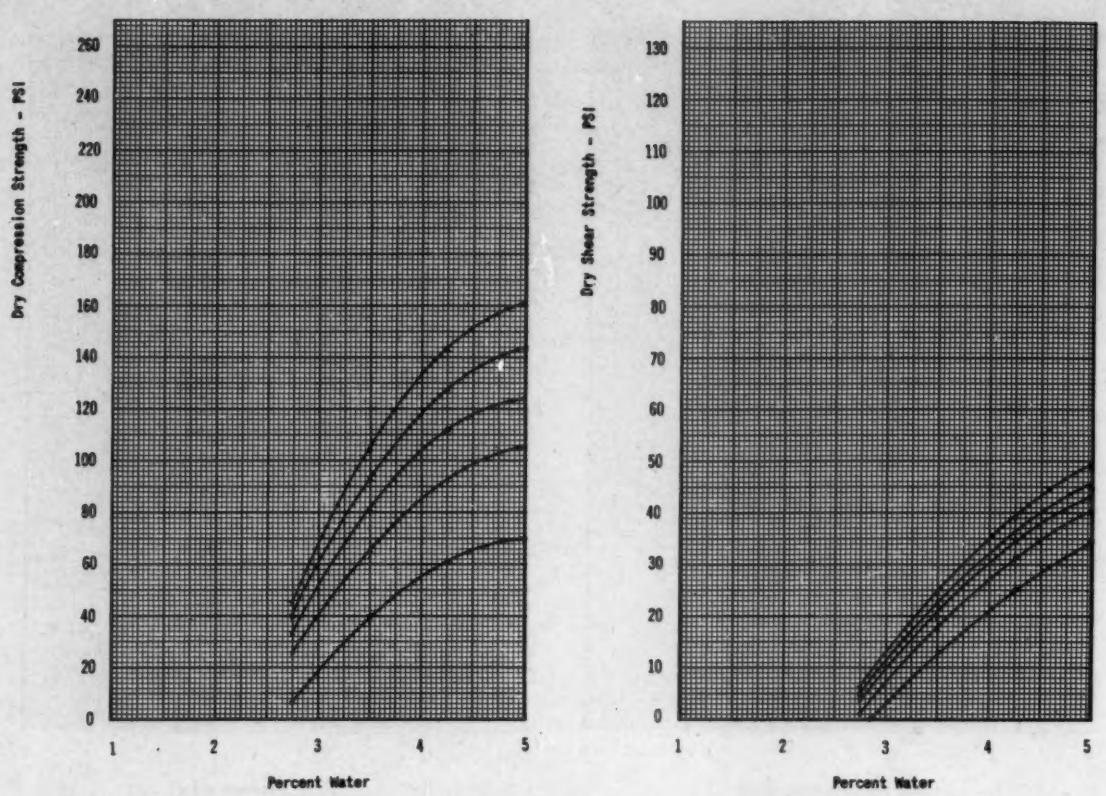



Fig. 9 — Water content and ramming energy effect on physical properties of 7.45 per cent southern bentonite bonded Portage silica sand. Mulled 2 min.

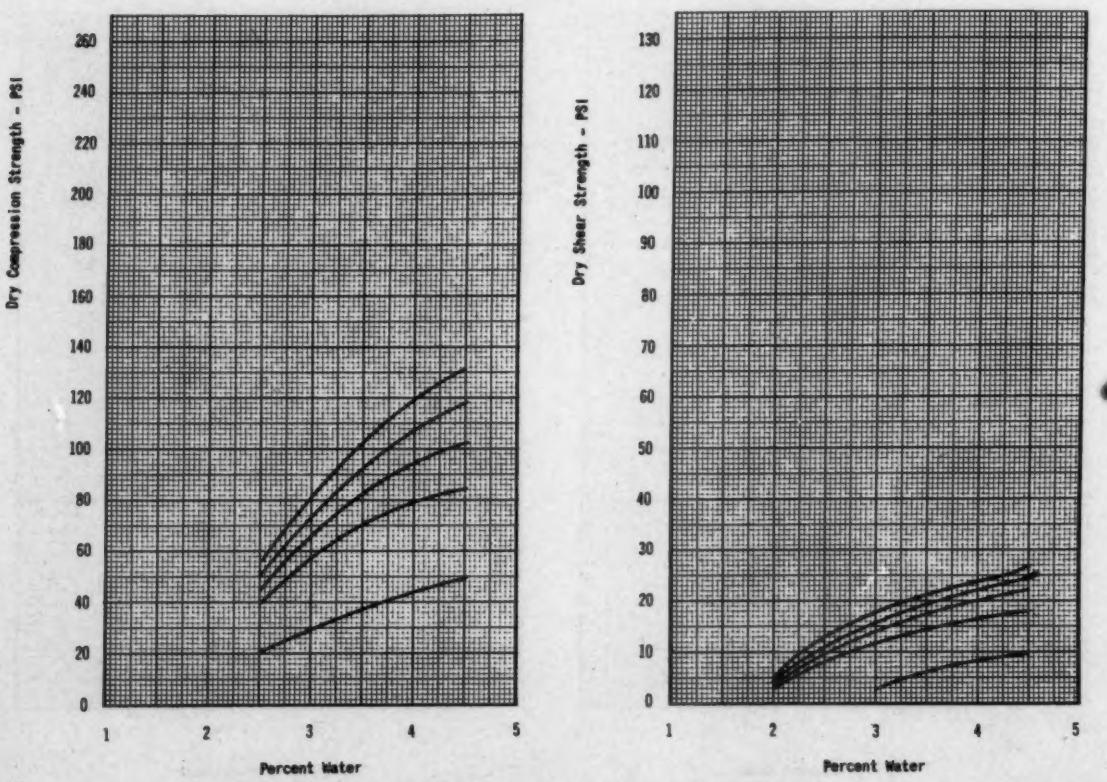



Fig. 10 — Water content and ramming energy effect on physical properties of 15 per cent fireclay bonded Portage silica sand. Mulled 6 min.

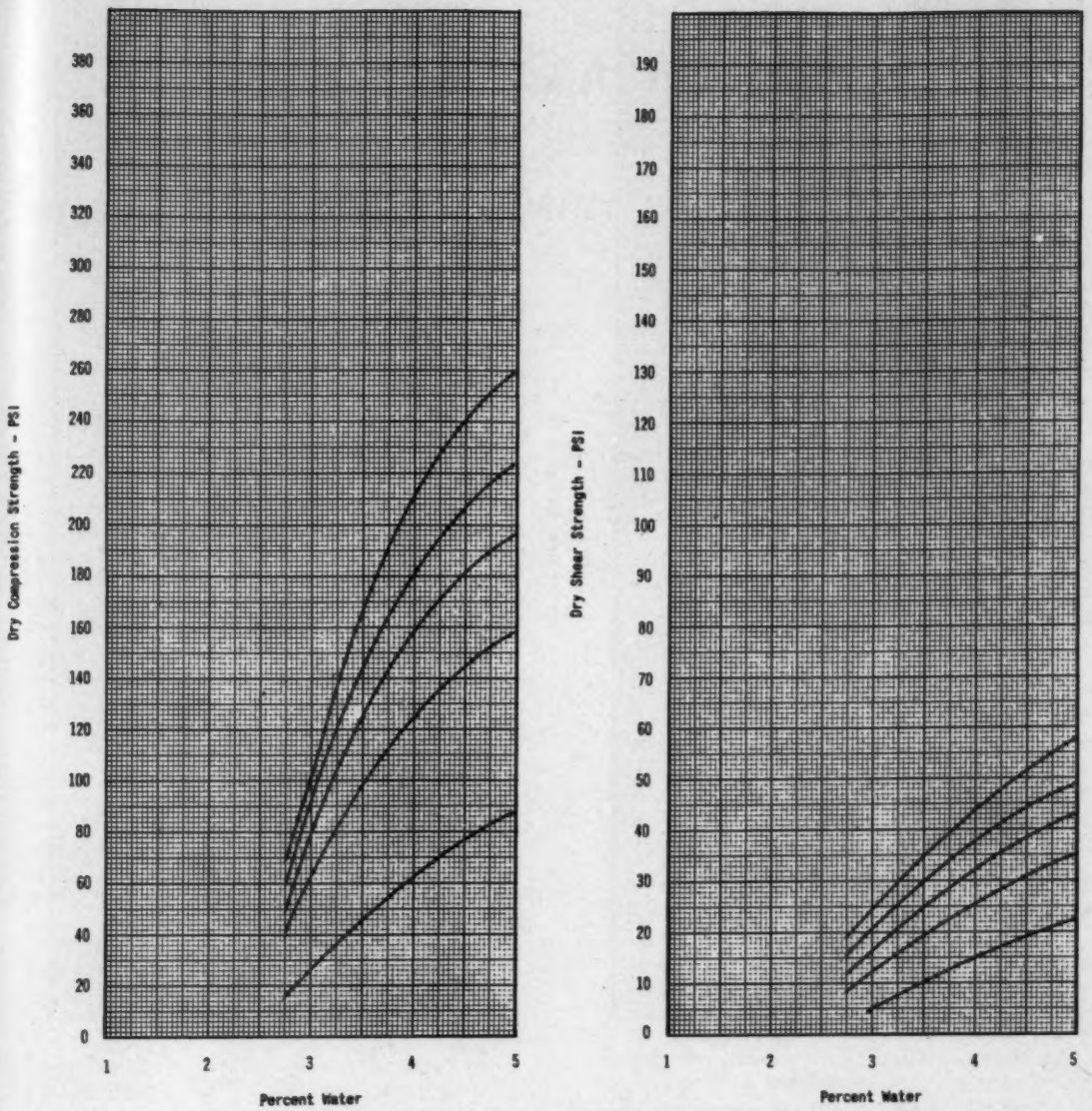



**Fig. 11 — Water content and ramming energy effect on physical properties of 10 per cent western bentonite bonded Portage silica sand. Mulled 6 min.**

ment of the maximum dry properties of western in only 6 min. As shown in comparing Figs. 11 and 12, western has the capability of imparting higher dry properties than southern, provided extensive or efficient mulling conditions exist.

The importance of mulling conditions as a variable is of much greater importance when using western bentonite than when using southern. This is true at least when new sand mixtures are being considered. Whether this holds true for rework sands remains to be seen, and will be covered in a later report of this series. The present data dealing with new mixtures would indicate that approximately two to three times as much mulling energy is required to develop the maximum properties of western than is required for southern. The mulling effect is in fact such an important factor in foundry sand design and control that an entire progress report of this series will be devoted to this important variable.

Dry properties of simple clay-sand-water systems are sensitive to small changes in water content. Regardless of clay type or per cent, increase in water is accompanied by an increase in dry strength. As pointed out earlier, the higher dry compression strength usually attributed to western over southern bentonite was not readily apparent. The dry shear strengths of western bentonite sands were, however, significantly higher than those bonded with southern.

Both dry compression and dry shear strengths ob-

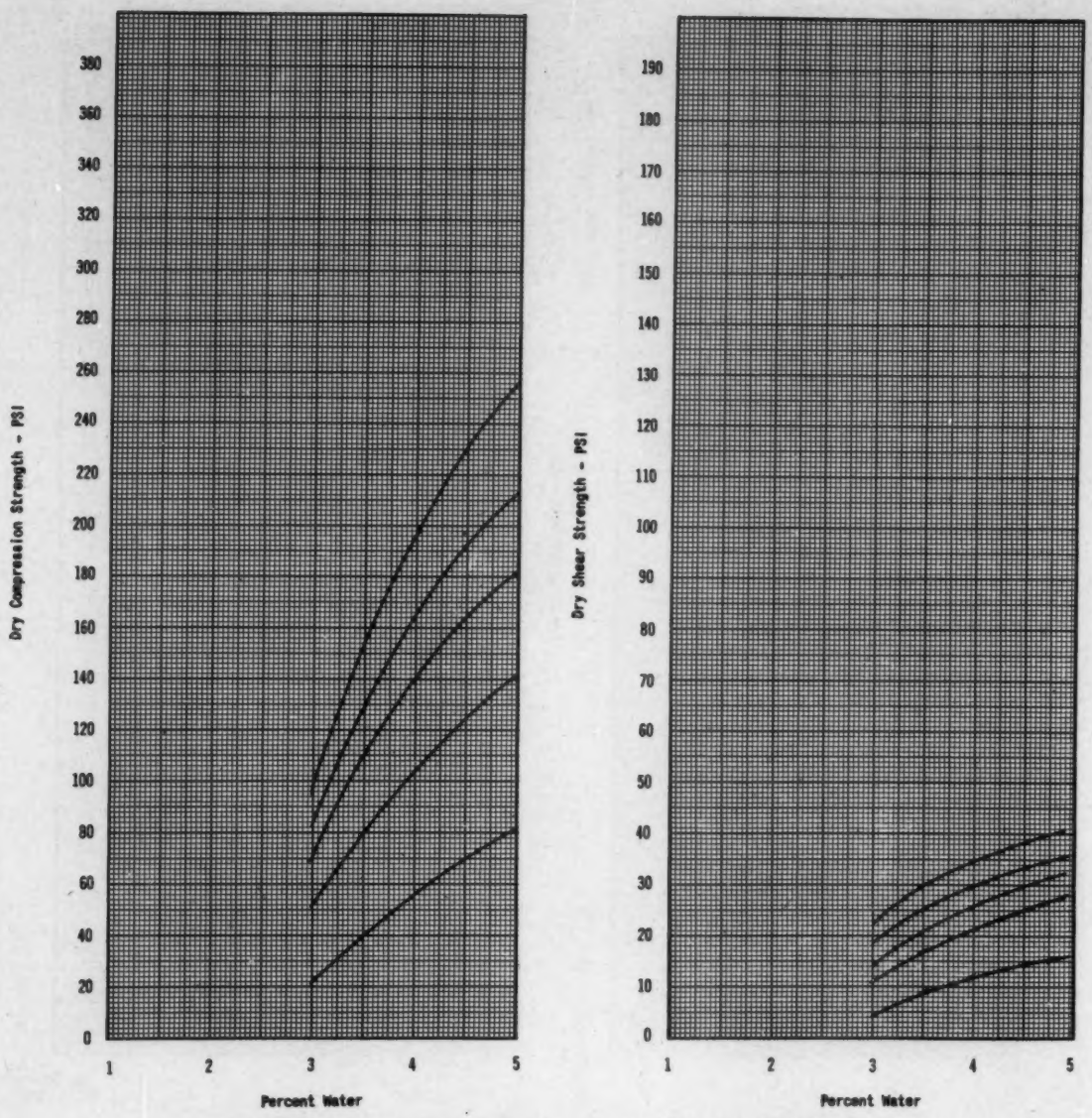



**Fig. 12 — Water content and ramming energy effect on physical properties of 10 per cent southern bentonite bonded Portage silica sand. Mulled 6 min.**

tained with fireclay bonded sands were much lower than obtained with western or southern bentonite.

## CONCLUSIONS

Intelligent discussion of the physical properties of clay bonded sands is impossible without a thorough understanding of the function of mulling as a variable. It is apparent from the data thus far presented in this report and in the previous progress report that almost any combination of physical properties can be had through a variation in clay and water contents and mulling efficiency. The tremendous variation in physical properties possible through variations in mulling alone helps explain why sand formulations which work wonders in one foundry often cause chaos when tried in another.

There can be no doubt that the properties of clay bonded sands are affected extensively by mulling. Under the mulling conditions considered in this paper, there was little apparent difference in the dry properties of western and southern bentonite. Under no circumstances did the dry properties of fireclay approach those of either southern or western bentonite.

## ACKNOWLEDGMENT

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# MALLEABLE IRON GRAPHITIZATION KINETICS

## *Some trace elements influence*

by G. Sandoz, B. F. Brown and W. A. Pennington


### ABSTRACT

**Trace amounts of the elements bismuth, antimony, tin, lead, cadmium and zinc were studied for their effects on the kinetics of each of the three graphitization reactions critical to the malleabilization of irons. The intrinsic affects of the elements were revealed with the use of formulas to correct for the nodule number differences between irons (differences in the number of graphite nodules per unit volume of iron). The effects of the trace elements were determined in irons with both low and high Mn-S ratios in order to detect possible interactions with sulfur.**

**The results show that some of the trace elements react with sulfur in irons with low Mn-S ratios and others do not. These affects are explained on the basis of thermodynamic estimates of the relative potentials of the trace elements to combine with sulfur in competition with iron. The trace elements do not form sulfides in irons with high Mn-S ratios, and in such irons the trace elements are all mildly retarding to graphitization rates.**


### INTRODUCTION

The rates of the graphitization processes that must take place during the malleabilization of white irons are sensitive to the composition of the iron. Accordingly, the concentrations of the major constituents, C, Si, Mn, S and P are closely controlled. However, a number of elements may be acquired adventitiously in small or trace amounts over which close control is impractical, and some of these are suspected of severely retarding graphitization rates. For example, the elements present in white metal, bushings, liners, etc., which may constitute a fraction of the scrap used in melting, are often held accountable for annealing difficulties encountered with irons of normal base composition.

Nevertheless, the data reported in the literature in support of such claims are meager, mostly qualitative and in some cases contradictory. Accordingly, an investigation was conducted with the aim of determining quantitatively the effects of controlled small amounts of elements Bi, Sb, Sn, Pb, Cd and Zn on the graphitization processes in irons—irons synthetically prepared, but close enough to commercial malleable compositions that the influences of impurity elements should be similar.

### GENERAL BACKGROUND

An evaluation of the effects of alloying elements on the graphitization of malleable irons requires that the effects be identified with respect to each of the three individual graphitization reactions which take place and are necessary to malleabilization. These reactions may be summarized as follows:

| | | | |
|---|---|---|---|
| First Stage: | Cementite | → | Austenite + Graphite |
| Second Stage: | Austenite | → | Ferrite + Graphite (direct) |
| | Pearlite | → | Ferrite + Graphite (indirect) |

As indicated, there are two main divisions into which the reactions fall, first and second stage graphitization. First stage graphitization involves one reaction—the conversion of eutectic cementite into austenite and graphite. Second stage graphitization involves two reactions, either or both of which may convert the material entirely into ferrite and graphite.

As an iron is cooled slowly through the critical temperature range and into subcritical regions, the austenite begins to transform directly into ferrite and graphite. However, before this reaction is completed, a part of the austenite may transform into an intermediate pearlite. This pearlite may itself decompose into ferrite and graphite at subcritical temperatures.

Since alloying elements may influence the rates differently, and thereby the relative participation of these reactions, it is clearly necessary to relate data on the effects of alloying elements to the specific reaction involved. In this paper, distinction between the two second-stage reactions will be made by the terms "direct" and "indirect."

Another important requirement to any evaluation of the effects of alloying elements on the graphitiza-

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tion kinetics of malleable irons is that the effects be identified with respect to the nucleation and growth features of each reaction. It is desirable to know whether an alloying element influences the rate of a particular reaction primarily by increasing or decreasing the number of graphite nodules per unit volume of iron developed during graphitization, or whether an intrinsic change in the graphitization processes such as relative thermodynamic stability of the carbide phase is involved.

A practical benefit of such information would be the possibility of predicting whether heat treatment—prequenching, for example—would be of help in malleablizing a contaminated lot of castings. Only castings contaminated with elements that act primarily through nodule number reduction would be expected to benefit from such treatment.

Finally, the effects of alloying elements should be identified as either direct influences on the processes limiting graphitization rates or as secondary effects resulting from chemical combination with other elements in the iron, particularly sulfur.

In searching the literature describing the effects of the trace elements, an attempt was made to evaluate the data in terms of the above considerations.

## LITERATURE REVIEW

Schwartz and Guiler[1] examined the effects of ladle additions of ½ to 1 per cent (amount added) of a large variety of alloying elements to malleable irons of "normal composition." All irons were put through the complete regular malleable annealing cycle, and subsequently fracture appearance and microstructure were examined to detect any retarding elements.

They found no overall retarding effects from the elements Pb, Bi, Zn and Cb. However, Sn and Sb were found to retard overall graphitization. Sn had about the same effect as Cr, and over 0.07 per cent was considered enough to make the iron completely unmalleablizable. There is no way of determining from this work, of course, the retarding influence on the individual graphitization reactions.

Palmer[2] investigated the graphitization behavior of three blackheart malleable irons, one containing 0.06 per cent Sb, the second 0.025 per cent Bi and the third 0.016 per cent Sn. Graphitization was measured by microscopic examination and by combined and graphitic carbon determinations. However, interpretation of the results is difficult because Palmer's irons were severely decarburized during first stage graphitization. In the second stage, samples of the irons were cooled at a rate of 3 C/hr and pulled out and quenched at various temperatures. Since the graphitization of pearlite is relatively slow, the results may be considered to apply primarily to the direct second stage reaction.

All of the three elements studied by Palmer, particularly Bi, retarded first stage graphitization to some extent. Much of the observed retarding action, however, was confined to the final traces of cementite. The bulk of the cementite graphitized nearly as readily as in the base iron.

In the second stage, 0.06 per cent Sb produced severe retarding of graphitization. A cooling rate of 3 C/hr almost completely suppressed ferrite formation. The presence of 0.025 per cent Bi, or 0.16 per cent Sn also retarded second stage graphitization but to a lesser extent, for substantial amounts of bullseye ferrite were present in each case (the unalloyed base iron was completely ferritic). At least part of the retarding effects in the irons containing Sn and Sb can be attributed to decreased nodule number, because this effect definitely is exhibited in the microstructures.

### *Sn, Pb and Sb Influence*

Shnay, Wilson and Rehder[3] have reported on the influence of trace amounts of Sn, Pb and Sb on first and direct second stage reaction rates. Time to complete first stage graphitization was determined microscopically, and second stage effects were followed by measuring the amounts of ferrite and pearlite in the microstructure resulting from various linear cooling rates. The effects of the three elements on the graphitization rate of pearlite were not determined. No attempt was made to measure nodule size and number.

These investigators found no effects on first stage graphitization rates caused by the presence of up to 0.011 per cent Sb, 0.029 per cent Pb or 0.046 per cent Sn. The element Pb also had no effect on the direct second stage reaction. The elements Sn and Sb, however, appeared to accelerate the direct second stage reaction in amounts up to 0.014 per cent and 0.003 per cent, respectively. Greater amounts of these elements produced a declining rate. The decline of graphitization rate as Sn was increased to 0.046 per cent. It was, however, mild, and certainly not indicative that 0.07 per cent Sn should make the iron completely unmalleablizable.[1,4]

In other work, Rehder[5] has ascribed difficulties in producing commercial blackheart malleable iron to the presence of 0.09 per cent Sn. In this case, however, the difficulties appear to be caused by retarded graphitization of pearlite in the second stage as much as with retarded direct second stage graphitization, and not at all with first stage graphitization.

Several investigations of the influence of Sn on gray and ductile irons are pertinent to the effect of Sn on the second stage graphitization of malleable irons. David, Krause and Lownie[6] have reported that the addition of up to 0.10 per cent Sn is beneficial in gray iron because of the suppression of ferrite formation.

They also have reported that Sn suppresses bullseye ferrite in ductile irons. This is supported by DeSy and Foulon,[7] who have reported that as little as 0.03 per cent Sn can prevent all ferrite formation in ductile irons normally 70 to 90 per cent ferritic. Gilbert and Palmer[8] have prevented ferrite formation in a normally 50 per cent ferritic ductile iron by adding 0.059 per cent Sn. It seems well accepted, therefore, that the presence of Sn in these materials retards the direct graphitization of austenite in the second stage.

## LITERATURE EVALUATION

The available data in the literature indicate that the elements Pb, Cd and Zn are without effects on the rates of any of the three graphitization reactions, although only the data on Pb[3] may be considered quantitative. The best data available indicate that the

TABLE 1— CHEMICAL COMPOSITIONS OF TRACE ELEMENT SERIES (WT. PER CENT)

| Heat No. | C | Si | Mn | S | P | Ni | Bi | Sb | Sn | Pb | Cd | Zn |
|---|---|---|---|---|---|---|---|---|---|---|---|---|
| H—100—A* | 1.90 | 1.10 | 0.04 | 0.03 | 0.01 | — | — | — | — | — | — | — |
| H—100—J | 1.90 | 1.10 | 0.04 | 0.03 | 0.01 | — | 0.056 | — | — | — | — | — |
| H—100—I | 1.90 | 1.10 | 0.04 | 0.03 | 0.01 | — | 0.090 | — | — | — | — | — |
| H—116—A* | 2.15 | 0.95 | 0.02 | 0.02 | 0.01 | — | — | — | — | — | — | — |
| H—116—D | 2.15 | 0.95 | 0.02 | 0.02 | 0.01 | — | — | — | — | 0.003 | — | — |
| H—116—C | 2.15 | 0.95 | 0.02 | 0.02 | 0.01 | — | — | — | — | 0.021 | — | — |
| H—116—B | 2.15 | 0.95 | 0.02 | 0.02 | 0.01 | — | — | — | — | 0.038 | — | — |
| H—116—G | 2.15 | 0.95 | 0.02 | 0.02 | 0.01 | — | — | — | 0.021 | — | — | — |
| H—116—F | 2.15 | 0.95 | 0.02 | 0.02 | 0.01 | — | — | — | 0.067 | — | — | — |
| H—116—E | 2.15 | 0.95 | 0.02 | 0.02 | 0.01 | — | — | — | 0.081 | — | — | — |
| H—116—J | 2.15 | 0.95 | 0.02 | 0.02 | 0.01 | — | — | 0.010 | — | — | — | — |
| H—116—I | 2.15 | 0.95 | 0.02 | 0.02 | 0.01 | — | — | 0.037 | — | — | — | — |
| H—116—H | 2.15 | 0.95 | 0.02 | 0.02 | 0.01 | — | — | 0.055 | — | — | — | — |
| H—116—L | 2.15 | 0.95 | 0.02 | 0.02 | 0.01 | — | — | — | — | — | — | 0.017 |
| H—120—A* | 2.10 | 0.95 | 0.04 | 0.03 | 0.01 | — | — | — | — | — | — | — |
| H—120—C | 2.10 | 0.95 | 0.04 | 0.03 | 0.01 | — | — | — | — | — | — | 0.003 |
| H—120—B | 2.10 | 0.95 | 0.04 | 0.03 | 0.01 | — | — | — | — | — | — | 0.022 |
| H—128—A* | 2.20 | 0.91 | 0.04 | 0.03 | 0.01 | — | — | — | — | — | — | — |
| H—128—F | 2.20 | 0.91 | 0.04 | 0.03 | 0.01 | — | — | — | — | — | 0.002 | — |
| H—128—E | 2.20 | 0.91 | 0.04 | 0.03 | 0.01 | — | — | — | — | — | 0.003 | — |
| L—56—A* | 2.20 | 0.80 | 0.14 | 0.02 | 0.01 | — | — | — | — | — | — | — |
| L—56—D | 2.20 | 0.80 | 0.14 | 0.02 | 0.01 | — | — | — | — | 0.023 | — | — |
| L—56—E | 2.20 | 0.80 | 0.14 | 0.02 | 0.01 | — | — | — | 0.045 | — | — | — |
| L—56—F | 2.20 | 0.80 | 0.14 | 0.02 | 0.01 | — | — | 0.083 | — | — | — | — |
| L—56—G | 2.20 | 0.80 | 0.14 | 0.02 | 0.01 | — | 0.083 | — | — | — | — | — |
| L—56—H | 2.20 | 0.80 | 0.14 | 0.02 | 0.01 | — | — | — | — | — | 0.002 | — |

* — Base Iron

elements Bi, Sb and Sn retard both the first stage and direct second stage graphitization reactions, except possibly when present in extremely small amounts. Some data indicate that Sn also retards the indirect second stage graphitization of pearlite. No pertinent data were found that describe the effects of Bi and Sb on the graphitization of pearlite.

In general, the investigations described in the literature have been confined to irons containing substantially greater amounts of Mn than of S. Possible interactions between trace elements and S have therefore been little considered. Moreover, there have been few attempts to measure and correlate the effects of trace elements with changes in graphite nodule number, and no attempt has been made to separate such effects from intrinsic effects.

It is evident that only qualitative estimates of the effects of most of the trace elements are available in the literature. More detailed information is needed, particularly if it is hoped to obtain clues as to the mechanisms involved. To summarize, the information which is required should provide:

1) Data on the intrinsic effects of the trace elements, i.e., those effects not associated with change of nodule number.
2) Data showing the effects of the trace elements on the process (course) of each graphitization reaction. For example, entire sigmoid first stage graphitization curves are needed, not just the times required to graphitize completely.
3) Data that permit clear distinction between the effects on the individual graphitization reactions.
4) Data on irons of both low and high Mn-S ratio in order to permit the detection of possible interactions between the trace elements and sulfur.

The experiments described in this paper were undertaken with the aim of obtaining this information.

## MATERIALS PREPARATION

All of the irons were prepared from armco iron, 50 per cent ferro-silicon, Mexican graphite and, where required, 80 per cent Mn ferro-manganese. All trace-element additions were made with chemically pure stock stirred into the ladle during pouring. Melting was done in a 300 lb basic induction furnace fitted with a cover. A helium atmosphere was maintained over the charge except during pouring. All melts were poured at 2750 F (1510 C) after holding in the furnace for 30 min. All series were cast as split heats in order to minimize the effects of melting variables.

All irons were cast into open finger molds fed by a large runner. The molds were made of synthetic steel facing sand. Fingers (3/8 x 1 x 5 in.) of these castings were broken off and sliced into 1/4-in. thick specimens for graphitization studies. Preliminary studies showed graphitization rates to be uniform throughout the fingers except for regions near the runner and near the extreme ends; hence no specimens were taken from areas less than 1 1/4-in. from each end of individual fingers. Table 1 gives the analyses of the heats prepared.*

### *Procedures for Measuring Graphitization Rates*

First stage graphitization was followed by measuring quantitatively the amounts of graphite formed in specimens heated to 1650 F (899 C) for various times. The specimens were heated in a tube furnace in an

*Analyses of the trace elements were done by Crippen and Erlich Laboratories Inc., Baltimore, Md.

atmosphere of helium which had been passed through drierite and hot copper turnings. Time was measured after the specimens first reached 1650 F (899 C) which took about 15 min. All specimens were water quenched at the end of the selected graphitization period.

The amount of graphite (and therefrom the per cent graphitization) of the specimens was measured by use of the point counting method of quantitative metallography. A series of experiments indicated that this method was more reproducible than attempting to measure a quantity of cementite, because of the presence of many fine spheroidal carbides throughout the austenite of most irons, particularly at the beginning of graphitization. Point counting was performed over an area of 0.3 sq mm at 500 ×, and a minimum of 1000 grid intersections were examined for each measurement. Independent checks with different samples established that measurements were reproducible within ±5 per cent graphitization.

### *Second Stage Graphitization*

Direct second stage graphitization was investigated by measuring quantitatively the amounts of pearlite formed in specimens as a function of linear cooling rate from 1650 F (899 C) (after first stage graphitization was complete). This is a legitimate procedure because any austenite that is not eliminated by the direct second stage reaction transforms to pearlite; hence the proportion of pearlite remaining in the matrix after the linear cool through the critical temperature range is inversely proportional to the amount of direct second stage graphitization that has occurred.

Some of the pearlite that forms, of course, begins to graphitize by the indirect second stage reaction, but this reaction is relatively slow and generally takes place by a "fading out" of the pearlite that does not involve a recession of the original boundaries. Graphitization of the pearlite was minimized by quenching the samples in water after they had reached 1200 F (643 C); this quench served further to assure that all possible direct second stage graphitization representative of a given cooling rate had taken place, as evidenced by the absence of any martensite (from untransformed austenite).

After a sample had been cooled in a helium atmosphere at a selected cooling rate, the point counting method was used to measure the amount of pearlite in the matrix. An area of 1.8 sq mm was examined at 200 ×, involving observations of at least 1000 grid intersections. Differences between measurements were 5 per cent of pearlite or less, when measured by two individuals.

It was found possible to reduce the number of individual specimens required to measure direct second stage graphitization by simply reaustenitizing one specimen as many times as needed to check each cooling rate. Several checks indicated the procedure was valid provided a specimen was held in helium at the reaustenitizing temperature of 1650 F (899 C) for at least 2 hr. Rehder and Wilson[9] have published data indicating that repeated annealing of one specimen increases direct second stage graphitization rates, but their reaustenitizing times of only 15 min and their use of a less neutral atmosphere probably account for their observations.

Indirect second stage graphitization was measured by following the rate of decrease of combined carbon of pearlitic specimens held at 1200 F (643 C) in helium. A series of specimens was obtained for each determination by air cooling from 1650 F (899 C) after first stage graphitization was complete. One specimen was retained as a base so that the combined carbon content at the beginning of the reaction could be determined; the others were reheated to 1200 F (643 C) for selected periods and then quenched in water. Slugs for combined-carbon analysis were cut out of the center of each specimen to eliminate any difficulty from surface effects.

### *Nodule Number Normalization**

In order to separate the intrinsic effects of the trace elements from their indirect effects caused by a changing of the nodule number (number of nodules/cu mm of iron) developed during graphitization, use was made of empirical expressions developed and reported on previously.[10] These expressions were derived from observations that in both first stage and direct second stage graphitization the changes in the log of reaction rates are linear with respect to the log of nodule number. The indirect second stage graphitization of pearlite is so insensitive to nodule number as to make normalization unnecessary.

Several symbols were adopted:

$T_F$ is the time to a given level of first stage graphitization, say 50 per cent.

$T_{FN}$ is the time $T_F$ that would have been observed had the nodule number remained constant after addition of an alloying element.

In all cases comparisons are made with a base iron, the original fraction of a split heat.

In the direct second stage reaction the term $R_C$ was adopted to indicate the linear cooling rate to give a certain percentage of pearlite in an iron, usually 50 per cent. The term $R_{CN}$, then, indicates the cooling rate that would have yielded 50 per cent pearlite had the addition of an alloying element not changed the nodule number of the base iron. The expressions developed are:

*First Stage Reaction*

$$T_{FN(alloy)} = T_{F(alloy)} - T_{F(base)} \left[ \left( \frac{N_{V(base)}}{N_{V(alloy)}} \right)^{0.29} - 1 \right]$$

*Second Stage Reaction*

$$R_{CN(alloy)} = R_{C(alloy)} - R_{C(base)} \left[ \left( \frac{N_{V(alloy)}}{N_{V(base)}} \right)^{0.45} - 1 \right]$$

The number of nodules/cu mm of iron, $N_v$, was obtained by counting the number of nodules observed (at 200 ×) within one sq mm of a polished cross-section of sample. Conversion from nodules/unit area to nodules/unit volume was done with the formula from Saltykov:[11]

$$N_V = 2.38\ N_A^{1.6}$$

*The term "normalization" is used throughout this paper in its mathematical sense and does not relate to the metallurgical heat treatment process.

where

$N_V$ = nodules/cu mm.
$N_A$ = nodules/sq mm.

Both expressions for normalizing nodule number differences between base and alloy iron heats are applicable to any specific level of graphitization. The nodule number values used in the formula for the first stage graphitization reaction should be measured at the level of graphitization at which $T_{FN}$ is determined. However, experience has shown that little error is introduced if the nodule number be measured at some other level.[10] For example, the nodule numbers measured after 100 per cent graphitization may be used when determining $T_{FN}$ at the 50 per cent level, and this was done in this paper.

The nodule numbers used in the formula for the direct second stage reaction are the same as those measured after 100 per cent first stage graphitization, provided the iron is held at the first stage graphitization temperature sufficiently long to minimize coarsening effects. Experience has shown that nodule number remains constant during direct second stage graphitization.[10]

## RESULTS

Because of space limitations, complete first and direct second stage graphitization curves will be given only for the element bismuth; for the other elements only the 50 per cent levels of graphitization will be considered. Complete graphitization curves of the other elements were obtained, however, and are generally available.[10] The results will be reported according to individual elements except for the indirect second stage reaction where the results will be presented collectively.

### *Bismuth*

Studies of the effects of Bi on first stage and direct second stage graphitization were conducted on two series of irons, one with an Mn-S ratio of 1.3, the other with an Mn-S ratio of 7.0. Figures 1 and 2 give the complete graphitization curves of the low Mn-S ratio series, Figs. 3 and 4 of the high Mn-S ratio series. Table 2 presents the data obtained at the 50 per cent levels of graphitization only.

It is seen that both reactions are retarded by the presence of Bi, regardless of the Mn-S ratio of the iron. The L-56-G iron, containing 0.083 per cent Bi showed a tendency for a few carbides to persist near the end of the first stage reaction. Figure 5 shows some of these persistent carbides.

Although such small amounts of carbide probably would have little influence on the properties of the iron, they may well account for statements in the literature that Bi severely retards the reaction, because some investigators have based measurements of the effects of trace elements on the time to eliminate carbide completely from the microstructure. Several of the other trace elements (Sb, Sn, Pb, Cd) were observed to develop a small amount of persistent carbides near the end of first stage graphitization, but there was never any effect on the graphitization curve until graphitization was about 95 per cent complete.

The graphitization rate values normalized for nodule number variation at the 50 per cent levels of

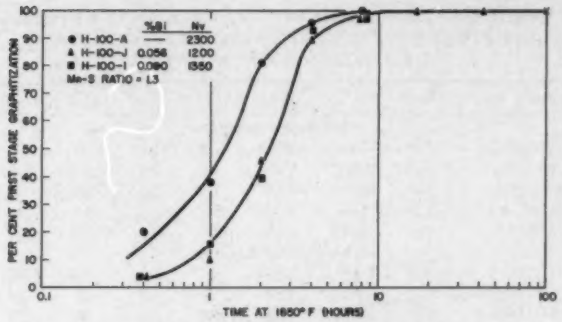



Fig. 1 — Bi effect on first stage graphitization of low Mn-S ratio irons.

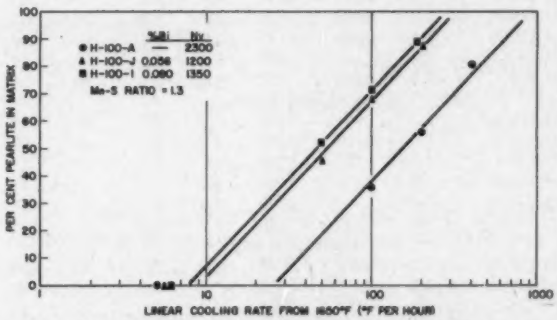



Fig. 2 — Bi effect on direct second stage graphitization in low Mn-S ratio irons.

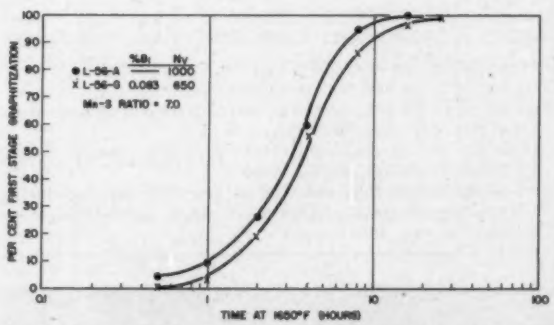



Fig. 3 — Bi effect on first stage graphitization of high Mn-S ratio irons.

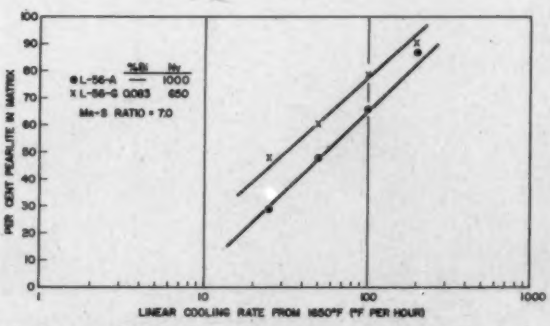



Fig. 4 — Bi effect on direct second stage graphitization of high Mn-S ratio irons.

TABLE 2 — TRACE AMOUNTS OF ELEMENTS EFFECT ON FIRST STAGE AND DIRECT SECOND STAGE GRAPHITIZATION KINETICS

| Heat. No. | Mn/S | Bi, % | $N_V$[1] | $T_F$[2] | $T_{FN}$[3] | $R_C$[4] | $R_{CN}$[5] |
|---|---|---|---|---|---|---|---|
| **Bismuth Series** | | | | | | | |
| H-100-A | 1.3 | — | 2300 | 1.2 | 1.2 | 150 | 150 |
| H-100-J | 1.3 | 0.056 | 1200 | 1.9 | 1.6 | 54 | 94 |
| H-100-I | 1.3 | 0.090 | 1350 | 1.9 | 1.65 | 46 | 76 |
| L-56-A | 7.0 | — | 1000 | 3.4 | 3.4 | 54 | 54 |
| L-56-G | 7.0 | 0.083 | 650 | 4.0 | 3.7 | 32 | 41 |
| **Antimony Series** | | | | | | | |
| H-116-A | 1.0 | — | 1550 | 3.7 | 3.7 | 34 | 34 |
| H-116-J | 1.0 | 0.010 | 1900 | 4.0 | 4.2 | 36 | 33 |
| H-116-I | 1.0 | 0.037 | 1700 | 4.0 | 4.1 | 25 | 24 |
| H-116-H | 1.0 | 0.055 | 1650 | 4.2 | 4.1 | 21 | 20 |
| L-56-A | 7.0 | — | 1000 | 3.4 | 3.4 | 54 | 54 |
| L-56-F | 7.0 | 0.083 | 900 | 4.0 | 3.8 | 21 | 27 |
| **Tin Series** | | | | | | | |
| H-116-A | 1.0 | — | 1550 | ~3.7 | 3.7 | 34 | 34 |
| H-116-G | 1.0 | 0.021 | 1850 | ~3.7 | 3.9 | 72 | 69 |
| H-116-F | 1.0 | 0.067 | 1850 | ~3.7 | 3.9 | 66 | 63 |
| H-116-E | 1.0 | 0.081 | 1150 | ~3.7 | 3.6 | 56 | 59 |
| L-56-A | 7.0 | — | 1000 | 3.4 | 3.4 | 54 | 54 |
| L-56-E | 7.0 | 0.045 | 650 | 4.2 | 3.5 | 30 | 39 |
| **Lead Series** | | | | | | | |
| H-116-A | 1.0 | — | 1550 | ~3.7 | 3.7 | 34 | 34 |
| H-116-D | 1.0 | 0.003 | 1550 | ~3.7 | 3.7 | 68 | 68 |
| H-116-C | 1.0 | 0.021 | 1850 | ~3.7 | 3.5 | 76 | 73 |
| H-116-B | 1.0 | 0.038 | 1400 | ~3.7 | 3.8 | 50 | 51 |
| L-56-A | 7.0 | — | 1000 | 3.4 | 3.4 | 54 | 54 |
| L-56-D | 7.0 | 0.023 | 540 | 4.2 | 3.5 | 35 | 47 |
| **Cadmium Series** | | | | | | | |
| H-128-A | 1.3 | — | 3900 | 3.3 | 3.3 | 36 | 36 |
| H-128-F | 1.3 | 0.002 | 3550 | 3.7 | 3.5 | 45 | 47 |
| H-128-E | 1.3 | 0.003 | 2000 | 4.2 | 3.5 | 45 | 55 |
| L-56-A | 7.0 | — | 1000 | 3.4 | 3.4 | 54 | 54 |
| L-56-H | 7.0 | 0.002 | 550 | 4.3 | 3.5 | 25 | 37 |
| **Zinc Series** | | | | | | | |
| H-116-A | 1.0 | — | 1550 | ~3.7 | 3.7 | 34 | 34 |
| H-116-L | 1.0 | 0.017 | 1450 | ~3.7 | 3.6 | 67 | 68 |
| H-120-A | 1.3 | — | 4600 | ~3.3 | 3.3 | 45 | 45 |
| H-120-B | 1.3 | 0.003 | 3950 | ~3.3 | 3.2 | 61 | 65 |
| H-120-C | 1.3 | 0.022 | 3100 | ~3.3 | 3.0 | 92 | 100 |

(1) Nodules/cu mm after 100% first stage graphitization.
(2) Time in hr for 50% first stage graphitization.
(3) Time in hr for 50% first stage graphitization after normalization for nodule number differences.
(4) Cooling rate in deg. Fahr./hr to give 50% pearlite during direct second stage graphitization.
(5) Cooling rate in deg. Fahr./hr to give 50% pearlite during direct second stage graphitization after normalization for nodule number differences.

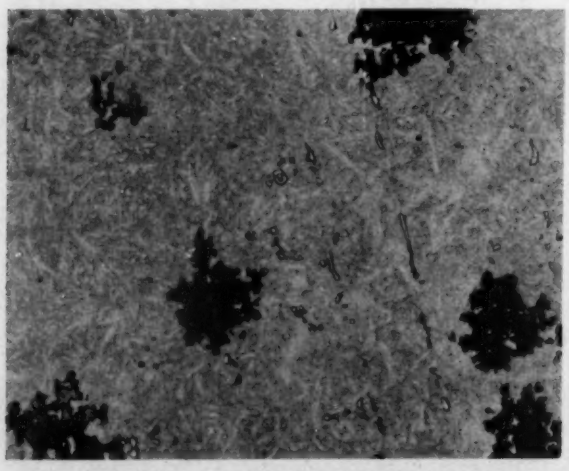


Fig. 5 — Persistent carbides in iron containing Bi near end of first stage graphitization reaction (L-56-G). 200 ×.

graphitization ($T_{FN}$ and $R_{CN}$) are given in Table 2, and are plotted in Figs. 6 and 7. It is seen that Bi retards intrinsically both first stage and direct second stage graphitization in irons with low and high Mn-S ratios. The effect of adjusting for nodule number is to make the retarding effects appear less severe. This is because Bi, along with several other of the trace elements, tends to reduce nodule number, a retarding influence in itself.

### *Antimony*

The two series of irons used to investigate the effects of antimony on the first stage and direct second stage reactions had Mn-S ratios of 1.0 and 7.0. The data obtained at the 50 per cent levels of graphitization are given in Table 2. Observing the changes in $T_F$ and $R_C$ with increasing Sb, it is seen that Sb slows the first stage and direct second stage reactions slightly, regardless of the Mn-S ratio of the iron. The one exception was the iron containing 0.01 per cent antimony (H-116-J) which graphitized slightly faster in direct second stage than did the base iron.

The 50 per cent graphitization rate values of the two series of irons normalized for nodule number variations are plotted in Figs. 8 and 9. It is immediately evident that the fast overall direct second stage graphitization of the H-116-J iron is accounted for by the fact that this iron developed a higher nodule number than did the base iron (H-116-A) probably because of some uncontrolled variable during melting and solidification. The normalized 50 per cent values plotted in Figs. 8 and 9 show that Sb acts similarly to Bi and retards both reactions intrinsically, whatever the Mn-S ratio of the iron.

### *Tin*

The two series of irons used to evaluate the effects of Sn on first stage and direct second stage graphitization were of Mn-S ratio 1.0 and 7.0. As indicated in Table 2, there was little effect of Sn on the first

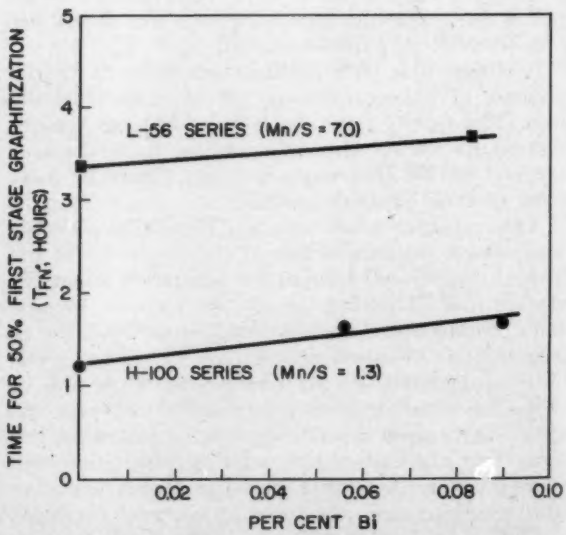



Fig. 6 — Intrinsic effects of Bi on first stage graphitization in irons of low and high Mn-S ratio.

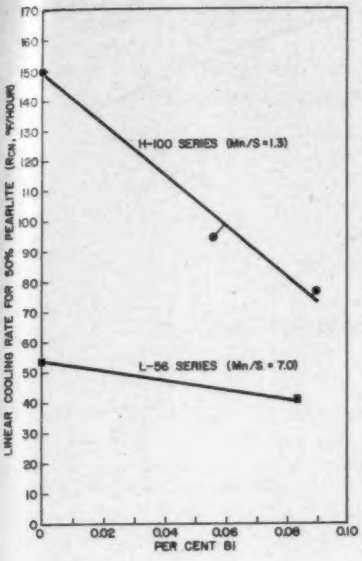



**Fig. 7 — (left) Intrinsic effects of Bi on second stage graphitization in irons of low and high Mn-S ratio.**

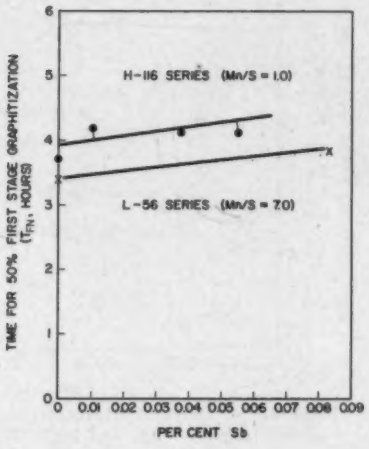



**Fig. 8 — (right) Intrinsic effects of Sb on first stage graphitization of low and high Mn-S ratio irons.**

stage graphitization rate in the series with the lower Mn-S ratio. On the direct second stage reaction, up to 0.02 per cent Sn increased the graphitization rate, but a gradual decrease followed as Sn was increased to 0.081 per cent. In the series with Mn-S ratio of 7.0, the presence of 0.045 per cent Sn decreased the rates of both the first stage and direct second stage reactions.

The 50 per cent values of graphitization normalized for nodule number variation are plotted in Figs. 10 and 11. Figure 10 shows that the intrinsic effects of Sn are negligible on first stage graphitization if the Mn-S ratio of the iron is 1.0. However, when the Mn-S ratio is 7.0 Sn retards the reaction slightly. Figure 11 shows that the intrinsic effects of Sn on the direct second stage reaction are similar to the overall effects; up to 0.02 per cent Sn promotes the reaction in the irons of low Mn-S ratio, but greater amounts cause a progressive retardation.

The presence of Sn retards the reaction when the Mn-S ratio is 7.0. It would appear therefore that Sn may interact with S in irons with low Mn-S ratio and thereby speed up the reaction; once the reaction with S nears completion, greater amounts of Sn may cause retarding because the iron then behaves as if the Mn-S ratio of the iron were greater.

### *Lead*

The results on Pb were almost identical with those on Sn, as may be seen in Table 2 and in Figs. 12 and 13.

### *Cadmium*

Two series of irons with Mn-S ratios of 1.3 and 7.0 were used to study the effects of Cd on the first stage and direct second stage graphitization reactions. As shown in Table 2, the presence of Cd retards the first stage reaction slightly in both series of irons. On the direct second stage reaction, Cd speeds graphitization in the irons with low Mn-S ratio, but retards graphitization in the irons with high Mn-S ratio.

The intrinsic effects of Cd on first stage graphitization are small in irons with both low and high Mn-S ratio, as shown in Fig. 14. The intrinsic effects of Cd on the direct second stage reaction, however, are again to promote the reaction in low Mn-S ratio irons and retard in high Mn-S ratio irons, as in Fig. 15. Thus, Cd apparently interacts with S in irons with low Mn-S ratio, but not in irons with high Mn-S ratio, behaving similarly in Sn and Pb in this respect. Probably Cd would retard the direct second stage reaction in irons of low Mn-S ratio should it be present in amounts greater than needed to interact with

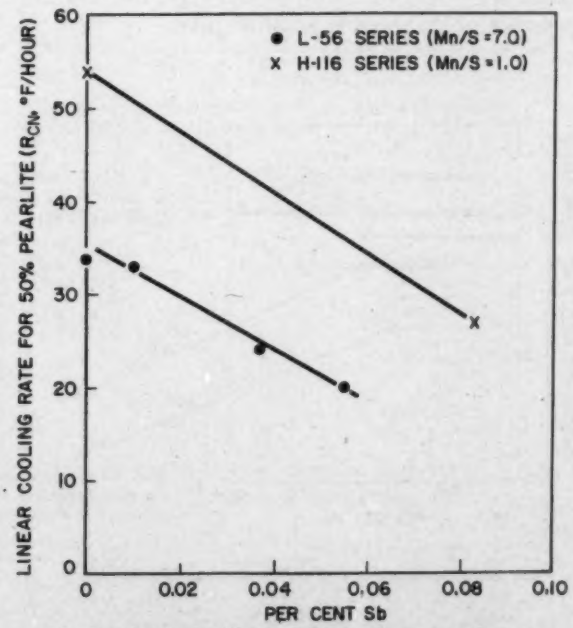



**Fig. 9 — Intrinsic effects of Sb on direct second stage graphitization of low and high Mn-S ratio irons.**

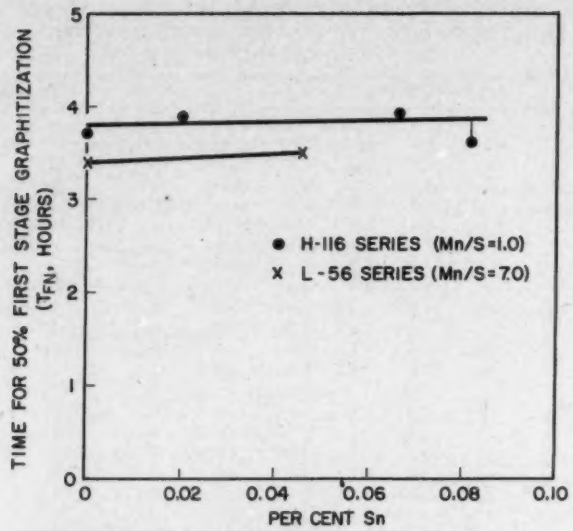



Fig. 10 — (above) Intrinsic effects of Sn on first stage graphitization of low and high Mn-S ratio irons.

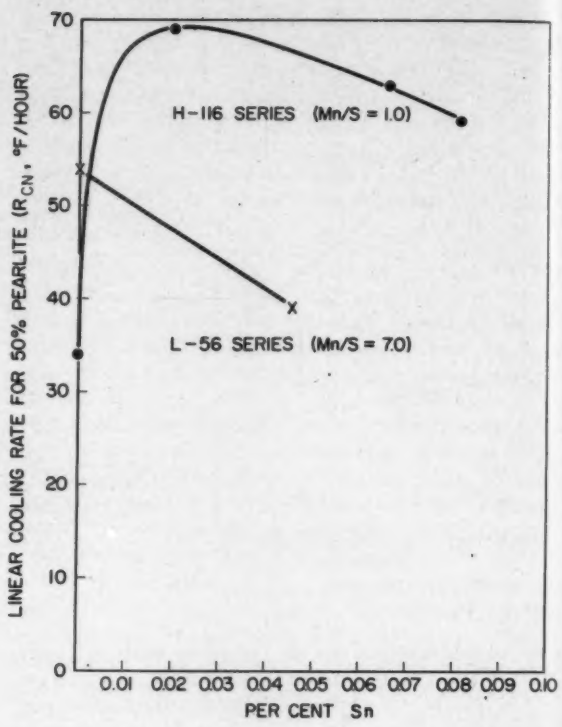



Fig. 11 — (right) Intrinsic effects of Sn on the direct second stage graphitization of low and high Mn-S ratio irons.

the excess S, similar to the characteristics of Sn and Pb.

*Zinc*

Investigations of the effects of Zn on the first stage and direct second stage graphitization reactions were limited to two series of irons with low Mn-S ratio (1.0 and 1.3). As shown in Table 2, there was no effect of Zn on the first stage graphitization rate of either series. On the direct second stage reaction, the presence of Zn speeded graphitization.

The intrinsic effect of Zn on the first stage graphitization reaction of irons with low Mn-S ratio are given in Fig. 16. A slight tendency for Zn to promote the reaction is indicated, but the effect is small and may not be significant. The presence of Zn does promote intrinsically the direct second stage reaction in these irons, however, as shown in Fig. 17. Thus, Zn is apparently another element that may interact with the S in irons with low Mn-S ratio.

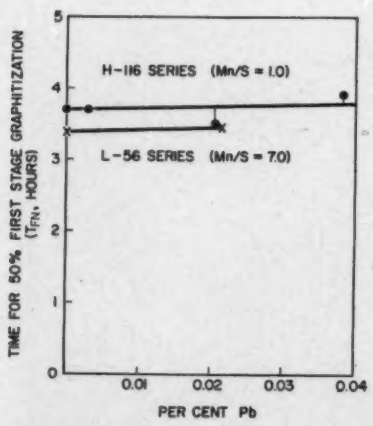
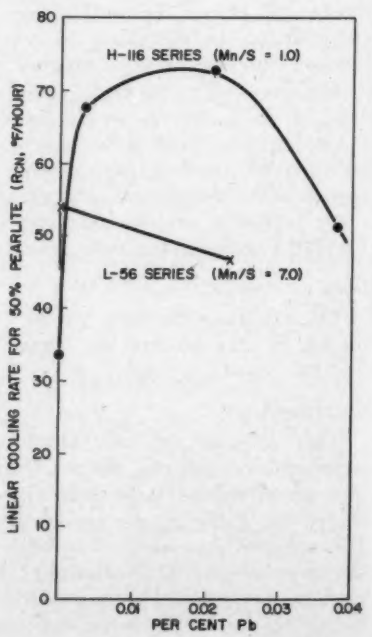



Fig. 12 — (left) Intrinsic effects of Pb on first stage graphitization of low and high Mn-S ratio irons.

Fig. 13 (right) Intrinsic effects of Pb on direct second stage graphitization of irons of low and high Mn-S ratio irons.



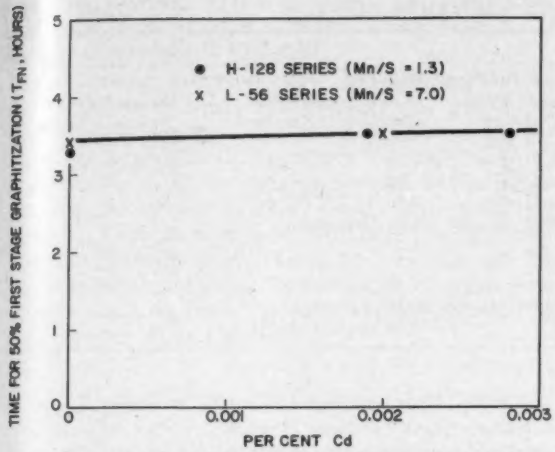



Fig. 14 — Intrinsic effects of Cd on first stage graphitization of irons of low and high Mn-S ratio irons.

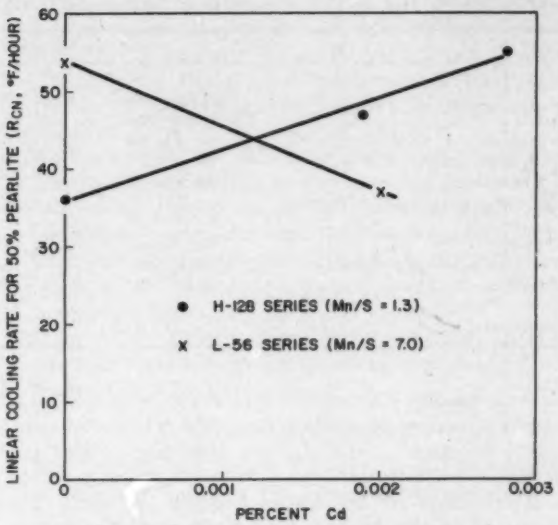



Fig. 15 — Intrinsic effects of Cd on direct second stage graphitization of low and high Mn-S ratio irons.

*Indirect Second Stage Reaction*

Studies of the effects of the trace elements on the indirect second stage reaction were limited to a series of irons prepared from a split heat and having an Mn-S ratio of 7.0. The results are plotted in Fig. 18, and show that trace amounts of the elements studied have little effect on the reaction in iron with high Mn-S ratio. Although no data were obtained on irons with low Mn-S ratio, the only possible effects that could appear therein would be some promoting action by interaction with S.

## DISCUSSION OF RESULTS

The results of the trace element graphitization studies have been summarized in Table 3. For the first stage reaction, all of the trace elements either retard slightly or have no definite effects. In irons with low Mn-S ratio, where nodule number effects are not normalized, the elements Sn, Pb and Zn have no definite effects, Sb retards slightly and Bi and Cd slightly. Some of these results are accounted for by nucleation or nodule number effects. Thus, when the results are normalized for nodule number, the apparent retarding effect of Cd vanishes and that of Bi is less.

In irons with high Mn-S ratio, the elements Bi, Sb, Sn, Pb and Cd all show a slight retarding effect on first stage graphitization, although a substantial part of the retarding is caused by nodule number effects. In addition, all of the elements (Zn was not studied)

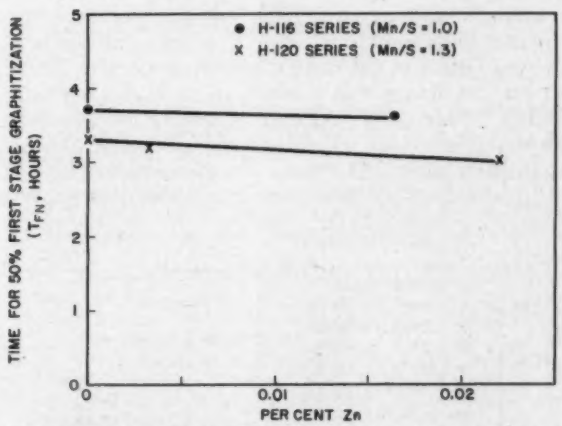



Fig. 16 — Intrinsic effects of Zn on first stage graphitization of low Mn-S ratio irons.

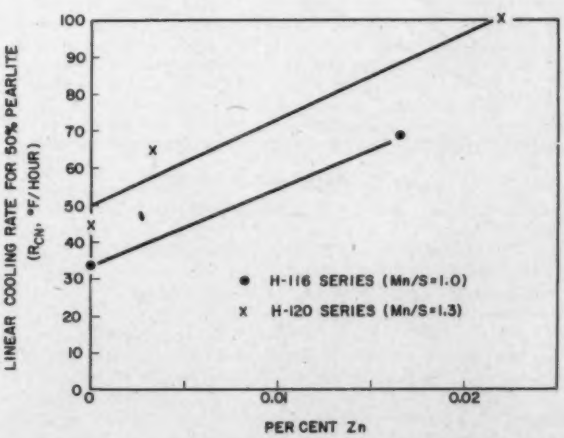



Fig. 17 — Intrinsic effects of Zn on direct second stage graphitization of low Mn-S ratio irons.

TABLE 3 — SUMMARY OF TRACE ELEMENTS EFFECTS

| | Low Mn-S Ratio Irons | | High Mn-S Ratio Irons | |
|---|---|---|---|---|
| | **First Stage Reaction** | | | |
| Effect | Not Normalized For Nodule Number | Normalized for Nodule Number | Not Normalized For Nodule Number | Normalized for Nodule Number |
| No Effect | Sn, Pb, Zn | Sn, Pb, Cd, Zn | — | — |
| Very slight retard | Sb | Bi, Sb | — | Bi, Sb, Sn, Pb, Cd |
| Slight retard | Bi, Cd | — | Bi, Sb, Sn, Pb, Cd | — |
| Persistent final traces of carbide | — | — | Bi, Sb, Sn, Pb, Cd | — |
| | **Direct Second Stage Reaction** | | | |
| Promote | Cd, Zn | Cd, Zn | — | — |
| Promote followed by retard | Sn, Pb, Sb | Sn, Pb | — | — |
| Retard | Bi | Bi, Sb | Bi, Sb, Sn, Pb, Cd | Bi, Sb, Sn, Pb, Cd |
| | **Indirect Second Stage Reaction** | | | |
| No effect | — | — | Bi, Sb, Sn, Pb, Cd | — |

show a tendency for traces of carbide to persist near the completion of the reaction, but not in amounts likely to greatly influence the ductility of the material.

The effects of the trace elements are more pronounced on the direct second stage graphitization reaction. In irons with low Mn-S ratio, where nodule number differences are not normalized, the elements Cd and Zn promote graphitization, the elements Sn, Pb and Sb promote graphitization in small amounts but retard in larger amounts, and the element Bi retards in any amount. After normalizing these results for nodule number effects, it is seen that the promoting effects of Cd and Zn are intrinsic. The effects of Pb and Sn are also intrinsic in their characteristic of first accelerating and then retarding the reaction. However, Sb is seen to be an intrinsic retarder like Bi.

In irons with high Mn-S ratio the elements Bi, Sb, Sn, Pb and Cd all retard direct second stage graphitization. Although not indicated in Table 3, the retarding effects of these elements are in some degree attributable to nodule number effects.

Table 3 also indicates that the effects of Bi, Sb, Sn, Pb and Cd on the indirect second stage reaction are negligible in irons with high Mn-S ratio.

### *Trace Elements Potentials*

The observations indicated in Table 3 are explainable in terms of the relative abilities of the trace elements to combine with S in competition with iron and manganese. The relative potentials of the trace elements in question to combine with sulfur bound as FeS may be estimated from published free energy of formation data.[12-15] Figure 19 shows schematically the results of such estimates. It is seen that the elements Bi and Sb would not be expected to remove much S from FeS whereas the elements Mn, Zn and Cd would; Pb and Sn are borderline cases. None of

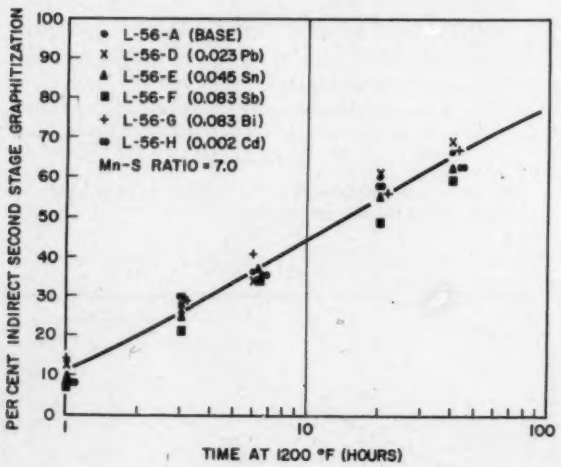



Fig. 18 — Pb, Sn, Sb, Bi and Cd effects on indirect second stage graphitization of pearlite in high Mn-S ratio irons.

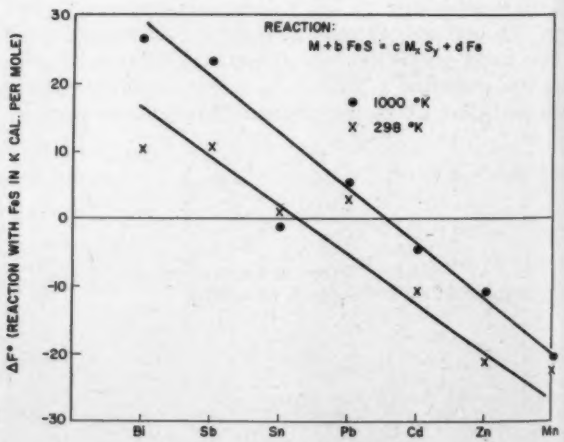



Fig. 19 — Schematic representation of relative free energy values for removal of S from FeS.

the trace elements would be expected to remove sulfur from MnS, however, for this is the most stable sulfide of the group.

It is evident, therefore, that the elements Bi and Sb would not be expected to accelerate graphitization rates by interactions with S in irons with low Mn-S ratio, whereas the elements Cd and Zn would probably do so. The elements Pb and Sn would be expected to interact with S to some extent. Referring again to Table 3, it is seen that this is the behavior observed experimentally. Thus, the elements Cd, Zn, Sn and Pb intrinsically promote the direct second stage reaction in irons with low Mn-S ratio, whereas the elements Bi and Sb retard.

The fact that the elements Sn and Pb also begin to retard the reaction as their amounts are increased is evidence that

a) these elements have rather low potential for combining with sulfur.
b) that the reactions with sulfur become essentially completed and thereafter the two elements act as if the iron were of higher Mn-S ratio.

The elements Cd and Zn would possibly also begin to retard the reaction if they were added in sufficient quantities.

The same general effects of the trace elements are observed on the first stage graphitization reaction in irons with low Mn-S ratio, but the effects are less pronounced. This is in accordance with the known less sensitivity of this reaction to the presence of excess sulfur.[16,17]

In irons with high Mn-S ratio none of the trace elements would be expected to interact with sulfur, for their potentials of combination with sulfur are all less than that of manganese. Thus, in irons with high Mn-S ratio the trace elements are free to manifest their intrinsic retarding effects. In such cases the intrinsic retarding actions observed may be associated with solution in and stabilization of cementite, or by influences on the relative pearlite and ferrite hardenabilities, as have been proposed elsewhere to explain the retarding effects of Mn and sulfur.[10]

## SUMMARY AND CONCLUSIONS

The effects of trace amounts of the elements Bi, Sb, Sn, Pb, Cd and Zn on the three major graphitization reactions in malleable irons were investigated. Irons with both low and high Mn-S ratio were employed to show possible interactions of the trace elements with sulfur. Normalization of the results with respect to nodule number differences between irons was carried out in order that the intrinsic effects of the trace elements could be revealed. The main conclusions are:

1) The effects of the trace elements, Sn, Pb and Cd vary with the Mn-S ratio of the iron. The effects of Bi and Sb do not change with Mn-S ratio.
2) In irons with low Mn-S ratio the elements Sn, Pb, Cd and Zn have negligible intrinsic effects on first stage graphitization. The elements Bi and Sb retard slightly.
3) On the direct second stage reaction in irons with low Mn-S ratio, the elements Cd and Zn promote intrinsically, the elements Sn and Pb promote in small amounts but begin to retard in larger amounts and the elements Bi and Sb retard in all amounts. These effects are explainable in terms of the relative potentials of trace elements for combination with sulfur in the presence of iron.
4) In irons with high Mn-S ratio all of the trace elements (Zn was not studied) retard slightly both the first stage and direct second stage reactions. The retarding of the first stage reaction is greatest near the end of the reaction, suggesting that the effects of the trace elements are associated with cementite stability.
5) The trace elements have little influence on the indirect second stage graphitization of pearlite in irons with high Mn-S ratio.

## ACKNOWLEDGMENT

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# FACTORS AFFECTING METAL-MOLD REACTIONS

by G. A. Colligan, L. H. Van Vlack and R. A. Flinn


## ABSTRACT

Previous work has determined the phase equilibria pertinent to iron and some iron-manganese alloys with silica. These investigations evaluated the equilibria at fixed temperatures and atmospheres ($CO_2/CO$ gas mixtures).

The actual atmospheres obtained in a sand mold during casting operations are of great concern. In order to correlate the equilibrium data with actual practice, the effect of manganese additions and atmosphere on interface reaction was qualitatively evaluated. Identical step castings were produced in green sand and shell sand molds from iron base alloys containing 0.23 to 8.60 per cent manganese. The surface sintered layers were removed from all castings and microscopically examined.

The effect of an oxidizing mold atmosphere was demonstrated in the green sand castings by extensive reaction at the interface. The atmosphere produced by phenolic resin bonded shell molds resulted in slight reaction and reaction products typical of a less oxidizing atmosphere than green sand molds.

Manganese additions caused extensive silicate melt formation. The reaction became more pronounced with increasing manganese content in green sand molds.

A practical means of minimizing metal-mold reactions may be possible through the use of organic sand binders and additives which can control the interface atmosphere.


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(This paper is based upon a thesis submitted by G. A. Colligan to the Horace H. Rackham School of Graduate Studies of the University of Michigan in partial fulfillment of the requirements for the degree of Ph.D. in Met. Engrg.)

## INTRODUCTION

Reactions at the mold-metal interface have produced serious problems in the field of cast metals. The reaction, which is the cause of fused sand, surface roughness and penetration, has been studied qualitatively by many investigators.[1-11] The preliminary work of the present authors[12] was the first attempt to study the pertinent phase equilibria in the iron-silica system where temperature and atmosphere were carefully controlled.

This preliminary work led to a comprehensive phase equilibria investigation which included the effect on interface reaction of manganese additions to the iron-base alloy. These results have been reported and are being published at the present time[13] Equilibrium diagrams summarizing these investigations are presented in Figs. 1 and 2.

Figure 1 shows that extremely reducing conditions are necessary to prevent a silicate melt (interface reaction). Furthermore, even small additions of manganese (Fig. 2) to the iron-base alloy make it difficult to avoid mold-metal reaction during solidification and cooling.

In view of these new equilibrium data, it seemed advisable to determine whether these effects were

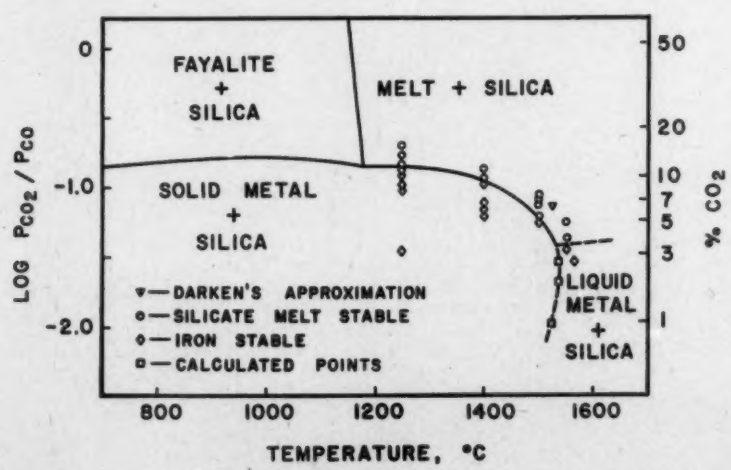



Fig. 1 — Fe-Si-O-C system from Darken[15] modified to include recent experimental results.[13] The system contains a strong excess of silica.

encountered under the nonequilibrium conditions experienced in commercial castings.

## PROCEDURE

The actual atmospheres obtained in a sand mold during casting operations are of great concern. In order to correlate the equilibrium data[13] with actual practice, the effect of manganese additions and atmosphere on interface reaction was qualitatively evaluated on test castings in the following manner. Identical step castings were produced in green sand molds, shell molds (Fig. 3) and molds containing a shell sand drag and green sand cope. The sand mixtures are summarized in Table 1. By using a green sand mixture, an oxidizing condition is produced at the mold-metal interface.[11] The resin bonded shell mold produces strongly reducing conditions as determined by Van Vlack, Wells and Pierce.[14]

TABLE 1 — SAND MIXTURES FOR TEST CASTINGS

| Description | Constituent | Weight, % |
|---|---|---|
| Green Sand | N. J. Silica Sand 140 mesh | 92 |
| | Western Bentonite | 3.5 |
| | Mogul Cereal | 1 |
| | Water | 3.5 |
| Shell Sand | N. J. Silica Sand 140 mesh | 95 |
| | Phenolic Resin | 5 |

### *Melting Procedure*

Cold-rolled A.I.S.I. 1018 bar stock was melted in a high frequency induction furnace. After melt-down, a small silicon addition was made to deoxidize the melt and allow recarbonization to the desired level. This level was calculated from experimental equilibrium data at 1550 C* (2822 F) under controlled atmosphere. The chemical compositions of these castings and pouring temperatures are recorded in Table 2, which contains a complete summary of experimental data. Low pouring temperatures were used initially to duplicate equilibrium data.

The manganese additions were made to the metal remaining in the furnace after the simple iron-carbon alloy was poured.

### *Casting Procedure*

Iron base compositions with 0.23, 0.75 and 4.40 per cent manganese were each initially cast in a green sand mold and a shell mold to determine the effect of atmosphere and manganese content on interface reaction.

In order to eliminate the problem of pouring temperature variation, duplex molds were poured with 0.50 and 8.60 per cent manganese. These molds each contained a shell sand drag and a green sand cope. In this way a measure of the effect of atmosphere at the interface on reaction could be determined in one mold. All castings were poured from small ladles after the metal was tapped from the furnace.

*See Appendix

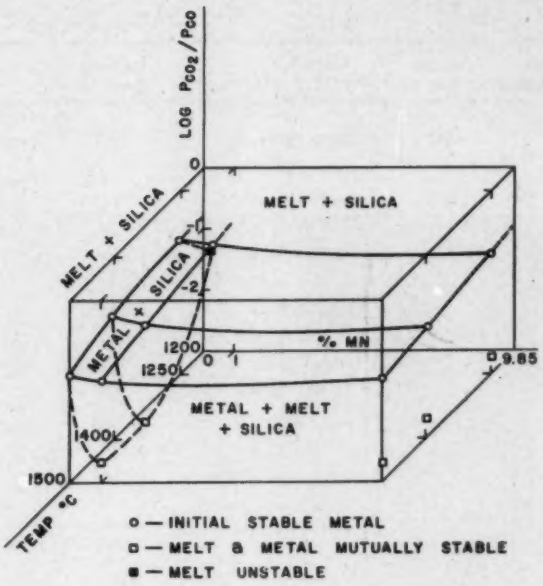



Fig. 2 — Fe-Si-O-C-Mn system in the presence of excess silica.

### *Interface Reaction Evaluation*

All castings were observed after cooling, and in some cases were photographed to provide a permanent record. The surface sintered layer was carefully removed from the castings for microscopic examination. Samples of appropriate sintered layer specimens were vacuum impregnated with thermosetting resin, mounted in thermoplastic resin and polished for high magnification observation.

## RESULTS

The results of this qualitative investigation may best be considered under the following categories— (a) macroscale evaluation and (b) microscopic evaluation of interface reaction.



Fig. 3 — Shell mold for test step casting. Silica sand bonded with 5 per cent phenolic resin.

TABLE 2 — TEST CASTINGS SUMMARY

| Heat No. | Casting No. | Mold Type | Pouring Temp., F | Surface Condition | Chemical Analysis (wt.,%) C | Mn | Si | P | S |
|---|---|---|---|---|---|---|---|---|---|
| 1 | E-1 | Green Sand | 2810 | poor surface, extensive reaction | 0.16 | 0.19 | 0.13 | 0.011 | 0.020 |
| | E-2 | Shell Mold | 2760 | smooth surface slight reaction | 0.20 | 0.23 | 0.21 | 0.014 | 0.036 |
| 2 | E-3 | Green Sand | 2810 | poor surface, extensive reaction, low fluidity | 0.18 | 4.10 | 0.33 | 0.011 | 0.019 |
| | E-4 | Shell Mold | 2730 | smooth surface, slight reaction, good fluidity | 0.20 | 4.40 | 0.45 | 0.010 | 0.018 |
| 3 | E-5 | Green Sand | 2950 | extensive reaction, penetration, fusion | 0.10 | 0.75 | 0.13 | 0.011 | 0.021 |
| | E-6 | Shell Mold | 2940 | slight reaction, fusion, no penetration | 0.12 | 1.03 | 0.26 | 0.013 | 0.021 |
| 4 | E-7 | Shell Mold Drag Green Sand Cope | 2950 | slight reaction on drag, extensive reaction on cope | 0.16 | 0.50 | 0.34 | 0.012 | 0.021 |
| 5 | E-8 | Shell Mold Drag Green Sand Cope | 2800 | slight reaction on drag, extensive reaction on cope | 0.15 | 8.60 | 0.41 | 0.010 | 0.019 |



Fig. 4 — Test step casting E-1, 0.19 per cent manganese-iron alloy cast in green sand mold.



Fig. 5 — Test step casting E-2, 0.23 per cent manganese-iron alloy cast in shell sand mold.



Fig. 6 — Test step casting E-3, 4.10 per cent manganese-iron alloy cast in green sand mold.

### *Macroscale Evaluation*

The first six step castings, each set of two containing a different nominal addition of manganese, were poured into green sand molds and shell molds. The details of the analysis, mold materials and pouring temperatures are recorded in Table 2. It should be noted that metal of a given nominal composition, as poured in the green sand mold, consistently yields lower manganese content in the final castings.

In general, the difference between green sand and shell sand was obvious at all levels of manganese. The oxidizing nature of the green sand mold produced extreme interface reaction, which resulted in poor surface finish, some metal penetration and extensive fusion of the silica at the interface. The reducing nature of the shell sand mold permitted little interface reaction. The surfaces were smooth, free of metal penetration and exhibited slight fluxing of silica.

Figure 4 shows the 0.19 per cent manganese alloy poured in a green sand mold. The heavy sintered layer and poor surface finish is obvious. Figure 5 is the same alloy poured in a shell mold. There is slight interface reaction and almost no sintered layer, i.e., fused zone of reaction products. The surface finish is smooth and the detail is far superior to the green sand casting in spite of the lower pouring temperature shown in Table 2.

Figure 6 shows the 4.10 per cent manganese alloy poured in a green sand mold. The extensive interface reaction has produced a heavy sintered layer and extensive silica fusion. Figure 7 illustrates the appearance of the 4.40 per cent manganese alloy poured in a shell mold. The casting exhibits good surface finish, some light reaction and silica fusion. The shell mold casting poured at the lower temperature has good detail and fluidity.

A second set of two castings was produced to eliminate the problem of pouring temperature variation. These molds consisted of a shell sand drag and a green sand cope. The first casting contained 0.50



Fig. 7 — Test step casting E-4, 4.40 per cent manganese-iron alloy cast in shell sand mold.

per cent manganese, and exhibited a heavy sintered layer at the green sand interface and light reaction at the shell sand interface. The second casting was an 8.60 per cent manganese alloy. The green sand interface of this casting was rough and contained a heavy sintered layer. The shell sand interface was smooth with slight reaction and had no heavy adhering sintered layer.

### *Microscopic Evaluation of Interface Reaction*

Microscopic examination of the sintered layer removed from the experimental castings yielded interesting results. Figure 8 contains the reaction products from the sintered layer of the 0.19 per cent manganese alloy cast in a green sand mold. There is extensive evidence of the formation of a liquid silicate melt at high temperature. This melt has essentially crystallized to fayalite on cooling. In addition to the quartz and cristobalite, the specimen contains a great deal of magnetite (the small skeletal crystals visible in Fig. 8).

Figure 9 shows the sintered layer from the 0.23 per cent manganese alloy cast in a shell mold. There is a small amount of silicate melt from the reaction at elevated temperature which has subsequently solidified as fayalite.

Figure 10 illustrates the typical structure of the reaction interface from the 4.10 per cent manganese

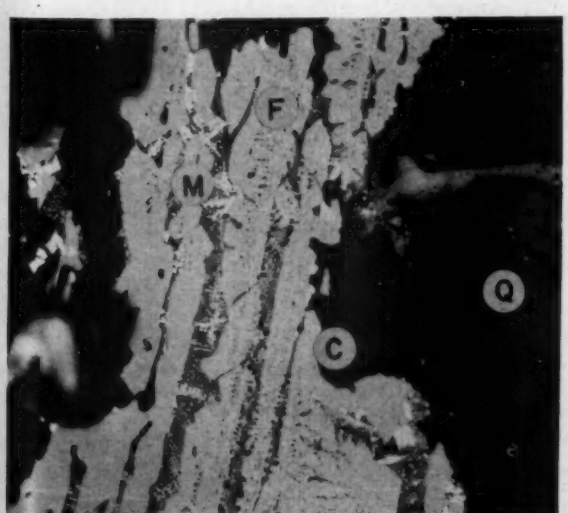



Fig. 8 — Microstructure of sintered layer from test step casting E-1, containing Magnetite (M), Fayalite (F), Cristobalite (C) and Quartz (Q). Iron-0.19 per cent manganese alloy cast in green sand mold. 500 ×.

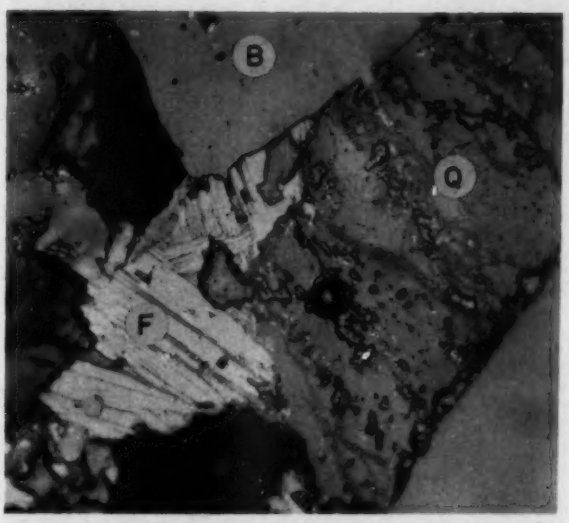



Fig. 9 — Microstructure of sintered layer from test step casting E-2, containing Fayalite (F), Quartz (Q) and Bakelite (B). Iron-0.23 per cent manganese alloy cast in shell sand mold. 500 ×.

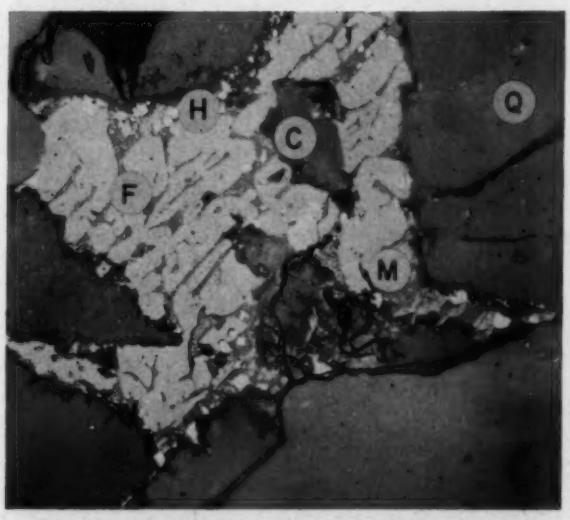



Fig. 10 — Microstructure of sintered layer from test step casting E-3, containing Hematite (H), Magnetite (M), Fayalite (F), Cristobalite (C) and Quartz (Q). Iron-4.10 per cent manganese alloy cast in green sand mold. 500 ×.




Fig. 11 — Microstructure of sintered layer from test step casting E-4, containing Iron (I), Fayalite (F), Cristobalite (C) and Quartz (Q). Iron-4.40 per cent manganese alloy cast in shell sand mold. 500 ×.

alloy cast in a green sand mold. There is extensive fayalite from the liquid silicate melt as well as magnetite and hematite. Figure 11 shows an alloy of the same nominal composition, 4.40 per cent manganese, poured into a shell mold. The sintered layer contains some fayalite from the liquid silicate melt and some spherical metallic crystals in the zone of the prior silicate melt.

The structure in the 0.75 per cent manganese alloys is the same. The green sand sintered layer reveals extensive reaction. The shell sand sintered layer indicates slight reaction with considerable reduction of iron from the silicate melt.

The duplex molds containing shell sand drag and green sand cope substantiate the results of the first series of castings. Figure 12 shows the sintered layer from the green sand cope of the 8.60 per cent manganese alloy. The extensive reaction zone contains fayalite and magnetite. The shell mold drag section of the same casting exhibits a sintered layer of the type illustrated by Fig. 13. The small amount of fayalite in this section is evidence that the reaction was slight, but there was extensive iron reduction when the liquid silicate melt was cooled.

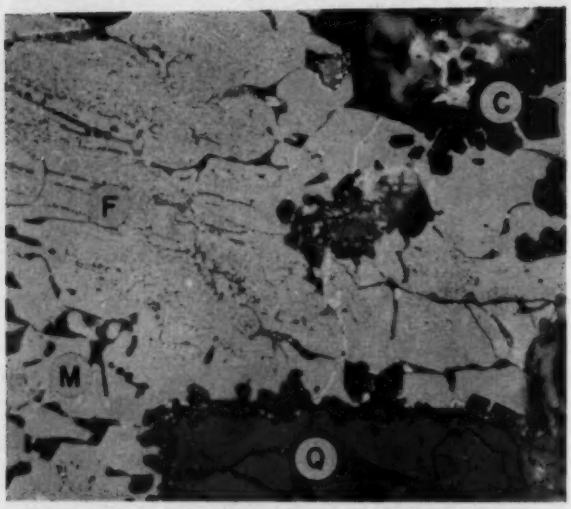



Fig. 12 — Microstructure of sintered layer from green sand cope of test step casting E-8, containing Magnetite (M), Fayalite (F), Cristobalite (C) and Quartz (Q). Iron-8.60 per cent manganese alloy cast in duplex mold. 500 ×.

## ENGINEERING SIGNIFICANCE

The engineering significance of this investigation will be discussed in the following order—(a) penetration and interface reaction, (b) problem of manganese additions and (c) effect of mold material.

### *Penetration and Interface Reaction*

There are two possible causes of metal penetration in a silica mold. Simple mechanical penetration occurs when the pressure exceeds the interfacial energy of the mold wall and causes the liquid metal to fill the interstices of the mold. In addition, chemical reaction at the interface may cause the mold wall to break down and allow metal penetration at low pressures.

There is extensive reference by earlier investigators of the formation of a low melting point slag at the

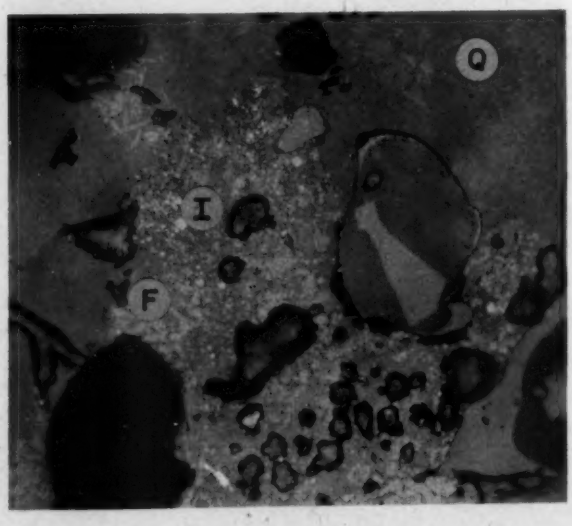



Fig. 13 — Microstructure of sintered layer from shell sand drag of test step casting E-8, containing Iron (I), Fayalite (F) and Quartz (Q). Iron-8.60 per cent manganese alloy cast in duplex mold. 500 ×.

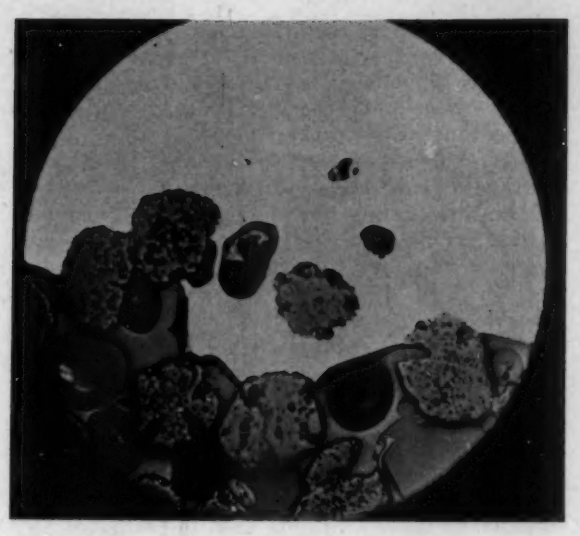


Fig. 14 — Microstructure of iron-silica interface after one hr at 1550 C (2822 F) in an oxidizing atmosphere of 3.9 per cent $CO_2$ — 96.1 per cent CO. The penetration of the iron is typified by the isolated quartz grains surrounded by iron. 25 ×.

mold-metal interface, but there is considerable confusion between the penetration of metal and the penetration of the low melting point slag. The same confusion has existed with respect to the terms "reducing" and "oxidizing" atmosphere, and the effect of these atmospheres on reaction and penetration.

The results of this investigation combined with the phase equilibria studies of the present authors[13] clarify the role of atmosphere in reaction and penetration. With pure iron-carbon alloys, the equilibrium diagram (Fig. 1) presents the range of atmosphere control necessary to prevent interface reaction, i.e., the formation of a liquid silicate melt. Observation of the phases present in the sintered layers of the green sand and shell sand molds indicates that normal cast metal practice operations lie above the boundaries between the liquid silicate melt field and the solid and liquid iron fields. The result is that extensive liquid silicate melt formation is experienced, with resultant loss of surface smoothness and detail.

The possibility of preventing interface reaction by the introduction of neutral or reducing gases in a mold is, therefore, slight and most impractical. Even vacuum treatment is not necessarily a potential solution. The atmospheres experienced in normal practice should more properly be classified on a relative oxidizing power basis, since their behavior is such a strong function of temperature.

### *Metal Penetration Problem*

The problem of metal penetration is quite severe in heavy castings, but it can also be a serious problem in small castings. Prior investigators have indicated that under oxidizing conditions the metal might wet the silica grains and facilitate penetration. Previous work[11,13] proved that the formation of a liquid silicate melt, in itself, does not cause a great decrease in the interfacial energy between the liquid metal and solid silica. This confirmed Atterton's analogous results.[7]

However, the liquid silicate melt does attack the mold wall and greatly enlarge the pore size of the silica compact. This will then lead to metal penetration at low pressures. Figure 14, from the phase equilibria studies, clearly shows this mechanism in operation. This sample of an iron-silica couple was treated at 1550 C (2822 F) under oxidizing conditions, 3.9 per cent $CO_2$. The sample size removes pressure as an important variable. The results of the mold wall attack have allowed metal penetration to completely surround many silica grains. Under reducing conditions, a sample of the same size exhibited no reaction or penetration at 1550 C (2822 F).

### *Manganese Additions Problem*

The data involving manganese content of the metal are quite significant. For many years it has been appreciated that manganese greatly aggravates the problem of producing castings with satisfactory surface finish. Experimental results[11,13] explain the reason for this difficulty. The effect of manganese on the stability of a silicate melt is graphically illustrated by the equilibrium data (Fig. 2). Even small additions, e.g., one per cent manganese, greatly lowered the reducing tendency of a given gas mixture.

It is essentially impossible, therefore, under the conditions investigated to prevent the formation of a liquid silicate melt in the presence of manganese. The higher the level of manganese addition to the iron-base alloy, the more severe was the interface reaction. This is verified in practice by the extremely severe mold wall attack of cast Hadfield steels which contain 13 per cent manganese. In view of these results, extensive use of manganese to improve hardenability in cast iron-carbon alloys should be avoided if good surface finish is a prime requirement.

### *Mold Material Effect*

The effects of mold material and the resultant atmosphere are also demonstrated by the data. Practically speaking, it is impossible to attempt to control the mold atmosphere by introducing inert or reducing gas mixtures into a mold cavity. The comparative results of green sand and shell molds, however, hold some promise. In all cases the use of a sand mixture bonded by a polymerized phenolic resin yielded good cast surfaces, little reaction and improved fluidity. This was true at all levels of manganese additions, 0.23-8.60 per cent.

Increasing manganese content had a less severe effect on the interface reaction in the resin bonded

shell molds than in the green sand molds. Actually, the severity of the reaction in the green sand mold was more pronounced at 0.19 per cent manganese than the reaction in the shell mold containing the 8.60 per cent manganese-iron alloy.

The effect of manganese content on interface reaction in a given type of mold is evident from these experimental castings. In both green sand and shell molded castings, the degree of liquid silicate melt formation increased with increasing manganese content.

Microscopic investigation of the sintered layers from these castings provided valuable information. The sintered layers which were removed from the green sand interface of all the test castings contained extensive fayalite formed from the liquid silicate melt. In addition, these reaction zones contained magnetite and some hematite. To produce these higher oxides of iron upon cooling, the atmosphere must be considerably more oxidizing than the range of mixtures used in the equilibration investigations cited in Figs. 1 and 2.

In contrast to this behavior, the reaction zones which were removed from the shell molded castings contained products identical to the equilibration samples previously reported.[11,13] The presence of fayalite confirmed the existence of a liquid silicate melt, but the extensive amount of iron reduced from the silicate melt on cooling indicates the relatively reducing nature of the atmosphere.

Since the same results were produced in a duplex mold with a green sand cope and shell sand drag, control of mold atmosphere at the interface may be possible. Organic sand binders similar to the phenolic resin used in this investigation may be a means of minimizing the undesirable interface reaction.

## CONCLUSIONS

The results of an extensive phase equilibria investigation[13] were evaluated and verified by the pouring of selected step castings. The effect of an oxidizing mold atmosphere was markedly demonstrated in castings produced in green sand molds. The surface finish and interface reaction zone exhibited the pronounced influence of manganese by extensive liquid silicate melt formation over the entire range of manganese additions. The atmosphere produced by phenolic resin bonded shell molds, which oxidizes to a lesser degree than that of the green sand molds, resulted in slight interface reaction over the range of manganese content of 0.23-8.60 per cent. In all cases, the minor attack and the resultant reaction products indicate that the atmosphere was extremely reducing relative to green sand molds.

These results indicate that a practical means of minimizing interface reactions may be possible through the use of organic sand binders which can control the interface atmosphere. In addition, the need for equilibria studies in other refractory oxide systems, and for the evaluation of the effects of other alloy additions to the iron-base melt, is apparent. Thus, by a better understanding of these equilibria combined with a knowledge of the kinetic aspects of reactions during casting, the problem of interface reactions may be controlled.

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## APPENDIX

The calculated carbon composition for these castings was based on the equilibrium diagram of Fig. 1. Since experimental results were available for a temperature of 1550 C (2822 F), this temperature was used to determine the atmosphere which would fix the carbon composition of the iron-carbon alloy. At 1550 C (2822 F), the boundary between liquid iron, silica and liquid silicate melt, silica is fixed at 3.9 per cent $CO_2 - 96.1$ per cent Co. By using available thermodynamic data, the carbon composition in equilibrium with this gas mixture may be calculated as follows:

$$CO_2 + \underline{C} \rightarrow 2\,CO$$

$$K = \frac{[P_{CO}]^2}{[P_{CO_2}]\ [\text{wt. \% C}]}$$

$$\log K = \frac{-\Delta H^\circ}{4.575\ T} + \frac{\Delta S^\circ}{4.575}$$

$$\text{at } 1550\,\text{C} \quad \log K = \frac{-32{,}350}{(4.575)\ (1823)} + \frac{27.70}{4.575}$$

$$\therefore K = 149.9 = \frac{[0.96]^2}{[0.039]\ [\text{wt. \% C}]}$$

$$\therefore \text{wt. \% C} = 0.157$$

The nominal carbon composition used in the charges for the experimental heats was this value of 0.157 per cent C.

# STEEL FOUNDRY SAND CLAY BONDING

## *A microscopic study*

by J. B. Caine, E. H. King and J. S. Schumacher


### ABSTRACT

**Microscopic examination of steel foundry sands casts doubts on accepted theories on "clay film thickness" and "clay distribution" by mulling. Methods for microscopic examination are given.**


### INTRODUCTION

There are many references in the literature to "clay film thickness" and the effect of mulling on "clay distribution." A study of these references reveals that in most, a uniform distribution of the bond clay about the sand grain is assumed but not proved. In mulling studies clay distribution is assumed to be better when green strength increases. Green strength does increase with mulling up to a point, but this is no proof that distribution in itself has changed.

Foundry sand technology abounds with such basic, unproved assumptions that may, or may not be, correct. In an effort to check the two previously mentioned basic assumptions a microscopic study was made. It should be emphasized that the findings reported in this paper merely present the problem. The reader should be warned not to jump to further conclusions that have not been proved. The problem is too important and too basic to compound present confusion by more assumptions not based on fact.

This particular investigation must be restricted to steel foundry sands, for such sands constitute the great majority of the sands studied. However, there seems no reason why the basic findings cannot be extended to all foundry sands, although there will be some differences in detail. The microscopic examination of sand requires different techniques than for metal. A detailed explanation is in order for the proper interpretation of the photomicrographs in this paper.

### INDIVIDUAL SAND GRAINS

Sand grains and their coatings can be studied by oblique illumination, as in Fig. 1. Such illumination is ideal for uncoated grains, but does not show much as to the coating — especially its thickness and distribution. Angular transmitted illumination, as in Fig. 2, gives a shadow picture of the relative amount, thickness and distribution of the bond coating.

Normal transmitted illumination, as is standard with the biological microscope, does not seem as good

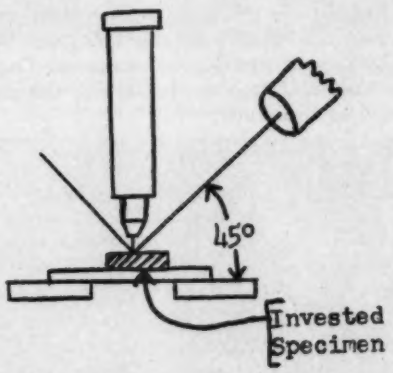



**Fig. 1 — Equipment setup for examining sand specimens under oblique illumination.**

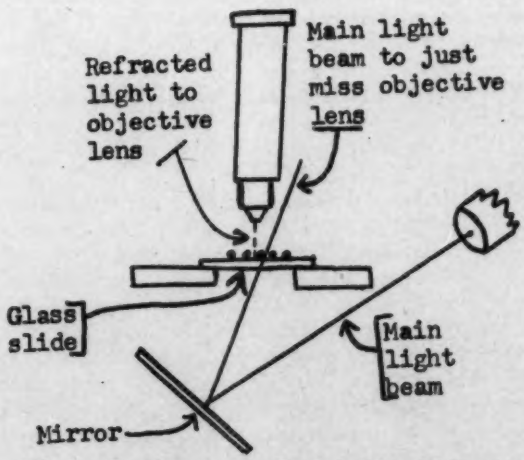



**Fig. 2 — Equipment setup for examining sand specimens under angular transmitted illumination.**

J. B. CAINE is Consultant, E. H. KING is Pres. and J. S. SCHUMACHER is Vice-Pres., The Hill and Griffith Co., Cincinnati, Ohio.

as the angular transmitted illumination of Fig. 2. The main light beam in normal transmitted illumination is focused directly into the objective lens and is unimpeded through the spaces between the loose sand grains blinding the viewer, or causing excessive photographic contrast. A study of loose, individual sand grains is restricted to magnifications less than 30 diameters (30 ×) due to too short a depth of focus at higher magnifications.

## RAMMED SAND COMPACTS

The study of rammed sand compacts requires different preparations than individual sand grains. First, the distribution of the compact must be preserved. This can be done by carefully investing the compact in a thermosetting plastic. Phenol resin core binder rendered opaque with a little lamp black is handy for the foundry, for the materials are available in most core rooms and pattern shops. The resin is set at as low a temperature as possible to prevent bubbles. This usually requires 24 hr at under 200 F. The specimen is then polished just like a metal specimen. Vertical illumination shows little, so oblique illumination is used. Such illumination restricts magnification to about 250 × with a 4 mm objective. Sufficient light is difficult to transmit to objectives of shorter focal length.

The technique used in other fields of examining thin sections by transmitted light has not been used due to lack of equipment for specimen preparation. It is not known if such specimens will tell more than those used in this study.

## MICROSCOPIC STUDY

Figure 3 shows a series of sand photomicrographs used to "zero in" on the study of individual sand grains. Figure 3a shows the sand unbonded, and is typical of many shown in the past. Figure 3b shows the same sand bonded with 6 per cent western bentonite to which a pinch of lampblack was added for photographic contrast under oblique illumination. At first glance the grains seem uniformly coated, but on close examination irregularities can be seen.

Transmitted light, as in Fig. 3c, shows a different picture. The light areas on the sand grains have either no coating or a thin coating of bentonite. Most of the bentonite is concentrated in spots or rings. If Fig. 3c is believed, it is necessary to revise our thinking about clay coatings on sand grains. There is no such thing as "clay film thickness." Instead a nonuniform coating seems to be a basic premise. It must be emphasized that Fig. 3c is not a freak; it is representative of several hundred specimens.

Figure 3d shows the same sand after dry reclamation to 3 per cent AFS clay. The decrease in the amount of coating is apparent. Figure 3e shows the same sand after the standard washing to remove clay. Only the clay that is in the pits remains.

It should be emphasized that the light and dark areas of Fig. 3c and those to follow do not designate areas of clay and no clay. Figure 3b shows that there is a coating over the whole grain. The light and dark areas shown by angular transmitted light only designate areas of varying clay film thickness.

Photomicrographs of this paper, especially Fig. 13, indicate that the dark areas have a clay film thicknesses of about 0.0005 in. The clay film on the light areas cannot be seen on a cross-section at 250 ×, and is probably in the order of 0.00005 in. thick or less. A basic question that cannot be answered at this time is how thick must the clay film be in order to bond two spherical particles.

## MULLING EFFECT

Figures 4 and 5 show sands after two mulling cycles in two types of mullers. Microscopic studies to date show little difference in clay distribution during the usual mulling cycles of the steel foundry. Either slight differences in distribution of the coating on the sand grains are significant, or other explanations are necessary for the changes in properties due to mulling. Mulling is usually evaluated by green strength. It has been the assumption that green strength is related to bond dispersion, and that increased dispersion of the bond results in increased green strength. Figures 4 and 5 cast doubt on this basic assumption.

There is at least one other possible explanation for the variation in mechanical properties of the sand due to mulling. This is – changes in the bond itself. Mulling "plasticizes" all clays by developing strength, adhesion and cohesion. It is just as probable that this plasticizing of the clay is responsible for the increase in green strength, deformation and dry strength, as that these properties are due to an unobservable increase in dispersion of the clay.

## DISCUSSION

If the discussion is restricted for the moment to typical low clay content steel sands, considering only green strength and deformation, mulling to maximum values may be detrimental. Flowability or moldability of any clay bonded sand decreases with increased green strength due to mulling. Only a minimum green strength is required for mechanical strength during molding. Any excess green strength, as distinguished from green strength as a measure of clay content for thermal stability, is a disadvantage.

The same can be said for deformation. Only a minimum deformation is required to prevent cracking and sticking during molding due to misalignment. Unless deformation is a measure of some phase of thermal stability, any excess is a disadvantage in decreasing flowability.

Dry strength is another property that is increased by mulling. Dry strength may be related to certain phases of thermal stability, which is the important property of a molding or core sand. If this is true there are ways to increase dry strength other than mulling, and retain moldability. One method is a slight increase in water content, combined with a shorter mulling cycle.

However, slight variations in water content in themselves do not seem to change clay distribution greatly. A number of sands were mulled in the laboratory with water contents from 2 to 25 per cent with 6 per cent bentonite. There was no increase in dispersion over that of Figs. 3 to 5 as the water was increased from 2 to 15 per cent. A definite increase in dispersion was found with over 20 per cent water, but these sands

were not moldable. They were no longer molding sands, but approached slurries.

## UNIFORM SAND COATING METHODS

Only two methods have been found to date to approach uniform coating of sand grains with clay: high clay content and reuse of the sand in the foundry. Figure 6 shows that 12 to 18 per cent western bentonite almost covers the sand grains. Figure 6c shows about the same amount of coating with Ohio fireclay. Reuse of sand is restricted in the steel foundry due to danger of scabbing, but the point is interesting. Figure 7 shows the sand from one steel foundry using a minimum of new sand. An appreciable number of grains are completely covered. Figure 8a shows a

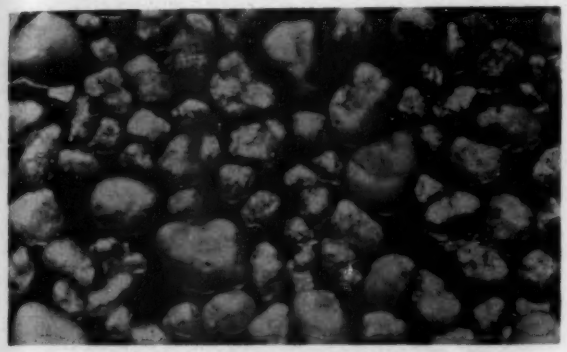


Fig. 3a — Washed and unbonded Ottawa type sand. Oblique illumination. 25 ×.

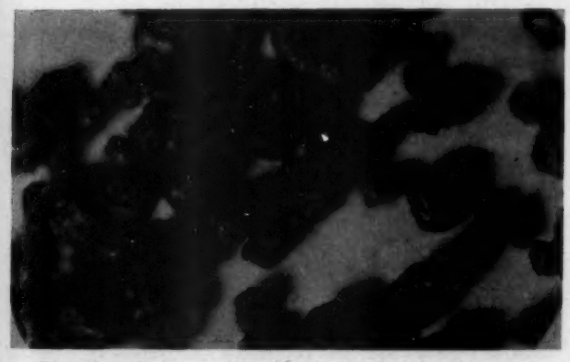


Fig. 3b — Same sand as in Fig. 3a, but bonded with 6 per cent western bentonite. Oblique illumination. 25 ×.

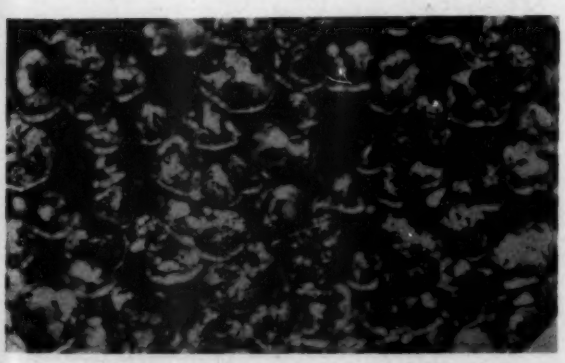


Fig. 3c — Same sand as in Fig. 3b, but using angular transmitted illumination. 25 ×.

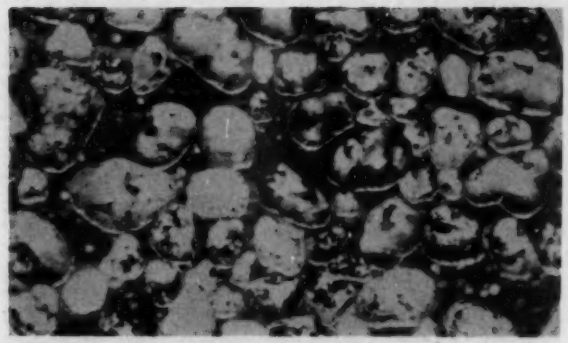


Fig. 3d — Same sand as in Figs. 3a and 3b, but after dry reclamation to 3 per cent AFS clay. Angular transmitted illumination. 25 ×.

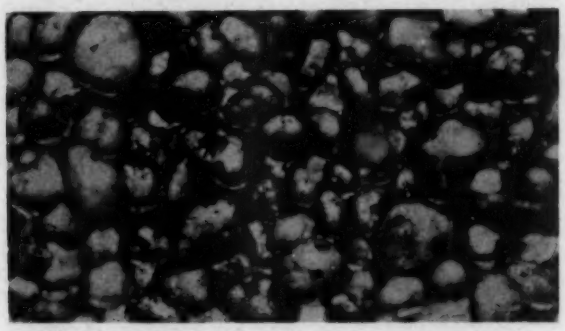


Fig. 3e — Same sand as in Figs. 3a, 3b and 3c, but after the clay has been removed by the AFS washing method. Angular transmitted illumination. 25 ×.

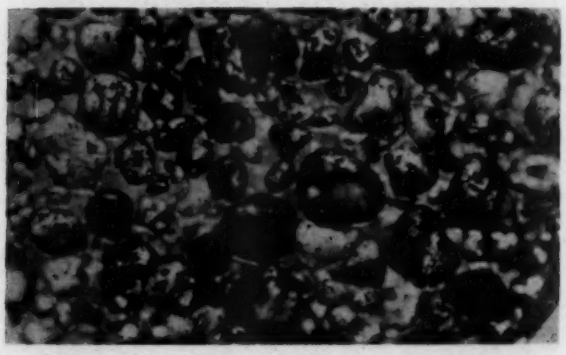


Fig. 4a — Sand with 6 per cent western bentonite added, mulled one min in a slow speed muller. Angular transmitted illumination. 25 ×.

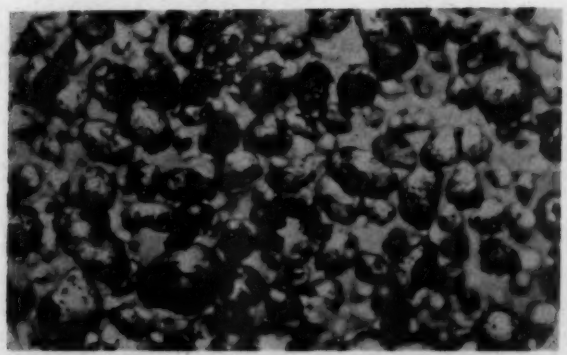


Fig. 4b — Same sand as in Fig. 4a, mulled 6 min in a slow speed muller. Angular transmitted illumination. 25 ×.

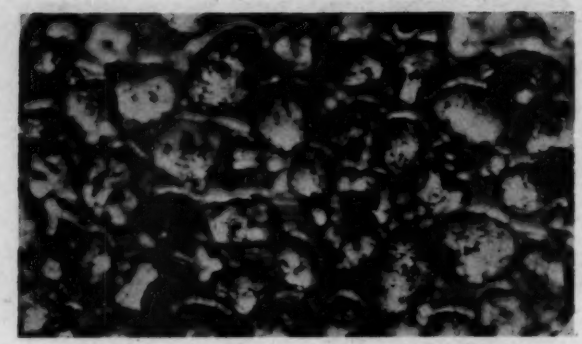


Fig. 5a — Sand with 6 per cent western bentonite added, mulled 20 sec in a high speed muller. Angular transmitted illumination. 25 ×.

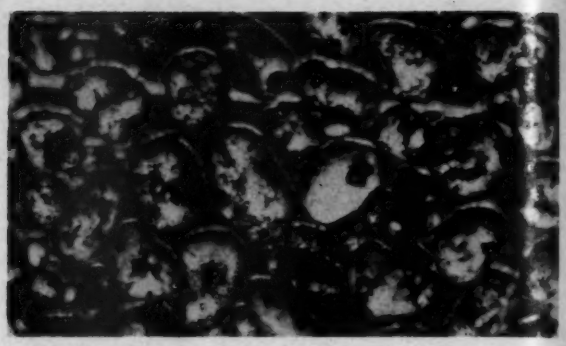


Fig. 5b — Same sand as in Fig. 5a, mulled 60 sec in a high speed muller. Angular transmitted illumination. 25 ×.

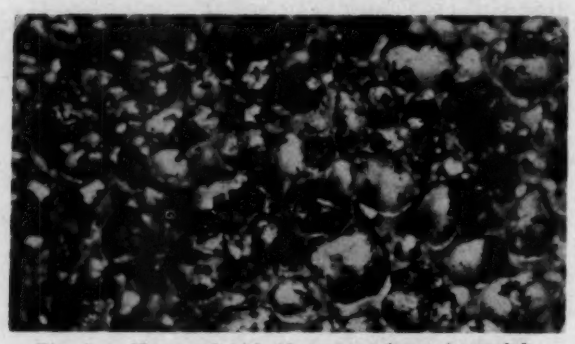


Fig. 6a — New sand with 12 per cent bentonite and 8 per cent water. Angular transmitted illumination. 25 ×.

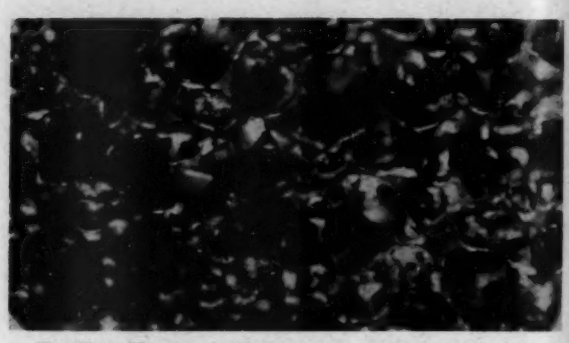


Fig. 6b — New sand, 18 per cent western bentonite, 6 per cent water. Angular transmitted illumination. 25 ×.

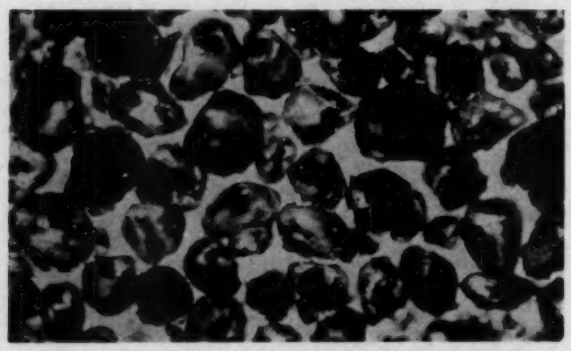


Fig. 6c — New sand, 18 per cent Ohio fireclay, 6 per cent water. Angular transmitted illumination. 25 ×.

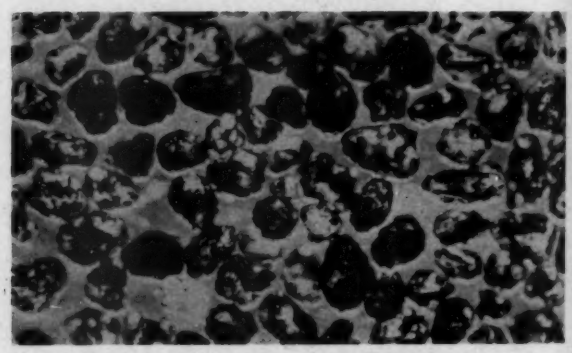


Fig. 7 — Steel system sand, 8 psi green strength, 10 per cent AFS clay. Angular transmitted illumination. 25 ×.

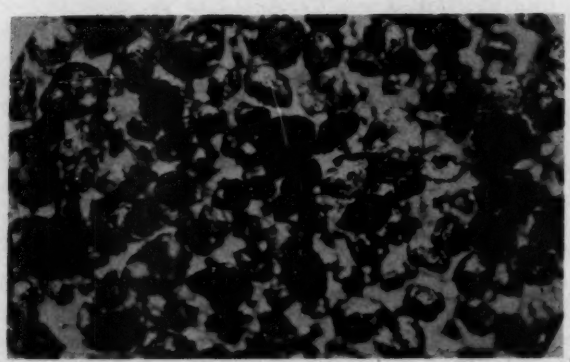


Fig. 8a — Iron system sand, 26 psi green strength, 17 per cent AFS clay, 5.1 per cent loss on ignition. Angular transmitted illumination. 25 ×.

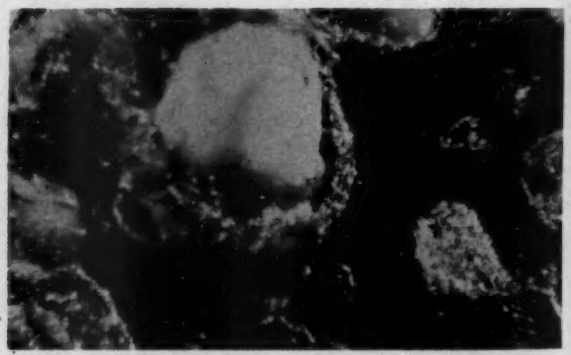


Fig. 8b — Same sand as in Fig. 8a, invested and polished. Oblique illumination. 125 ×.

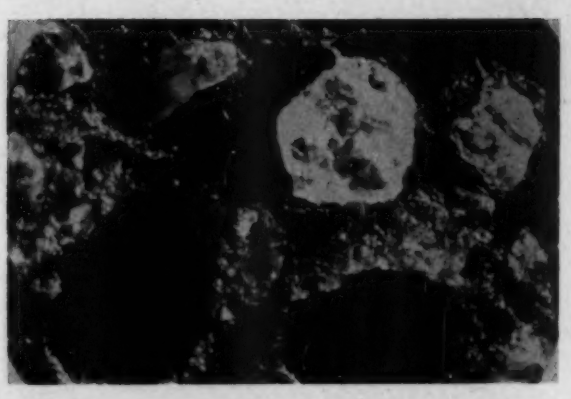


Fig. 9 — New steel core sand containing 25 per cent silica flour, invested and polished. Oblique illumination. 125 ×.

system sand used in an iron foundry whose grains are almost completely covered.

The assumption should not be made at present that it is desirable to have uniformly covered sand grains in the steel foundry. Other factors must be taken into consideration. For example, the steel sand of Fig. 7 does not make particularly good castings, in fact the castings could be classed as under average. Figure 8b shows the iron sand of Fig. 8a after investing with plastic, polishing and examining at higher magnification. The loose shell around one sand grain is evident. This sand makes excellent iron castings, but such loosely bonded sand grains can be suspected as one reason steel sands fail on continued reuse without reclamation.

Figure 8b introduces the second microscopic technique for sands, that of the study of rammed sand compacts, invested with a mounting medium and at relatively high magnification. A fundamental point here is the identification of the phases. There is little doubt of the silica grain of Fig. 8b, and that this silica grain is surrounded by a relatively thick coating of another phase. It seems safe to assume that the surrounding phase is nongranular in the sense that the silica grain is granular.

The main nongranular materials present in this sand are western bentonite and sea coal from the additions. This bentonite and sea coal would be dehydrated and charred from heating during reuse. The "coky" appearance of the shell of Fig. 8b is typical of reused sands, which is different than the appearance of fine silica particles, as the silica flour particles of Fig. 9.

## HIGH MAGNIFICATION TESTS

Figure 10 is shown as the basis of comparison for the evaluation of bond coatings at high magnification on unbonded, white silica sand. This specimen was vibrated in a test tube to 61 per cent density, invested, solidified and the test tube broken away from the specimen. Note the chains of white refraction dots around the sand grains. These dots, at present,

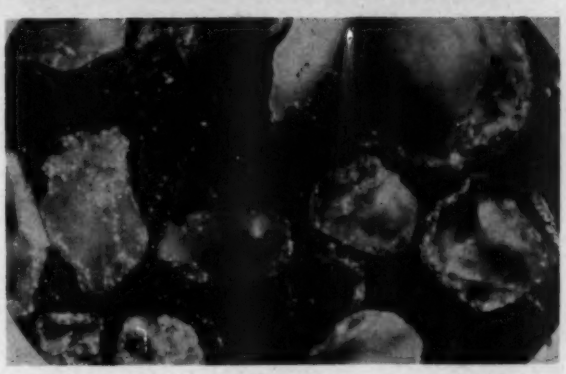


Fig. 10 — Washed, unbonded sand, invested and polished. Oblique illumination. 125 ×.

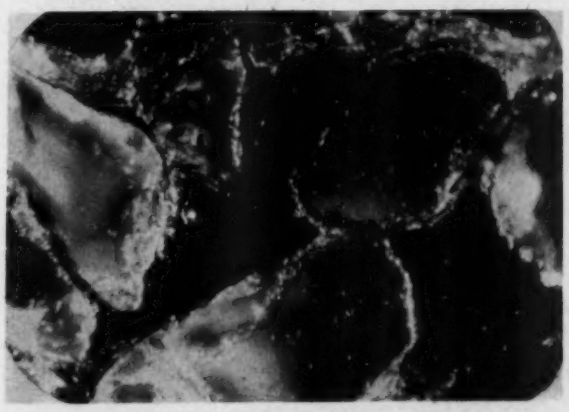


Fig. 11 — Surface coated with western bentonite slurry to distinguish bentonite, invested and polished. Oblique illumination. 125 ×.



Fig. 12 — Coating obtained with 18 per cent Ohio fireclay, invested and polished. Oblique illumination. 125 ×.



Fig. 13a — Production steel facing sand, 8 per cent bentonite. Clay areas 0.00056 in. thick. 125 ×.

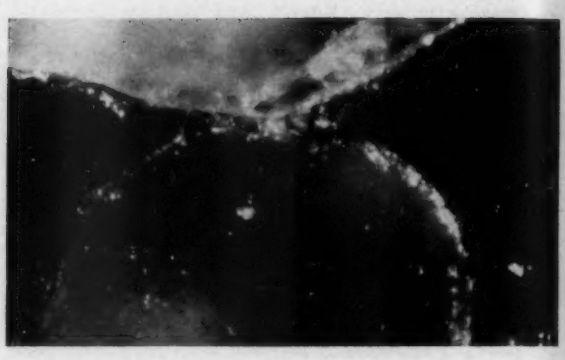


Fig. 13b — Production steel facing sand with 8 per cent bentonite. Clay areas 0.0005 in. thick. 250 ×.

interfere with the evaluation of small amounts of bond coatings. The white refraction dots of Fig. 10 must sometimes be used as sort of a blank in the evaluation of thin bond coatings.

Figure 11 is an attempt to establish the appearance of western bentonite at relatively high magnification. It shows a coating of western bentonite along the top, horizontal surface made by dipping a sand specimen in a bentonite slurry. The bentonite appears "pearly" under visual examination. A normal bonded contact between two sand grains is seen in the right center of Fig. 11. Figure 12 illustrates a relatively thick bond coating, formed in this instance by 18 per cent Ohio fireclay. The fireclay does not seem as translucent as western bentonite, but the difference is slight.

Figure 13 illustrates the appearance of what is now considered normal bentonite coatings, 6 to 8 per cent, at successively higher magnification. These are production steel facing sands. Note the bond at the grain contacts. Figure 13b shows bond at the grain contact, a gap and a spot of bentonite away from the contact. It should be emphasized that areas showing bond contacts are not easy to find under the microscope.

This may, or may not be as it should. First, the examination of a single plane through a three dimensional mass cannot be used, except statistically, as a criterion of the mass. Perhaps particles that do not seem bonded on a single plane are bonded on a different plane. Then, too, the nonuniform coating makes it improbable that every contact point will be bonded.

This last point indicates that a uniform coating may be an advantage. This is probable, except for one point – there is the possibility of two grains sliding against their unbonded surfaces during ramming until they hit a spot of bond. If this action could be relied upon it would enhance flowability and moldability without sacrifice of bonding. These points can be studied and evaluated quantitatively, if this is done statistically. However, the study quickly becomes complicated to the point of the order-disorder mechanics of atomic physics.

## SUMMARY

It was stated in the introduction that the findings of this paper merely present the question or problem. The results to date can be summarized as follows:

1) The theories of uniform distribution of clay substance on sand grains and clay film thickness, within the ranges of clay content and water used in the steel foundry, are possibly in error. Perhaps all that can be hoped for with present bonding techniques is a varying degree of nonuniformity.
2) Mulling over short intervals of time does not increase clay distribution as evaluated by this paper. The increase in green and dry strengths with increased mulling may not be due primarily to improved distribution of the bond. The increase in green strength may be detrimental in that moldability is decreased. The role of dry strength in steel sands is not clearly understood.
3) Techniques of specimen preparation for microscopic examination are given. Preliminary qualitative methods for differentiation of the various bond constituents are given.

# LIGHTWEIGHT CELLULAR METAL

by L. Polonsky, S. Lipson and H. Markus


## ABSTRACT

**A process for the production of cellular metals is described. The process consists of preparing a refractory mold, filling the mold with soluble granules which correspond to the size and shape of the pores desired in the metal and infiltrating the molten metal into this soluble aggregate. After leaching with a suitable solvent, the metal has a void structure of interconnecting cells which correspond to the shape and size of the soluble aggregate. The apparent density of the metal body can be controlled from a maximum of approximately one-third of its base density to some lower density in the order of one-fifth the base density.**

**In addition to process development, which includes casting of shapes as well as billets, a limited investigation of the mechanical characteristics of aluminum-base cellular metal was made. Potential applications for materials of this type are suggested.**


## INTRODUCTION

The development of foamed plastics and their application as materials of construction have lent a certain impetus to the development of similar metal base material. The first successful attempt to produce an aluminum foam was the result of an Air Force contract with the Bjorksten Research Laboratories, Inc.[1,2] The method used was similar to that employed for plastic foams in that a gas forming component was added to the molten aluminum alloy.

The first efforts were with the aluminum-magnesium eutectic composition (55 per cent Mg-45 per cent Al) with either zirconium or titanium hydride dispersed through the melt as the foaming agent. This produced a metallic foam which had a specific gravity of 0.5 to 0.6 (gm/cc) with a void volume of approximately 75 per cent of the specimen volume. Another process for the production of porous metal products was developed by Armour Research Foundation.[3] This is called film metallurgy, and the material, which is produced by a felting process, is sintered in much the same manner as is done with metal powder products.

While these newly developed materials had many desirable characteristics, they suffered from certain limitations insofar as the specific Ordnance applications which were then being considered for materials of this type. In an effort to develop a material more nearly suited to these specific applications, a radical departure was made from the foaming principle. The material produced as a result of this new process was called cellular aluminum rather than foamed aluminum, since foaming played no part in the method of production.

Briefly, in this process use is made of a refractory mold which is filled with graded soluble granules. The alloy is infiltrated throughout the compact by methods which are discussed later in this report. After solidification, the soluble granules are leached out of the structure, leaving a cellular metal structure which has a void volume of approximately 70 per cent.

Preliminary tests showed that this method was feasible. The interconnecting nature of the voids was an interesting characteristic, and suggested a variety of other possible applications as an ordnance material. This report is a description of processing techniques developed to produce cellular aluminum and other metals, as well as a limited evaluation of mechanical and physical properties of the material.

## MATERIALS AND METHODS

### *Description of General Process*

Cellular metal is produced by a foundry process. Molds are prepared according to standard practice. Once prepared, the molds are filled with a suitable grade of soluble granules. Since substantially all of the work described in this report deals with metals having casting temperatures no higher than that of aluminum, the soluble granules used were graded crushed rock salt. In filling the mold, the particles tend to pack in the most effective manner permitted by the particle shape.

Vibration is used to promote this packing. Since none of the particles can remain suspended in space, each particle must contact at least one other neighboring particle. In practice, all of the particles touch many adjoining particles.

The mold is then heated in preparation for infiltration by the molten metal. During the heating cycle mold moisture is vaporized and small amounts are condensed on the surface of the granules. This provides a liquid binder for the particles. Upon

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further heating, the liquid is evaporated, and the salt, crystallizing from solution, welds the particles together at the points of junction. The mold cavity, at this point, contains a coherent salt briquette. The spaces between the particles are eventually filled with the metal which forms the cellular structure. Figure 1 shows part of a salt briquette which was removed from a mold after it had been heated to the casting temperature.



Fig. 1 — Salt compact.

The metal is now melted, brought up to its pouring temperature, and poured into the salt-filled mold. By methods which are described later, the molten metal is infiltrated into the spaces between the salt particles and, once in place, is permitted to solidify. The casting may now be shaken out of the mold and is ready for any necessary machining. The casting is a composite material, one component being the salt briquette and the other the metal infiltrant. The cellular metal structure is exposed by leaching away the salt crystals from the metal-salt composite.

The product may be regarded now as a negative of the porous type structure obtained by powder metallurgy. The metal component may be compared to the void component of the powder metal compact, and, conversely, the void component to the metal component of the powder metal product.

It may be seen that this process permits control not only of void or cell size, but also of cell geometry and size distribution. Figure 2 shows a range of cell sizes produced by this process. The details of accomplishing the steps outlined above are described in this report.

### *Soluble Granules*

Ideally, the soluble material used for making cellular metal should be a high-melting point salt of high-aqueous solubility, noncorrosive to the metal infiltrant, stable, nonhygroscopic, readily available, and inexpensive. Since this investigation primarily pertained to the production of aluminum alloys in cellular form, a salt had to be selected to fulfill the requirements as related to these alloys. A review of potentially suitable materials quickly disclosed that sodium chloride fully satisfied most of these requirements.

In fact, the only serious limitation was the corrosive nature of sodium chloride brines. Although reference data revealed several other promising salts, on the basis of the criteria mentioned above sodium chloride was selected as it was far superior to all those considered.

The salt used in this work was a commercial grade which is 99 + per cent sodium chloride with insolubles limited to negligible percentages. It was obtained in moderately coarse lump form, being predominately 4 to 8 mesh. These lumps were further reduced by crushing and the crushed salt was graded according to particle size. Crushing was accomplished by a standard reciprocating jaw crusher with adjustable jaw clearance.

Figure 3 shows samples of a variety of crushed salt crystals, as well as a sample of spherical particles which was produced by solidification of molten droplets of salt.

## MOLDING

Mold materials for making cellular aluminum castings must be permeable in order to provide an exit for the air contained in the mold, yet must be resistant to metal penetration. They must be strong up to temperatures as high as 1400 F (760 C) and inert to both salt and metal. These requirements are fulfilled to a greater or lesser degree by silicate-bonded and plaster-bonded silica molds. Other types of molding media may be modified to satisfy these requirements.

The process requires special flasks to encase the molds. These flasks must be able to withstand the high mold temperatures, and permit water cooling of the flask in order to induce the thermal gradients necessary for solidification control. Since the type of flask needed is determined by factors other than molding, these requirements are discussed in the section dealing with casting and solidification.

Molding problems encountered in cellular aluminum casting production fall within two categories —those problems associated with casting of billets and those dealing with shaped castings.

### *Billet Castings*

Billet castings are produced in conforming steel flasks. Usually, no pattern is needed and the mold walls are protected by a refractory coating approx-

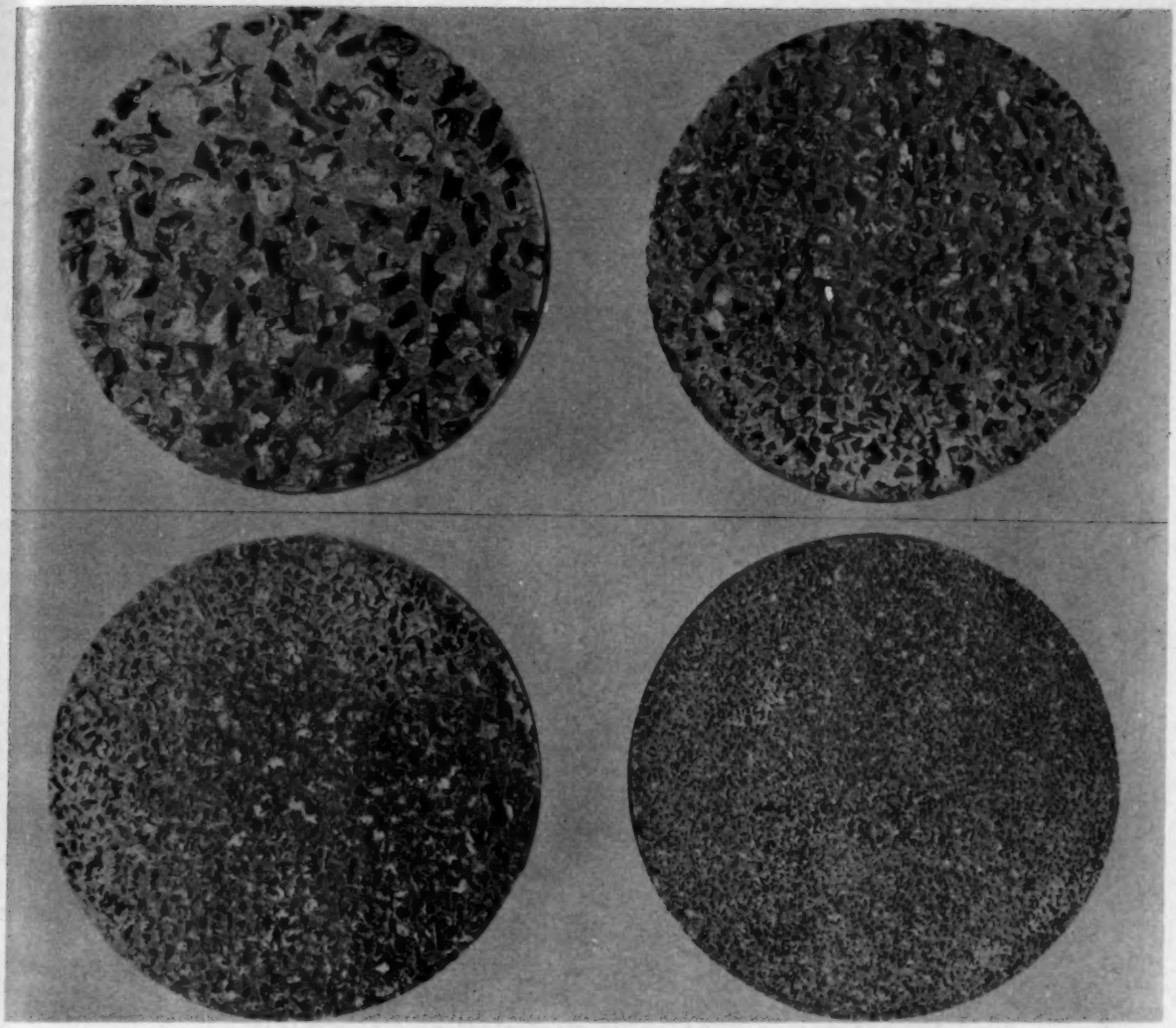


**Fig. 2 — Cellular 356 aluminum alloy discs showing range of cell sizes. 2 ×. *Left to right — top* — 2.36 mm and 1.65 mm, *bottom* — 0.83 mm and 0.41 mm.**

imately ¼-in. thick. A suitable material for this refractory coating is a metal casting plaster of the type used for investment molding. The coating is air dried over night, and then the mold may be filled with the graded salt crystals. The filling operation is facilitated by the use of vibration, which decreases the percentage of voids in the aggregate.

Space must be provided at the top of the mold for the feed metal necessary to compensate for solidification shrinkage. The feed metal requirements are discussed in greater detail in the section dealing with casting and solidification.

The billet flasks may also be lined with asbestos paper or with silicate-bonded sand rammed against a pattern. The rammed mold is less satisfactory than either the plaster wash or the asbestos paper lining. This is due to the greater thickness of the rammed lining, which reduces the rate of heat transfer through the flask wall.

### *Shaped Castings*

Either plaster-bonded or silicate-bonded molds are satisfactory for shaped castings. Here again it is desirable to use flasks of approximately conforming shape in order to control solidification. The casting technique for infiltration of the salt granules is so drastic that it is difficult to perfectly seal a parting line against leakage. For this reason molds are prepared without flasks and are then bedded down in a one piece flask. Silicate-bonded sand is used for this purpose.

A number of shaped castings were made in silicate-bonded sand molds. In order to reduce the tendency for metal penetration, a special facing sand was prepared.

| Material | Weight, % |
|---|---|
| Silica sand (140 mesh) | 75 |
| Silica flour | 20 |
| Silicate binder | 5 |

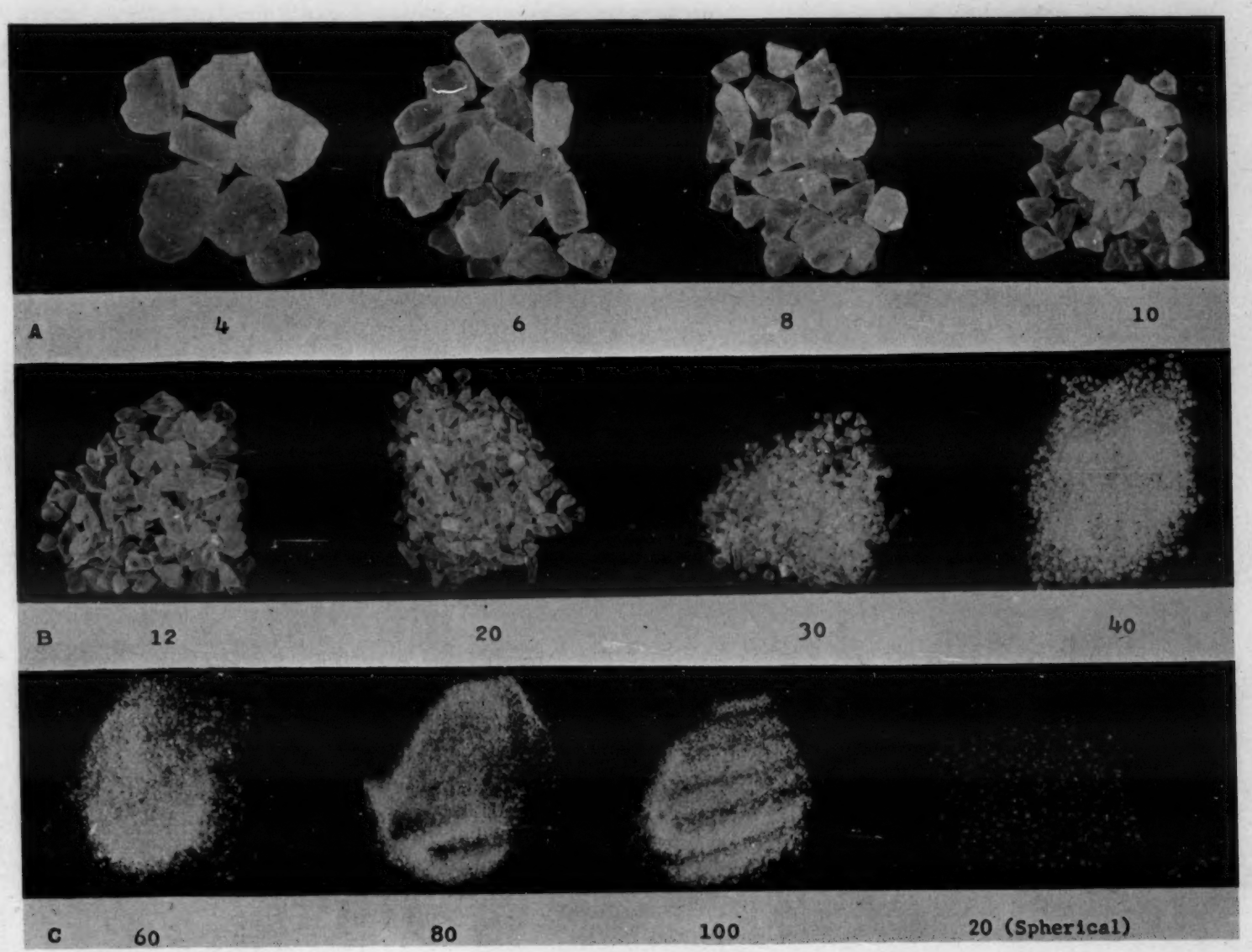



Fig. 3 — Salt particle mesh sizes. 2 ×.

The backup was a silicate-bonded sand of approximately 80 AFS fineness. Before filling and closing, the mold surface may be sprayed with silicate binder as further insurance against metal penetration.

## CASTING AND SOLIDIFICATION

In order to infiltrate a salt aggregate, the molten metal must follow a tortuous path between the salt granules. It must overcome the surface tension of the liquid metal and the friction of the channels through which it flows. It must force the air contained in these channels through the mold walls, which are of limited permeability.

Another factor that must be considered is that even though the product of this process is a cellular metal skeleton, it is essential that the fundamental principles common to all solidification processes be observed. Unfed solidification shrinkage results in spongy cell walls, which impair the quality of the cellular material in the same way that spongy metal impairs the quality of a conventional casting.

To pour a cellular casting successfully, three basic requirements must be fulfilled. While all of these are inter-related, they are discussed separately.

### *Mechanical Factors*

Certain elements of the process are mechanical, since they pertain to forcing the liquid metal to infiltrate the granular compact. These elements determine the kind of equipment and methods necessary to produce successful infiltration. It was found that molds could be successfully infiltrated by gravity pouring vibration and pressure pouring. Gravity pouring can be used with coarse grades. However, more drastic methods are required to infiltrate with progressively finer granules.

Gravity pouring has been found satisfactory for pouring large billets of coarse cell structure. A tall sprue is necessary to provide the metallostatic head to develop the pressure needed to overcome surface tension and frictional effects. With sufficient metallostatic head to initiate entry of the metal into the compact, the lengthening head caused by infiltration into the compact keeps the process in operation until infiltration is complete. This method is limited to mesh sizes of No. 6 and coarser.

Vibration is useful in extending the size range of cell structures that can be infiltrated without recourse to pressure techniques. Its mechanism is apparently related to the cyclic pressures which overcome the surface tension of the liquid metal, permitting it to enter the narrow channels between the salt crystals. The method extends the range of granule sizes that can be infiltrated down to approximately No. 12 mesh. Figure 4 shows a mold clamped to a vibrator. It is essential that the mold be firmly fixed to the vibrator in order that the energy be effectively transmitted to the mold.

Pressure pouring permits infiltration of extremely fine aggregates. Castings have been made with granules as fine as 100 mesh. However, with smaller granule sizes, the possibility for entrapment of air becomes greater. An effective technique is to employ suction at the base of the mold simultaneously with the application of pressure. The suction removes the air contained in the mold. It also increases the effective pressure on the metal by the increment to which it increases the pressure differential between mold surface and metal head surface. Figure 5 is a schematic illustration of a suction-pressure apparatus used for much of this work.

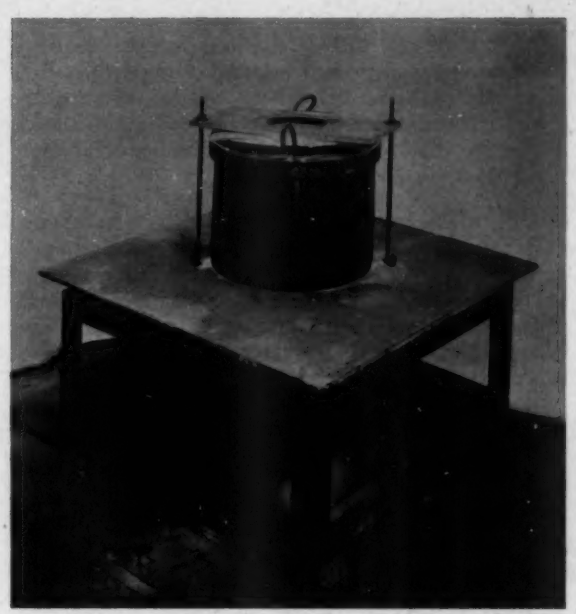


**Fig. 4 — Vibration casting fixture with mold fixed in position.**

A clamp was used to secure the mold directly to the vacuum box. The asbestos gaskets shown are for the purpose of preventing air leaks. Perforations on the top of the box provide a passageway through which air is exhausted from the mold, the refractory bottom of the mold being permeable to gasses but not to liquid metal. Air pressure on the head was controlled by a pressure regulator. It should be noted that the volume of metal in the sprue must be enough to fill the voids within the aggregate and, also, to provide a reservoir for solidification shrinkage.

### *Thermal Factors*

In order for the metal to infiltrate completely the salt compact, the duration of its fluid life (time during which the metal remains liquid) must be adequate. The fluid life is a function of the amount of superheat put into the metal and the rate of heat transfer to the mold walls and salt granules. Since the aggregate surface area of the granules is large, and since they occupy a volume of approximately twice that of the metal, it is apparent that the loss of heat from metal to salt must be kept low to fill the mold with metal.

This can be done by narrowing the temperature differential between the mold and the metal. Experience shows that it is necessary to preheat the mold close to the melting range of the alloy in order to avoid excessive amounts of superheat to

sustain the fluid life of the molten metal. If infiltration were the only consideration, the required mold temperature would encompass a fairly narrow band and be inversely related to the mesh of the granular compact. However, controlled solidification of the metal introduces other requirements for mold temperature.

### *Solidification Factors*

In conventional foundry molds, solidification is controlled by means of thermal gradients developed in the solidifying casting. This is achieved by taking advantage of casting geometry, gating, and riser placement, as well as metal and mold temperatures. Since cellular metal castings must be produced in molds which are at nearly the same temperature as the molten metal, only limited thermal gradients can be anticipated. On the basis of existing knowledge of solidification, these would be inadequate to promote feeding of the solidifying casting.

In order to produce well-fed cellular metal castings, it is necessary that the casting be poured and solidified as two separate and distinct steps in the process. To obtain sound metal it is neither necessary nor desirable that the mold temperature be kept at a minimum value consistent with successful infiltration. It is important that, immediately after casting, the metal and mold be at a temperature higher than the melting range of the alloy. Thermal gradients may now be produced in the casting by cooling the lower end of the mold or flask surface.

To insure feeding of the casting, the mold and metal temperature just after pouring must be high enough so that an adequate thermal gradient can be established between the chilled area at the bottom of the mold and the feed metal at the top. The exact temperatures selected are dependent upon the length and volume of the casting to be solidified. Naturally, it is desirable to avoid temperatures in excess of those required to generate the necessary thermal gradients. Experience has indicated that it is necessary to develop thermal gradients of not less than 5 F/in.

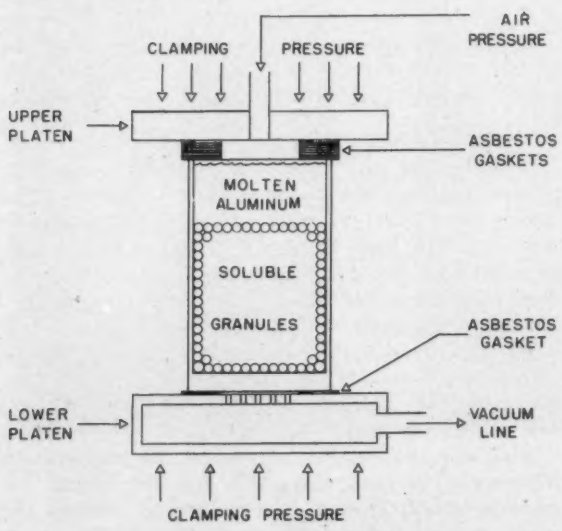



**Fig. 5 — Suction-pressure casting apparatus.**

### *Water Cooled Mold*

Ideally, the mold should be cooled by means of water circulating through a water jacket at the base of the mold. It has been found, however, that adequate progressive solidification can be obtained either by use of an open trough to contain the water or by spraying water against the mold surfaces. It is not necessary to circulate the water in the trough, since water temperature may be regarded as a negligible factor compared to the heat sink provided by the water boiling in the trough. This type of cooling should be applied over the lower 25 to 50 per cent of the surface.

It is necessary, however, to avoid wetting the refractory in the mold. This is done by either using a flask which is sealed at the bottom or designing the water trough in such a way that spillage due to the boiling is kept to a minimum.

Figure 6 is a cross-sectional drawing showing the design of a flask having a cooling trough incorporated in it. A false bottom is provided in order to permit the use of suction as an aid to infiltration. The bottom is supported either by refractory brick or a bed of coarse sand. For nonferrous metals, the bottom plate may be a wire mesh reinforced plaster disc which rests on the support. If a shaped casting is being made, the mold containing the soluble granules is placed in the flask and bedded down with sand. If a billet is being made, the granular aggregate is simply poured into the flask which has previously been lined with a suitable refractory.

After the mold has been heated to the proper temperature for casting it is withdrawn from the furnace, the suction pipe is connected to a vacuum line and the metal is poured into the mold. For fine mesh sizes it may be necessary to employ a small positive pressure, as shown in Figure 5. In any event, the top of the mold should be covered with a transite sheet or asbestos paper to reduce the heat loss from the surface and guard against splashing of water into the molten metal.

Sufficient water is now poured into the trough to start the solidification process. The exact level required is based upon previous experience with the same casting. The water begins to boil almost immediately, and it is the heat of vaporization that produces the heat sink necessary to absorb the large amount of heat contained in the mold. Water is added from time to time to make up for evaporation losses.

## LEACHING

The removal of salt from the aluminum-salt composite is accomplished by leaching with water. This process may be regarded as consisting of two steps:

1) the solution of all the salt in the cells.
2) the rinsing or removal of all the brine formed by solution of the salt.

It is futile to do the rinsing unless the solution process is complete.

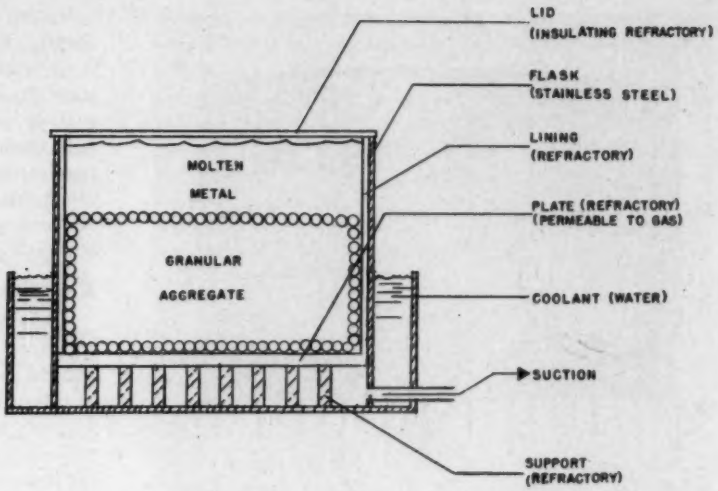



Fig. 6 — Mold with water cooling trough for solidifying cellular metal castings.

*Solution Process*

The leaching process as applied to these composites proceeds for the most part as a diffusion process. The leaching liquor in contact with the salt rapidly approaches saturation and the salt ions diffuse back toward the liquor of lower concentration. Due to the nature of the metallic structure, agitation of the liquid results in little benefit because of the confining nature of the cells. The rate at which salt goes into solution is affected by the area of contact and the length and geometry of the path through which the ions must diffuse.

Figure 7a is a model of a string of horizontal cells. The first cell in the string ($C_1$) is rapidly freed of salt because of its proximity to the dilute leaching liquor. Before the salt in $C_2$ can be removed, the salt in the passage between $C_1$ and $C_2$ must be leached and the ions must diffuse through this passage and out into the main body of liquor. As the solution process proceeds, the liquid column for diffusion becomes longer and longer, with more and more restricted passages to retard the diffusion process.

In addition to the diffusion process, the denser liquid at the exit end of the passage gravitates to the bottom of the leaching tank, which tends to maintain a lower concentration solution near the specimen. Figure 7b shows concentration gradients which develop in the tank during the leaching process.

Where the cell structure is coarse or the specimen is thin, leaching can be accomplished rapidly because the number of restricted passages to the surface are fewer. In addition, the currents set up by the movement of dense liquid toward the bottom act not only on the surface of the specimen, but through it. Where the cell structure is fine, the chief mechanism for solution of the crystals contained in the metal cell is the diffusion process previously described.

*Peripheral Leaching*

Figure 8 shows a section of a partially leached specimen, showing the peripheral mode of leaching. This specimen was suspended in the tank so that the right side was closest to the surface. This is the location which benefits least from the concentration currents in the tank and, as a result, shows the shallowest depth of salt removal.

If the liquid used for leaching contains some gas, either in solution or in dispersion, some of this gas may collect in the cells and form an air lock which will arrest the diffusion process. Consequently, the leaching through the blocked channel is

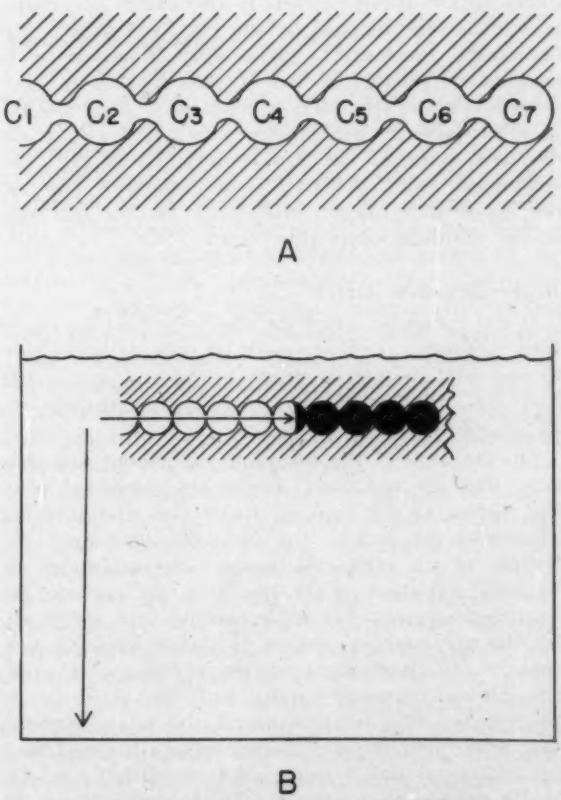



Fig. 7 — *A* — A series of cells in cellular metal. *B* — Concentration gradients developed in the tank during leaching of cellular metal.

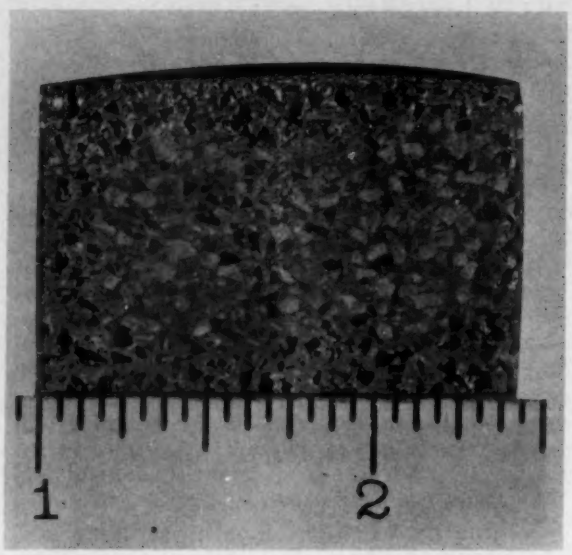



**Fig. 8 — Cross-section of partially leached cellular 356 aluminum alloy.**

halted. For this reason, the water used for leaching should be conditioned either by standing prior to use or, preferably, deaerated by vacuum technique.

Gas is sometimes formed as a product of the chemical action between the brine and the alloy being leached. This not only results in undesirable corrosion of the specimen, but it also causes the channels to become blocked by the gas lock. When this occurs the gas must be periodically removed by a vacuum outgassing of the structure. This is best done while the specimen is submerged in the tank. If a partially leached specimen is withdrawn from the tank, some of the liquid contained in the cells will drain away. When it is resubmerged, any air not replaced by liquid will block further leaching in the channels where this occurs.

### *Brine Corrosive Attack*

If the corrosive attack by the brine on the specimen is severe, it is necessary to provide some protection for the metal either by adding a suitable corrosion inhibitor or by providing electrolytic cathodic protection.

On the basis of the foregoing discussion, it is evident that the specimen should be suspended near the surface of the tank in order that the currents caused by the flow of the dense liquor toward the bottom of the tank can be used to maximum advantage. Agitation of the liquor in the tank would disturb the natural favorable currents and, therefore, impede the leaching process. If a tank having a volume of ten times the specimen volume were used, the salt concentration would reach one third of saturation for that temperature if the salt concentration were regarded as uniform through the tank and no additional liquid were added or withdrawn. Actually, the liquid in contact with the specimen would contain considerably less salt than equilibrium conditions would indicate.

If the liquid is slowly drawn from the bottom during the leaching operation and replaced with clear water at the top, only three changes would insure that the brine concentration at the specimen surface would never exceed one or 2 per cent. This is 3 to 5 per cent of saturation. Reducing this salt concentration further would only have a negligible effect on the rate of leaching. This is because the concentration gradient would not be significantly affected.

### *Rinsing*

After all the salt in the composite has gone into solution, the specimen is ready for rinsing. This should be done under a heavy flow of water, preferably through the material. After rinsing, a small amount of drippings from the specimen should be collected and checked for chloride ion. If the concentration of chloride ion in the drippings is no greater than that of the rinse water, the material has been properly leached and rinsed.

If the chloride ion concentration remains high after two or three attempts at rinsing the leaching process is not complete, and the material should be returned to the tank for further leaching. Any air entrapped in the structure must be removed after the material is submerged. By this technique, a leaching cycle is established for each material being produced. Additional pieces can be successfully leached by adhering to the established cycle.

All of the castings produced in connection with this investigation were successfully leached by the methods already described. Certain obvious improvements in leaching technique, however, were not investigated because they had no bearing on the main purpose of this study. One of these potential improvements is the use of high frequency, low amplitude vibration, such as is commonly used in electrolytic cleaning processes. This would serve the dual purpose of accelerating the rate of diffusion of the solute in the liquor of high salt concentration toward the surface of the specimen, and dislodging gas bubbles in the cell cavities which tend to block off certain cavities from action by the solvent.

## CELLULAR METALS OTHER THAN ALUMINUM

The methods described can be used with appropriate modifications to produce cellular metals other than aluminum. For metals and alloys having melting points below that of the sodium chloride, the procedures followed are substantially identical to that described for aluminum except that mold temperatures and casting temperatures are appropriately adjusted. The table presents the mold and casting temperatures used to produce cellular structures in some metals other than aluminum.

| Metal | Melting Point, F | Temperature, F | |
|---|---|---|---|
| | | Mold | Casting |
| Pb | 621 | 665 | 750 |
| Sn | 459 | 500 | 600 |
| Zn | 787 | 840 | 920 |

Certain metals, however, while falling within the melting range given above, must be treated differently because of inherent difference in their characteristics. A case in point is that of magnesium and its alloys. Because of their reactive nature it was found that these metals must be cast in molds which have been properly inhibited. One satisfactory mold medium tested was plaster bonded silica similar to that used for investment casting except that one per cent boric acid was added to the mix as an inhibitor. The castings were produced in a suction-pressure apparatus, as previously described.

Just before the metal was poured, a small amount of sulfur was dusted over the salt compact. This produced an $SO_2$ atmosphere in the mold, and completely inhibited burning when the metal was poured through it. Pressure to produce infiltration was applied, using argon gas at 10 to 20 psi. The largest casting produced by this technique was a 3 in. diameter billet. For this small casting it was possible to generate thermal gradients in the solidifying casting by cooling the lower extremity of the mold by the application of damp rags to the flask surface. For larger castings it would be necessary to have a water jacketed flask, since an open water spray for cooling would be too much of a hazard.

Leaching of the magnesium salt composite proved to be considerably more difficult than leaching aluminum. This was apparently due to the reaction of magnesium to water, which was accelerated by the presence of chloride ion.

It was possible to leach thin slices of this material without excessively attacking the magnesium. This was done under a heavy flow of water which decreased the chloride ion concentration. It was apparent, however, that it would be necessary to find satisfactory corrosion inhibitors or, perhaps, means for providing cathodic protection of the metal if heavy sections were to be satisfactorily leached.

Several cellular ferrous castings were also produced. One of these was an austenitic stainless steel of the 18-8 type. This was cast in a silicate-bonded mold with calcined limestone (CaO) chips as the soluble aggregate. Since no attempt was made in this limited experiment to bond the chips, it was necessary to fit a ceramic strainer core over the filled cavity in order to keep the aggregate in place during the casting operation. The mold was heated to 1800 F (982 C), and maintained at temperature for a sufficient period for the limestone to become completely calcined. The molten alloy was then infiltrated by the gravity technique.

Leaching was accomplished with a dilute nitric acid solution. This part of the process was rather slow, but the test did demonstrate the feasibility of producing cellular metal of a ferrous composition. It is planned to investigate this area in greater detail during the second phase of this program.

## COMPOSITE STRUCTURES

### *Secondary Infiltration*

The nature of cellular metal suggests the possibility of producing composite structures which may have unique and interesting properties. These structures would be a marked departure from the type currently being used and developed. The cellular skeleton represents a continuous phase of one material. The volume represented by the interconnecting cells can be filled with another material which will also be present in the composite as an independently continuous phase. A composite cast structure of this type may subsequently be subjected to working operations which, under certain conditions, will produce bonding between the two phases.

The cellular metal development promises the possibility of developing a new family of materials, which are composed of two independent but continuous matrices with a variety of properties resulting from the characteristics of the combination. While this possibility was immediately apparent when the cellular structures were first conceived, it was decided to defer development of this phase until the initial process development represented by this report had been completed.

Several composite structures, however, were prepared in order to demonstrate the feasibility of the idea. Double matrix composite materials are produced by infiltration of a cellular metal skeleton of a second material. The composites produced by this technique are – aluminum-lead, aluminum-tin, aluminum plastic and aluminum-aluminum alloy (No. 43).

These composites are representative of the many combinations that are possible. A more complete study of this aspect was deferred to the second phase of this project.

### *Nonsoluble Aggregates*

Composite structures containing nonsoluble aggregates are produced by substantially the same process as is used with soluble aggregates. Where the aggregate material will not form a coherent briquette prior to infiltration, it is necessary to provide some means to prevent the particles from moving during casting. This is conveniently done either by capping with a coarse salt aggregate, where conditions permit, or by the use of a coarse silicate-bonded sand. The mesh size of the aggregate in all cases corresponded to the range of sizes possible with the soluble aggregates. These composites were prepared – aluminum-graphite, aluminum-silica, aluminum-boron carbide and a combination of aluminum and a proprietary ceramic material in the form of hollow spheres which have a bulk density of 0.3 to 0.4 gm/cc.

The aluminum-boron carbide composite was clad by casting a layer of pure aluminum and rolling out into strip. This corresponded to a properitary material. The working operation was successfully accomplished. The boron carbide content of this strip was approximately twice that of the proprietary material. This particular composite is being studied under another program. The other composites were prepared as illustrative of this material concept. The aluminum-graphite may be useful as a high temperature, dry bearing material. The aluminum-silica composite may be used as a lightweight abrasion resistant material. The combination of aluminum and the proprietary material may be used as a high-strength-to-weight ratio material.

Cellular aluminum can be most conveniently machined if the machining operation is performed prior to leaching. It was found generally that slower machining speeds, but comparatively heavy feeds, are most satisfactory. The abrasive nature of the salt tends to dull the cutting tool rapidly and, for this reason, carbide tools were used whenever possible. Leached material could be successfully machined without excessive smearing or fracturing of the cell walls if the cell size is smaller than 20 mesh.

## TEST PROCEDURES AND RESULTS

### *Infiltration Efficiency*

It is interesting to determine what efficiency is actually attained in the infiltration operation. In this respect, efficiency is regarded as the ratio of the weight of metal actually infiltrated to the theoretical weight possible to infiltrate. When this ratio is less than unity, either the interstices between the salt grains are not completely filled during the process of casting or, after casting, voids develop in the metallic phase due to such factors as gas precipitation or unfed volumetric shrinkage. The measure of efficiency of infiltration, therefore, is one measure of the quality of the material.

The efficiency of infiltration can be easily ascertained in a simple geometry sample, such as a cylinder. By measuring its diameter and height, the total volume $V_c$ can be determined. The weight of the composite $W_c$ can also be determined. After leaching, the weight can be redetermined, giving $W_a$ the weight of the aluminum. The difference between these two weighings is the actual weight of salt in the sample composite. From the known density of the aluminum $D_a$ and the known density of the salt $D_s$ it is possible to calculate the theoretical weight of aluminum needed to fill completely the interstitial spaces between the salt particles.

This computation is only correct if each salt crystal, which completely fills its cavity at the casting temperature, still completely fills its metal cavity at room temperature. However, this is not valid since a significant difference does exist in the coefficients of thermal expansion of the aluminum and the salt. In cooling the aluminum alloy from its solidification point to room temperature the volumetric contraction is approximately half that of the salt crystals.

This means that after the composite has cooled to room temperature the salt crystals are smaller than the cavities in which they lie. It is therefore necessary to correct for the void volume resulting from the greater contraction of the salt. The exact volume of voids created by this contraction is difficult to determine, due to the fact that the data relating to the contraction coefficient over the entire temperature range are not readily available. It does appear, however, that this factor would be approximately 5 per cent. In order to bring the calculated absolute values of efficiency of infiltration into closer relationship with reality, it was decided to multiply the expression $(W_c - W_a)$, which represents the actual weight of the salt component, by a constant $K$. The value of this constant for the aluminum alloy-NaCl composite is 1.05. The theoretical weight of aluminum $W_a$ can be expressed as:

$$W_a = \frac{D_a}{D_s} [\, V_c D_s - K (W_c - W_a) \,]$$

The per cent efficiency of infiltration, EI, is:

$$EI = \frac{W_a}{W_a} \times 100$$

This equation was used to calculate the *EI* of the specimens used for the compression tests.

As work on this project progressed, as might be expected, a gradual improvement in processing techniques occurred with consequent improvement in the quality of the cellular metal produced. It is important to demonstrate the current level of development with respect to quality.

The most significant parameter which can be related to quality is efficiency of infiltration. As a demonstration of the current quality level, four cellular castings (1⅝-in. in diameter by 6 in. long) were prepared. The cell sizes corresponded to mesh sizes of 6, 10, 20 and 40. In order to emphasize the severity of this test, these castings were made with high purity aluminum rather than the 356 alloy which had been used for most of the prior work. The pure metal substantially increased the problem of overcoming solidification shrinkage. Three one in. high cylinders were removed from designated locations along the length of each casting. These cylinders were machined to 1½-in. diameter, carefully measured and leached, and the efficiency of infiltration and apparent density were calculated. The results of these tests are shown in Fig. 9.

These data show a significant improvement in the efficiency of infiltration of all the cell sizes. All nine samples representing the 6, 10 and 20 cell sizes showed a constant *EI* of more than 98 per cent along the full length of casting tested. The slightly lower *EI* values obtained for the 40 cell size material are probably due to the larger number of interstices between adjacent grains that have to be filled. Incomplete filling is probably responsible for deviations from the desired 100 per cent *EI*.

The apparent density values are relatively constant over the length of each specimen. The variation in apparent density among specimens is due to differences existing in the particle size distributions of the sieve fractions used. Even though a single sieve fraction was taken for each cell size, certain variations in relative particle size distribution would be expected.

## COMPRESSIVE STRENGTH

The mechanical properties of cellular aluminum have yet to be completely defined. Because of its simplicity and the relationship that it bears to the structural loading anticipated for the material, compressive strength was selected as the principal parameter for study. Specimens for test were machined to 1.500 in. diameter cylinders 1.000 in. high. They represented cell dimensions corresponding to granule sizes of 4 to 60 mesh with both angular and spherical shapes.

The 356 alloy (7 per cent Si - 0.3 per cent Mg) was selected as a representative aluminum alloy and

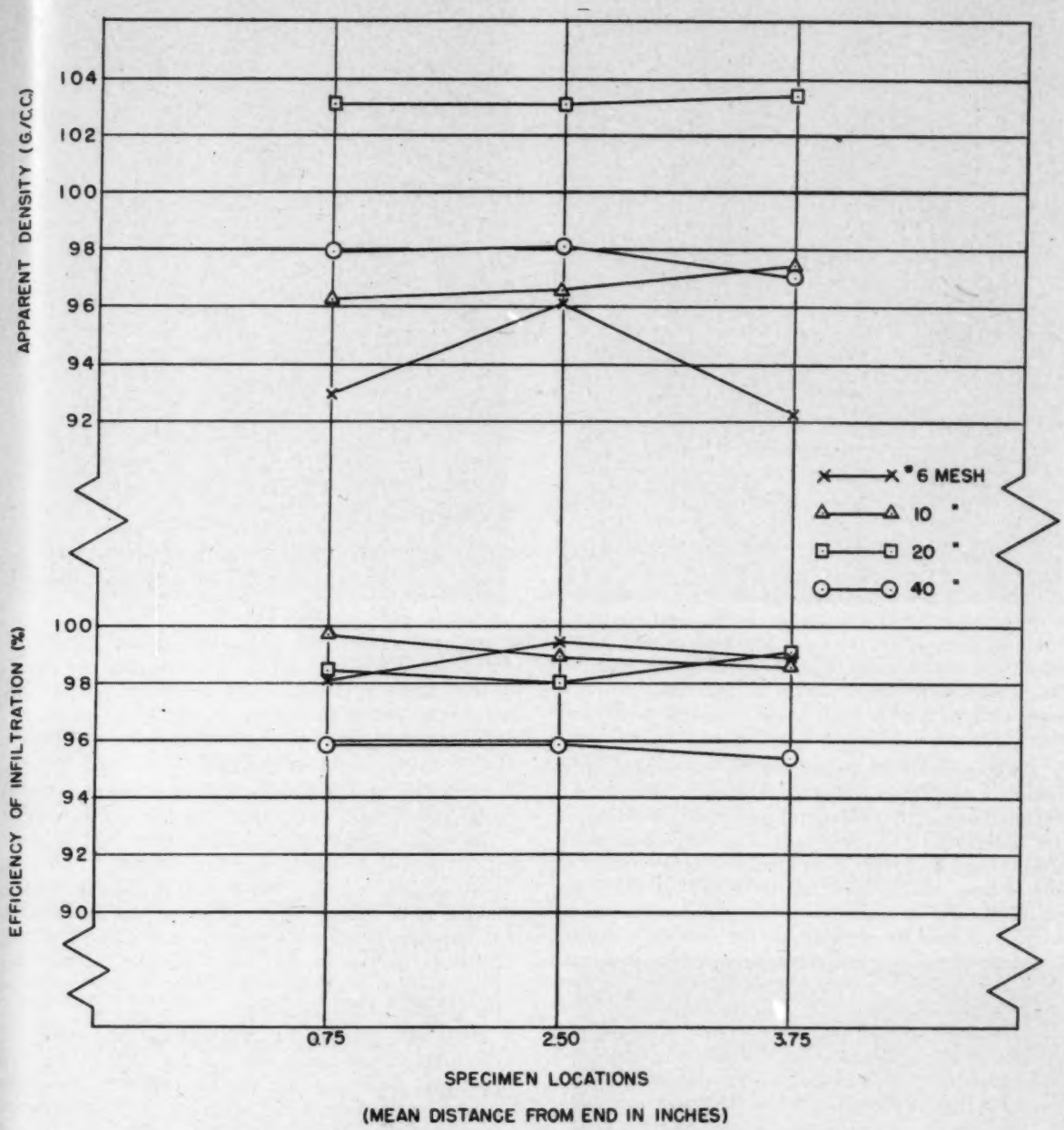



**Fig. 9 — Apparent density and efficiency of infiltration of high purity aluminum cellular castings.**

the specimens were tested in both the as-cast and heat treated condition. The compressive index figure was the one commonly used for materials of this type, which is the stress level recorded at 10 per cent compression. In addition, the stress level at 0.2 per cent offset was also recorded. One series of specimens was tested as metal-salt composites and a few metal-plastic and metal-ceramic specimens were also included.

The table gives a few representative values obtained by compression testing cellular and composite samples:

| Material | Cell Size, mesh | Yield Strength, psi, 0.2% offset | Compressive St., psi, 10% def. | Apparent Density, g/cc | Efficiency of Infiltration, % |
|---|---|---|---|---|---|
| **Cellular Materials** | | | | | |
| 356 As Cast | 10 | 1870 | 3020 | 1.018 | 93.7 |
| 356 As Cast | 30 | 2150 | 3700 | 1.102 | 92.5 |
| 356 T-6 | 10 | 2780 | 4300 | 1.009 | 91.3 |
| 356 T-6 | 30 | 2510 | 4390 | 1.067 | 90.6 |
| **Composite Materials** | | | | | |
| 356 + Epoxy Resin | 60 | 8300 | — | 1.687 | |
| 356 + Propietary Material | 60 | 17,300 | — | 1.643 | |
| 356 + Salt | 20 | 10,650 | — | 2.30 | |

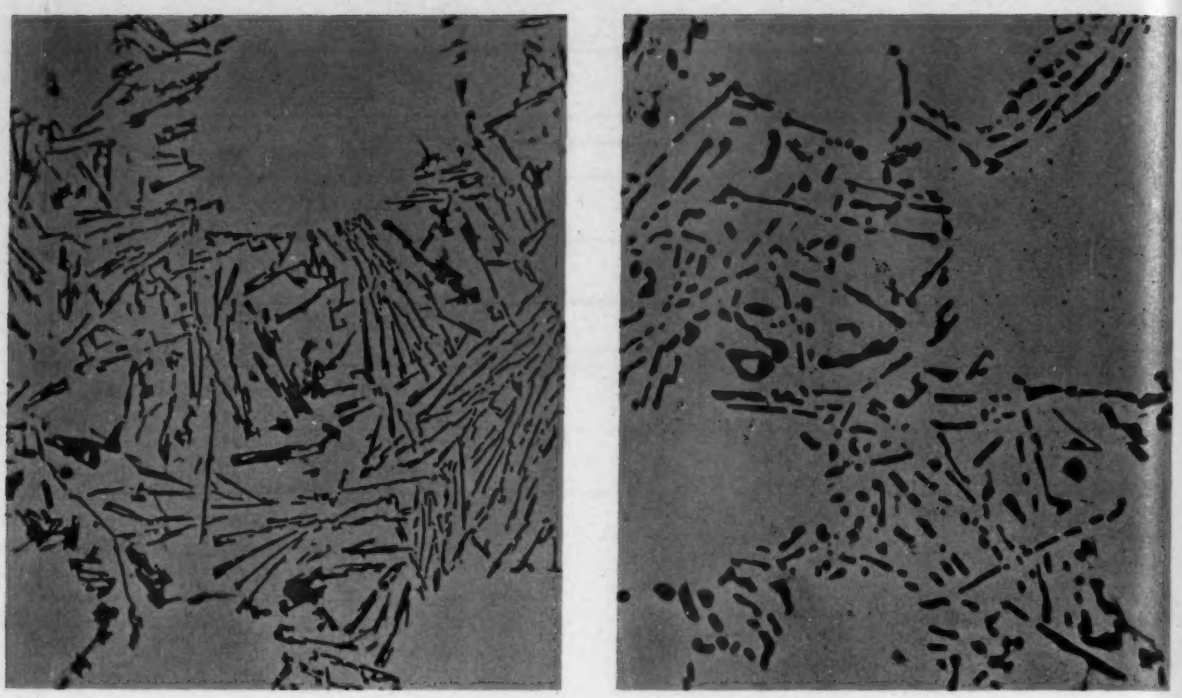


Fig. 10 — Microstructure of cellular 356 aluminum alloy no. 10 (1.65 mm) cell. 250 ×. *Left* — as-cast structure. *Right* — T-6 structure.

These data are illustrative of the range of properties that can be expected from these materials. An examination of all the data indicates that the strengths are independent of mesh size. The variations which were found could be attributed to characteristics of the sample which were related either to efficiency of infiltration or apparent density values. The use of a filler to provide support for the cell walls during compressive loading substantially increases the load-carrying capacity of the material. The aluminum alloy-proprietary composite is a particularly interesting example of this kind of structure. Though the density of the material is only about 60 per cent of solid aluminum alloy its compressive yield strength approaches that of the solid alloy in the as-cast condition.

## MICROSTRUCTURE

Metallographic samples were prepared from cellular 356 alloy of an intermediate cell size (10). One of the samples represented the as-cast condition and the other the solution treated and aged condition (T-6). These structures are shown in Fig. 10. The silicon constituent is present as a coarse phase in both conditions. Prior to heat treatment, the lamellar silicon structure is typical of alloys of this type.

After heat treatment (which consisted of solution treating at 1000 F (538 C) for 16 hr, quenching, and aging at 310 F (154 C) for 4 hr, the silicon particles became rounded. A few finely dispersed voids evident in the structures are probably due to unfed solidification shrinkage.

## DENSITY

The final density obtained in a cellular casting is intimately related to the classic problem of packing of particulate matter. This problem is treated in great detail by Parisi, Nutter and Michalowski.[2] It is most conveniently handled by assuming the particulate matter to be spheres. On the basis of this assumption, a container filled with uniform size spheres of any diameter packed in the closest possible manner and neglecting edge effects will occupy 74.0 per cent of the volume of the container.

It is also possible to accommodate one sphere of size 0.414 D, and two spheres of size 0.225 D, for each principal sphere of size D. The smaller spheres fit in the space between the adjoining principal spheres. Ideally, it is possible to pack an additional 7.1 per cent of the 0.414 D spheres and 5.3 per cent of 0.225 D spheres into the same cavity. These percentages are based on the volume of the principal sphere fraction. It is therefore theoretically possible to fill the mold to 83.17 per cent capacity, using the appropriate percentages of these three size ratios.

This would indicate that it is possible under these conditions to produce a cellular metal having a void volume of approximately 83 per cent, and, conversely, an apparent density of approximately 17 per cent of the true metal density.

Since ideal particle shapes, size distributions and packing efficiency cannot be attained in practice, the apparent densities obtained are substantially higher than those indicated from theoretical considerations.

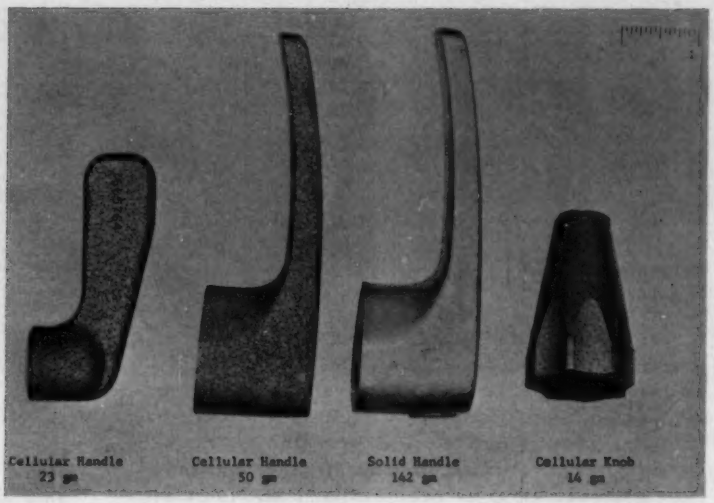



Fig. 11 — Cellular aluminum alloy castings.

In addition, since this was the initial phase of the project, it was believed that it would be desirable to limit this work to cellular structures having a single size distribution.

## CASTINGS

One of the advantages of the cellular metal technique is the capability of producing shaped castings to closely controlled tolerances. As a demonstration of this characteristic of the process, several investment molds were prepared from existing patterns. These molds were invested and dewaxed by standard techniques and then filled with 40 mesh salt crystals. The molds were then heated to the casting temperature for 356 alloy, infiltrated and cooled in the manner already described. Figure 11 shows three such castings.

For comparison purposes, one solid investment casting is also shown. It is evident that, aside from the inherent porosity of the material, the surfaces and detail are equivalent to those of conventional castings. One of the characteristics of the as-cast surface is that, in comparison with the cell dimensions, only small pinholes appear at the surface. This is due to the fact that the junction between the crystals and the mold wall, for the most part, is a point of contact. These surface holes could be further reduced in size by the use of spherical particle shapes and by an increase in the pressure used for infiltration.

Another shaped casting that was produced was the electronic mounting plate shown in Fig. 12. This casting was made in a silicate-bonded mold. Suction in combination with 60-cycle vibration was used to effect infiltration. A 12-mesh crystal was used. Since the requirements of this component were such that a solid flange was desired the pattern was molded in the flange-up position, and only that part of the mold which was intended to be cellular was filled with the crystals. Figure 13 also shows the casting in cross-section, and reveals the solid flange and the cellular structure of the body.

A number of billets approximately 10 in. in diameter by 24 in. long were cast by the techniques described for billet castings. One of these billets was required to have a solid core running through the axis of the casting. This was accomplished by positioning a metal core in the center of the mold before filling. The core was lubricated and wrapped with kraft paper prior to insertion in the mold to prevent the salt crystals from adhering to its surface. The mold was then heated sufficiently to drive off the mold moisture.

The salt solution on the crystals, which was formed by the condensation of the mold vapors, was evaporated during the heating cycle. The crystallization of the salt at at the point of junction of the salt crystals made the salt component a coherent mass. The metal core was then withdrawn. Any of the paper wrapping not removed at this time was consumed during the later firing of the mold. The mold was then heated to the proper temperature for casting and infiltrated with molten aluminum.

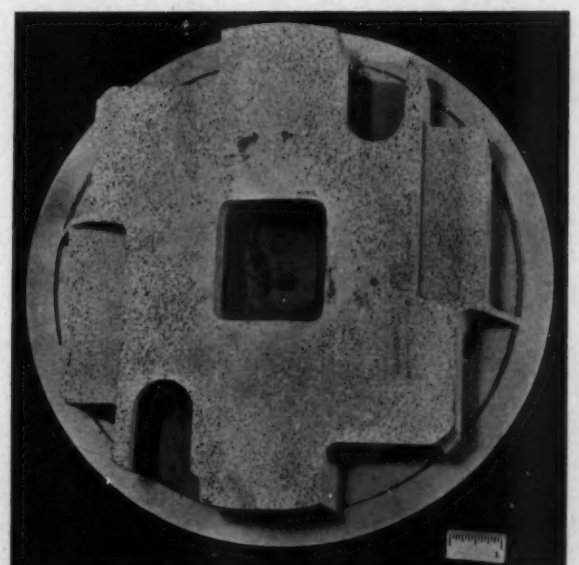


Fig. 12 — Cellular 356 aluminum alloy mounting plate casting.

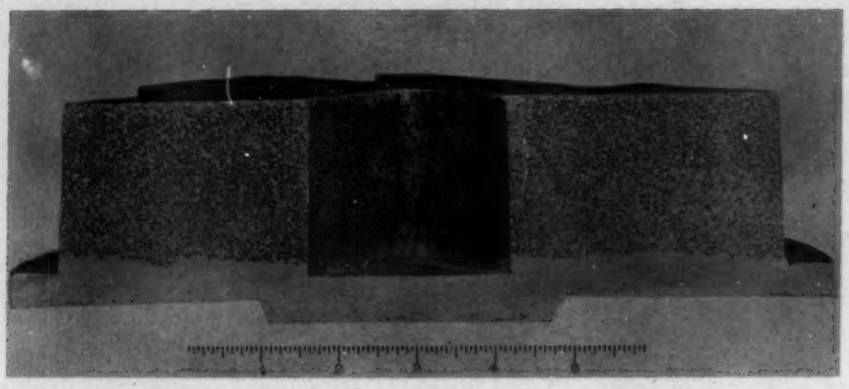


Fig. 13 — Cross-section of cellular aluminum mounting plate casting.

In this case the crystal size was 10 mesh and only vibration was used as an aid to infiltration. Figure 14 shows a transverse section taken from this casting. Similar techniques can be used to produce other combinations of solid-cellular geometry.

## POSSIBLE APPLICATIONS

While the text of this report makes occasional reference to possible applications for cellular metals, it is appropriate that this report should contain a consolidated description of those applications which have either been envisaged by the authors or suggested to them. It is emphasized, however, that these suggested uses are speculative, and are indicated for the purpose of defining some of the areas where the unique properties of this material may be used to advantage.

The successful development of any single application depends, first of all, upon careful design to realize the full potential of the particular characteristics of cellular metal which are being exploited for the application. In addition, the material itself, which can be produced with an infinite variety of characteristics, must be tailored for the specific application.

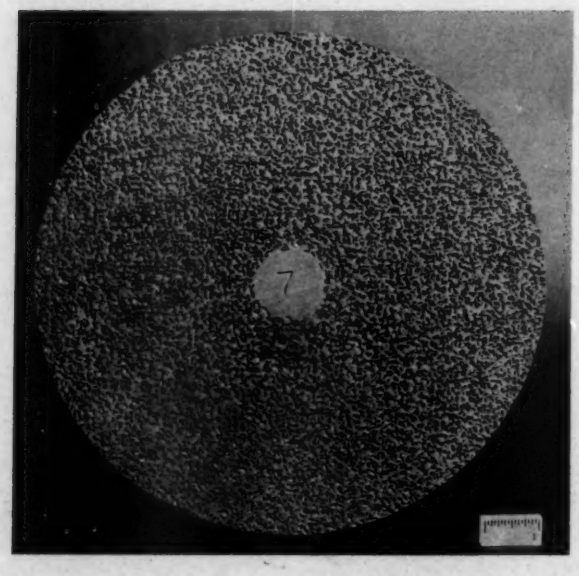


Fig. 14 — Cross-section of cellular 356 aluminum billet.

### *Metallic Filters*

The large void volume promises the possibility of filter elements which would have substantially greater particle retention characteristics than powder metal filters. It would be necessary to refine techniques used for cellular metal production in order to exercise greater control of the dimensions of the intercellular channels. This would probably require the use of accurately controlled spherical soluble granules.

### *Exhaust Muffler Elements*

Powder metallurgy components have recently come into use as diffusers for the exhaust of pneumatic tools. Applications of this type can no doubt be filled by cellular metals with equal or better facility. This suggests the possibility of use as exhaust elements for internal combustion engines. Components made of cellular material may be much more efficient than conventional baffle elements and may possibly be designed to reduce the amount of exhaust particles released to the atmosphere.

### *Bearing Materials*

The cellular skeleton may be used as a means for providing structural support for lead and tin base bearing compositions. The characteristics of the material permit much larger percentages of the active bearing component to be incorporated in the material than is possible with either powder metallurgy or conventional alloying.

### *Sandwich Core Material*

The nature of cellular metal is such that it is not competitive with honeycomb structures because of its substantially greater density than that typical of

honeycomb construction. It may, however, find application in structures where contoured sections must be supported by a lightweight core. This may be in such areas as leading edges of wings or fin surfaces. This can be accomplished by using cellular cores and casting the desired skin against it. The metal would be firmly keyed to the cells at the surface of the core and would benefit from the chilling that the cellular core would provide.

Such a structure would probably require neither brazing nor organic adhesive for assembly. More important than this, however, is that it may be possible to provide transportation cooling for such structures by circulation of suitable coolants through the interconnecting cells that make up the core.

### *Jigs and Fixtures*

Cellular metal, particularly in fine mesh sizes, is capable of being easily deformed to take reasonably accurate impressions of a solid metal body of greater strength. This property suggests a possible use in the construction of machining jigs and fixtures. A stock part may be used to hob its impression in a block of specially prepared cellular material. This impression can then be mounted in the particular machine or fixture intended for the operation, and subsequent parts can be accurately positioned by means of the hobbed cavity.

The porous nature of the material may also be of benefit in permitting coolant liquid of gasses to be directed against the component while it is being machined or otherwise processed. The low density of the material may be a further significant advantage, particularly if the fixture requires manual handling.

### *Energy Absorption Material*

One of the unique properties of cellular metal in comparison with other metallic materials is the nature of its response to compressive loads. Cellular metal can absorb energy by the mechanism of relatively easy and controlled collapse of the cell walls. This suggests that in a combination of two materials – one strong and the other weak, the weaker material may afford protection for the stronger material by its sacrificial collapse and attendant absorption of energy. This characteristic suggests applications where it is desired to absorb energy released by accidental impacts.

A cellular metal element of the proper energy absorbing capacity may be useful in preventing serious damage to a vehicle and its occupants when it is involved in an accidental collision.

### *Heat Transfer*

The limited study of the thermal properties of cellular metal suggests a variety of applications based upon its unique thermal characteristics. In dynamic systems, the large, controlled surface area can be exploited for efficient heat exchange, particularly between a heat source and a gas flowing through the system. In passive systems, the low thermal conductivity may be used for thermal insulation, a function not normally possible with metals.

### *Electrochemical Processes*

The combination of permeability and large surface area may be employed to advantage in electrochemical processes. This would be principally in connection with electrodes which can be prepared to provide increased surface area. These applications include plating processes, battery electrodes and anodes for galvanic corrosion protection of such structures as water tanks, pipe lines and ship hulls.

### *Electronic Packaging Systems*

One of the most critical segments of modern missile systems is the electronic package which is a part of the fuse system. In order to obtain maximum reliability in this area, a heavy weight penalty is paid by the use of a massive, rigid chassis to support the electronic components. The chief function of this chassis is to provide electrical shielding and to protect the components from damage due to vibration generated during launching and in flight.

It is this particular application that originally stimulated the development of cellular aluminum. The cellular structure promises to reduce chassis weight by almost two-thirds while providing effective shielding and acting as an absorber for the vibrational energy generated during flight. Several prototype components have already been produced and are being tested.

The process for cellular metal manufacture may be adapted to the production of composite materials. This can be accomplished either by means of secondary infiltration of a lower melting material into the cellular skelton or by using the secondary substance as the aggregate for filling the mold. The former method is illustrated by the proposed application for bearing materials. The latter method may be used for the production of composites for radiological shielding. Some of the systems thus far suggested include lead-boron carbide and aluminum-boron carbide. By this means relatively large volumes of boron carbide can be retained in an aluminum matrix, which will not only support the active shielding material but will be of structural benefit to the system.

## ACKNOWLEDGMENT

The authors wish to express their gratitude to Messrs. R. W. Warren and R. R. Palmisano of the Diamond Ordnance Fuze Laboratory, and Mr. H. Rosenthal of Pitman-Dunn Laboratories, for their interest and the many constructive suggestions offered during the course of this study.

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# WELDABLE AS-CAST NICKEL ALUMINUM BRONZE COMPOSITION AND MICROSTRUCTURE

by M. L. Foster and S. Goldspiel


## ABSTRACT

**The potentials of Ni-Al bronze for propeller castings, which include good mechanical properties and excellent cavitation resistance, could be enhanced by a determination of compositions and structures which give consistently weldable as-cast material. This is a report of the first phase of an investigation designed for this purpose and deals with the relationship between aluminum content, mechanical properties, microstructure and weldability. The sample material included conventional keel blocks and specimens taken from a groove-welded wedge-block casting, grading in thickness from ½ to 3½-in., and simulating during welding restraints which are encountered in actual propeller repairs. The results of tests indicate that the minimum percentage of Al needed for a weldable structure is close to 9 per cent, and that the optimum as-cast structure in the thickness range considered is obtained with 9.5 per cent Al at 5, 4, and 0.5 per cent Ni, Fe and Mn, respectively, which were kept relatively constant. Pb and Si, elements considered as "detrimental," did not adversely affect weldability at levels of 0.10 and 0.18 per cent, respectively, in an otherwise optimum composition. Optimum mechanical properties obtained to date and relationships between metallographic structure, mechanical properties and weldability, are discussed.**


## INTRODUCTION

Of the several alloy types being considered for propeller castings, nickel-aluminum bronze is one of the more promising. Under optimum conditions, castings of this alloy type combine high strength and ductility with excellent corrosion (cavitation and erosion) resistance. To date, propellers of nickel-aluminum bronze as large as 50 tons have been cast for commercial use. Some propellers of this alloy type have been used successfully by the Navy in ships which operate under extremes in marine conditions, as represented by arctic waters.

This alloy is lighter than the high tensile brasses, and hence makes possible (a) considerable savings in fuel consumption, (b) reduced wear and tear on main propeller shafts and (c) the use of thinner sections in design, due to its higher strength-to-weight ratio. Its resistance to corrosion allows propellers to retain smooth surfaces for a longer time, thereby increasing propeller efficiency and reducing maintenance costs. The major obstacle, reported for the alloy type, is that its fabricability (particularly bending and welding properties) often present difficulties.

It is the purpose of this investigation to study compositions and structures of castings of this material which, with due allowance for mass effects, will yield consistently fabricable material, including good castability, bendability and weldability.

## LITERATURE REVIEW

The search for improved propeller alloy materials for maritime as well as naval use within the past ten years has resulted in the appearance of a number of articles in the technical literature on nickel-aluminum bronze. A brief review of the literature will highlight results of previous work and indicate the manner in which the current investigation has been planned to supplement and extend it. Vanick,[1] in a comprehensive review article, considers the Ni-Al bronze casting alloy with relation to tin bronze, red brass and manganese bronze, the latter being used as a yardstick as far as properties are concerned.

In addition to nominal compositions, he presents comparative data on salt water fatigue, cavitation resistance, corrosion fatigue and corrosion in sea water under varying types of exposure. He also treats briefly the subjects of heat treatment, heat and wear resistance, fabricability including machinability, welding and brazing. The two compositions of Ni-Al bronze alloy discussed in this paper are based on earlier development work of W. C. Stewart and W. L. Williams.[2] It is noted that the elongation quoted for the two alloys considered is lower than specification requirements[10] for the currently used alloys.

As far as weldability is concerned, Vanick states that the use of standard aluminum bronze filler with metal arc or oxyacetylene produces satisfactory welds, with better ductility than can be achieved with Ni-Al bronze filler. The work does not indicate the extent to

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NOTE: The opinions or assertions contained in this paper are the private ones of the authors and are not to be construed as official or reflecting the views of the Naval Service at large.



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which the nominal composition must be adhered to in order to achieve the good mechanical properties and fabricability claimed. In addition, the effects of mass on structure, properties and fabricability, while briefly mentioned, leave much to be desired for actual guidance and production. Impact properties for the alloys were not presented.

Cook, Fentiman and Davis[3] give a rather extensive discussion of the properties of wrought Ni-Al bronze as related to composition and metallographic structure, and, thus, provide considerable background for an understanding of factors which must be controlled in the development of cast alloy counterparts.

*Inert Gas Metal-Arc Welding*

Welding of Ni-Al bronze castings is treated in considerable detail by Cahill.[4] The author discusses welding using the inert gas metal-arc process with both Ni-Al bronze and Al bronze filler wires. He concludes that the latter is considered preferable at the time of the study, but that continued experimental work in progress on modifications of the former wire may lead to higher strength deposits. It is of interest to note that the procedure developed by Cahill had the decided advantage that little or no preheat was required compared to the preheat, of as high as 800 F, needed for the method used previously, i.e., manual metal-arc welding with Al bronze. Cahill's work is, however, limited to a uniform section thickness, and takes no account of variations in nominal composition or the effects of impurities (normally present in commercial ingot and returns) on weldability.

A discussion of all types of aluminum bronzes, from a point of view which is of interest to designers, is contained in an article by W. L. Williams.[5] Applications considered include shipboard machinery, condenser components and propellers, with emphasis on requirements for naval vessels. Comparative chemical composition, mechanical properties, corrosion, fatigue and magnetic property data are included. The author makes the important observation that "aluminum bronzes can be used for a wide variety of shipboard hardware, provided designs are consistent with properties. It is important, however, that the individual alloys be screened carefully in selecting materials for specific applications."

The current investigation at the Material Laboratory is being conducted with the latter thought in mind, because Ni-Al bronze for large propeller castings must (a) tolerate ingot and remelt with reasonable amounts of impurities to keep cost of material down, (b) be fabricable (bendable and weldable) in the as-cast condition, since heat treatments of most ship propellers are impractical, due to size and (c) be repairable, if they contain defects associated with original casting. In addition, the choice of the individual alloy should be backed by a sufficient amount of information on heat treatment to improve fabricability, if necessary, and thus obviate sizeable financial losses otherwise entailed.

Valuable information on casting practices for Ni-Al bronze as applied to the production of large propellers, is contained in a paper by A. J. Smith.[6] Factors considered in detail include melting practice, gas absorption, molding methods and materials, refractories, fluxes and pouring temperatures for both small and large castings. His treatment of large casting practices is of particular value because they are significantly different than for smaller castings. This knowledge is limited to a small number of large producers.

Comparative advantages to manganese bronze are also discussed, and include improved propeller efficiency, 18 per cent reduction in weight, low notch-sensitivity and overall economy, considering service life and lower maintenance costs. The author's statement that "aluminum bronze is readily repairable" is, however, not completely borne out by experiences reported by a number of investigators and observed in connection with a limited number of repair jobs, in which the Material Laboratory was involved. In this connection, it should be noted that this factor represents one of the prime considerations in the Laboratory's current investigation, since under certain conditions of solidification, critical amounts of alloying elements or presence of impurities may influence weldability.

## PREVIOUS WORK

For purposes of completeness, a summary of some preliminary work by the Material Laboratory, which was reported to the Bureau of Ships[7] as a basis for the support of a formal investigation, is presented. This preliminary work was based on results obtained from 16 heats of nickel-aluminum bronze produced in the Shipyard foundry, since the alloy was first cast here in June 1956. The interest in the alloy at that time came about as a result of a request from the Shipyard welding engineer that the foundry cast a patch of a specified composition (Item A, Table 1 shows the actual analysis). This patch was to be used for the salvage by repair welding of an 18 ft diameter ship propeller.

Since no stock of the required ingot was available, the foundry, with technical assistance from the Material Laboratory, produced and pigged a small quantity of metal of the desired nominal analysis. Four attempts were necessary to overcome melting and casting difficulties to produce the actual patch which had the shape of a blade tip with chord length maximum width and maximum thickness of 5 ft, 2 ft and 3½-in., respectively. The final casting, produced as shown in Fig. 1, had the specified composition and mechanical properties shown in Item B, Table 1, and a soundness which satisfied rigid radiographic and surface penetrant tests. This patch was satisfactorily welded to the ship propeller under supervision of the Shipyard welding engineer.

Subsequently, in Jan. 1957, at the request of the Shipyard welding engineer, two additional heats were melted and used for casting of test plates which also welded satisfactorily. The compositions and tensile properties of these heats are shown in Table 1, items C and D.

In Feb. 1957, a second patch, similar in shape to the first one described above, was cast for the repair of another ship propeller. The analysis of the pro-

TABLE 1 — DATA FOR PRELIMINARY EXPERIMENTAL CAST NICKEL ALUMINUM BRONZE HEATS

| | | Chemistry | | | | | | | | Mechanical Properties | | | | |
|---|---|---|---|---|---|---|---|---|---|---|---|---|---|---|
| Item | Identification | Cu | Ni | Fe | Mn | Pb | Sn | Si | Al[a] | TS in Kips[j] | YP in Kips[j] | Elong., % in 2 In. | Hardness, Rb | Weldability |
| A | Initial Anal. | 80.5 | 5.2 | 4.0 | 0.5 | — | — | — | 9.4 | 87.5 | 36.7 | 14.0 | 82 | Satisfactory |
| B | 15S[d] | 81.0 | 5.4 | 4.0 | 0.6 | 0.03 | 0.01 | 0.1 | 9.0 | 84.5 | 37.5 | 19.5 | 82-85 | Satisfactory |
| C | 82X[e] | 79.7 | 4.8 | 4.6 | 0.6 | 0.03 | 0.01 | 0.1 | 10.3 | 88.5 | 46.2 | 10.5 | 85-88 | Satisfactory |
| D | 59Y[e] | 79.1 | 4.8 | 4.3 | 0.6 | 0.03 | 0.01 | 0.1 | 11.2 | 87.6 | 32.6 | 15.0 | 83-85 | Satisfactory |
| E | Propeller[f,e] | 80.5 | 5.5 | 5.1 | 0.9 | 0.01 | 0.02 | 0.05 | 8.0 | — | — | — | 86-92 | Satisfactory |
| $F_1$ | A4[g] | 82.1 | 5.4 | 3.6 | 0.7 | 0.06 | 0.01 | 0.1 | 8.3 | 78.5 | 39.5 | 25.0 | 71-74 | Cracked Badly |
| $F_2$ | Patch A4[h] | — | — | — | — | — | — | — | — | 78.0 | 23.7 | 23.0 | 76-80 | Cracked Badly |
| G | R100 6A[b] | — | — | — | — | 0.02 | 0.01 | 0.2 | — | — | — | — | 92-95 | Satisfactory |
| H | R100 6C[b] | — | — | — | — | 0.07 | 0.01 | 0.05 | — | — | — | — | 74-79 | Cracked Badly |
| I | R100 6E[b] | — | — | — | — | 0.09 | 0.04 | 0.05 | — | — | — | — | 74-79 | Satisfactory |
| J | R100 6G[b] | — | — | — | — | 0.13 | 0.01 | 0.05 | — | — | — | — | 76-82 | Cracked Badly |
| K | R100 6H[b] | — | — | — | — | 0.15 | 0.01 | 0.05 | — | — | — | — | 76-85 | Satisfactory with peening |
| L | G51 | 80.2 | 5.0 | 4.4 | 0.79 | 0.01 | 0.01 | 0.05 | 9.5 | 83.0 | 38.0 | 19.5 | 78-81 | Satisfactory |
| M | G99 | 79.8 | 5.2 | 4.0 | 0.57 | 0.04 | — | — | 10.3 | 80.0 | 37.5 | 15.0 | 85-87 | Satisfactory |

NOTES: a. By difference.
b. Material furnished by industrial firms.
c. Plus Mg.
d. Satisfactory patch for one ship's propeller repair.
e. Weld test plate casting.
f. Second ship's propeller proper.
g. Separately cast test bar for unsatisfactory patch intended for ship (f) propeller repair.
h. Tensile bar cut from patch casting.
j. 1000 lb/sq. in.

peller material is shown under item E, Table 1. The manufacture of this patch involved essentially the same procedure established in casting the first successful patch. The composition and mechanical properties of the separately cast test bar and the patch itself are shown in Table 1, items $F_1$ and $F_2$, respectively. The patch cracked on welding.

This behavior could not be explained by the composition and properties of the separately cast test bar which indicated a lower aluminum content, higher elongation and lower tensile strength compared to the first patch. These differences would generally be expected to result in improved weldability. A check was therefore made on the bar machined from the patch proper to determine significant differences between it and the separately cast test bar. The results showed no significant differences. Test plates from succeeding heats made with the same procedure as for the satisfactory patch also cracked on welding.





Fig. 1 — Satisfaction patch casting produced for repair of one ship's propeller. *Top* — riser end view. *Bottom* — gate end view.

### *Chemical and Metallographic Tests*

At this point, it was decided to check whether the difference in weldability might be associated with (a) presence of tramp elements; (b) microstructure as affected by composition, mode of solidification and rate of cooling.

A chemical check for phosphorus and sulfur on (a) the successful patch, (b) the ship propeller proper and (c) the unsuccessful patch indicated no significant differences, all results being below 0.01 per cent. A comparative spectrographic analysis for metallic impurities of the several heats made indicated that the ship propeller contained magnesium, and that the patch which cracked had higher lead than the other samples. The results are shown in Table 1, under items B, C, D, E and $F_2$.

The metallographic structures for the several samples discussed and presented in Fig. 2 show major differences. The structures can be grouped into three significantly different types. Figure 2a shows the microstructure representative of the three foundry heats

which welded without trouble. It will be noted that it contains essentially two phases, one of which is in the form of elongated grains. Figure 2b represents the microstructure of the ship propeller casting. It is also a two phase structure, but differs from the former in that the grains are more equiaxed.

It is to be noted that the propeller material also differs from the foundry castings in that it contains Mg which was probably used to aid deoxidation. No deoxidant additions were used in the Shipyard castings, since it was considered that the relatively high aluminum content of the alloy type would provide the necessary deoxidation under conditions of melting used. The propeller casting welded satisfactorily. Figure 2c represents the structure of the heat from which the unsatisfactory patch, intended for the repair of the ship propeller, was cast. At 120 × the structure appears to be essentially single-phased and equiaxed.

However, the presence of a second phase at grain boundaries is indicated by Fig. 3a, which represents the structure of this material at a magnification of 500 ×. This material consistently cracked on welding. Figure 3b represents the microstructure in the heat affected zone after welding; the fact that the cracking is intergranular in nature appears to be significant.

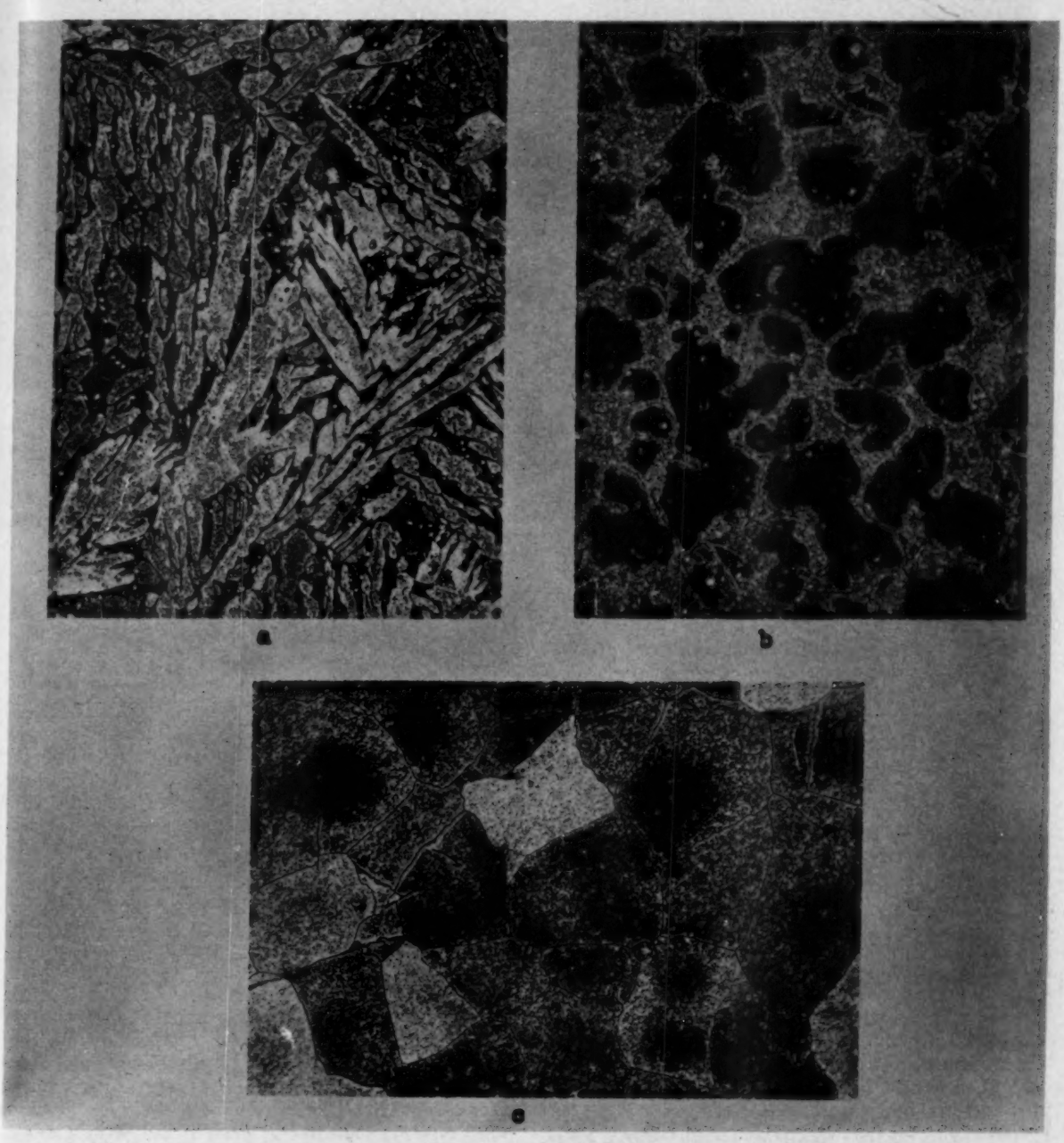



Fig. 2 — Representative photomicrographs of experimental Ni-Al bronze propeller patches in the as-cast condition. (a) Weldable yard foundry patch casting, (b) original, damaged ship's propeller casting and (c) yard foundry patch casting which cracked upon welding. $NH_4OH$ and $H_2O_2$ etch. 120 ×.

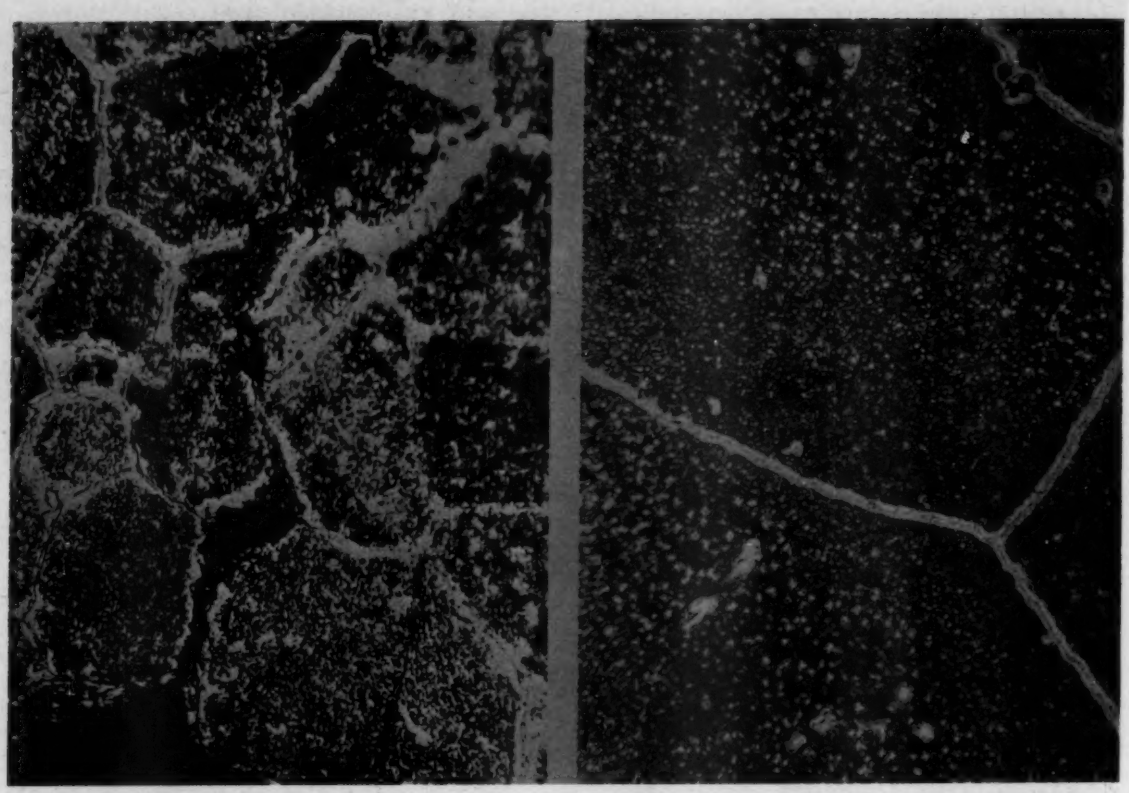


Fig. 3 — Photomicrograph of yard foundry as-cast Ni-Al bronze which consistently cracked upon welding. Item F-1, Table 1. *Right* — casting. *Left* — heat affected zone near weld. $NH_4OH$ and $H_2O_2$ etch. 500 ×.

Of course, the degree of this significance remained to be established by more extensive tests.

Rockwell B hardness values for the several samples, listed in Table 1, indicate that the unweldable structure is softer than the ones which welded without difficulty.

During the progress of the preliminary experimental work several specimens of nickel-aluminum bronze of varying degrees of weldability were received from an industrial firm which uses this alloy type in the manufacture of cast propellers. While the samples were furnished with no other data than performance in welding, they were used to derive additional information for possible correlation of microstructure and tramp elements with weldability. The sample material available was insufficient for a complete chemical analysis.

### *Spectrographic Analyses*

Table 1 lists the results of spectrographic analyses for impurity elements (Pb, Sn and Si) and Rockwell B hardness tests for five samples of the above material (items G, H, I, J and K) against weldability, as indicated by the source of supply. Figure 4 shows representative microstructures for four of these samples. The structure of item G resembles Fig. 2a, and is therefore not shown again.

The results for the proprietary samples do not agree completely with trends indicated by data obtained for the Shipyard foundry samples, and thus raised a number of questions to be resolved. Samples I and H, which have equivalent hardness ranges, comparable impurity content and somewhat similar (nearly single phase) microstructures, show good and bad weldability, respectively. The difference in sharpness of the grain boundary material, which remained to be identified, was expected to hold a clue to the difference in weldability reported. Samples H and J appear to bear out the trend indicated by comparable data for Shipyard foundry samples.

Their lead contents are high, hardnesses are low and structures are similar to those for the foundry material (Fig. 3) and all cracked on welding. Sample K, with highest lead content of all samples considered (i.e. 0.15 per cent), intermediate hardness and a "good" microstructure cracked slightly without peening but welded satisfactorily with peening.

In order to eliminate the possible effects of tramp elements on weldability, it was decided to run subsequent heats with material carefully screened for low Pb, Sn and Si. The heats were calculated to contain Al, Fe, Ni and Mn at levels of about 10, 4, 5 and 1 per cent, respectively, which were estimated to be comparable to the optimum composition for hot worked alloy reported by Cook, Fentiman and Davis,[3] i.e., Al, Fe and Ni at levels of 10.5, 5 and 5 per cent, respectively. Representative data for heats of this type are indicated as items L and M in Table 1.

The hardnesses compare favorably and the tensile strengths are somewhat lower than for the weldable

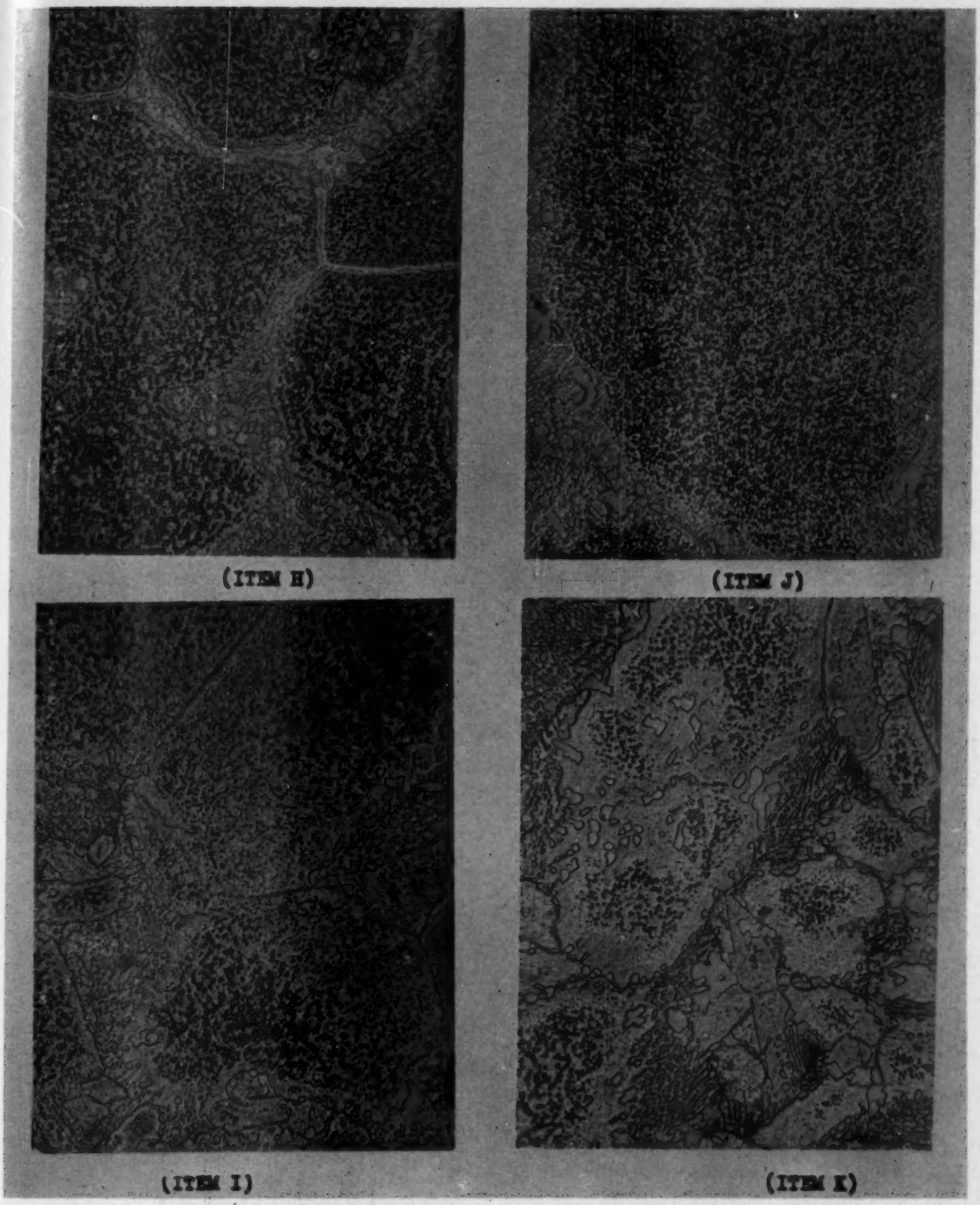



Fig. 4 — Representative microstructures for proprietary cast Ni-Al bronze samples. Items H, I, J and K, Table 1. $NH_2OH$ and $H_2O_2$ etch. 500 ×.

alloys, items B, C, and D. The microstructures of these heats are shown in Fig. 5. It is interesting to note that while both materials (items L and M) were weldable, there is a pronounced difference in the microstructures which could not be explained at the time.

*Preliminary Trends*

On the basis of this limited data reported for cast Ni-Al bronze, the preliminary trends indicated were as follows:

1) Impurity content (Pb, Si, Sn, etc.) may adversely affect weldability, but is not the sole factor involved.
2) Microstructure is an important factor related to weldability.
3) A "weldable" microstructure is dependent on a critical composition range.

In addition, it was considered that the cooling rate during solidification has an effect on the microstructure, and that for each nominal composition there may be a critical cooling rate range which produces a weldable structure.

Since the indications from this limited work were recognized to be trends at best, and since, in addition, they revealed unexplainable anomalies, it was considered that a more extensive investigation was justified, so that while the advantages of the alloy could be realized, the difficulties of the inconsistent weldability and fabricability could be reduced to a minimum. Accordingly, the Material Laboratory requested the Bureau of Ships to sponsor a formal investigation to establish firm trends and to resolve anomalies indicated by its preliminary work.

The project was actually authorized at the end of 1957. The importance of this work has since been highlighted by reports in the literature,[8,9] which shows an increased interest in and use of this cast alloy type and suggests that improvement in weldability to obtain consistent repairability is needed. With regard to welding, the latter report, for example, states "the know-how for welding aluminum bronze (which includes the type with nickel) is gradually being acquired though there have been difficulties to overcome in the process."

The overall objectives of this investigation were:

1) To determine the composition limits which yield a weldable structure under various jobbing foundry melting conditions, with considerations of effects of impurities, mass and microstructure on weldability.
2) To develop suitable heat treatments to convert unfavorable as-cast structures to structures which can be welded consistently without difficulty.

The current progress report deals primarily with aspects which fall under the first objective.

## EXPERIMENTAL PROCEDURE

The experimental procedure used in this investigation was designed to supplement related work already published, or known to be in progress on the relationship between weldability and composition of as-cast nickel-aluminum bronzes, by (a) introducing considerations of the effect of mass (or cooling rate) and (b) obtaining data which could be useful in design, production and salvage operations.

In order to avoid contamination with tramp elements, the sample material for each composition studied was made from ingot prepared from virgin materials at the Shipyard Foundry. The desired and actual compositions of the ingot material made are:

| Element | Desired, % | Actual, % |
|---|---|---|
| Aluminum | 8.0 | 8.2 |
| Nickel | 5.0 | 5.1 |
| Iron | 4.0 | 4.0 |
| Manganese | 0.5 | 0.5 |
| Copper | Rem. | 82.2 |

This ingot material was used to cast seven heats varying in aluminum content from 6 to 12 per cent. Table 2 shows the expected and actual analyses, as determined from test bar coupons for each of the heats.

A uniform foundry practice was adopted and maintained for all heats in this investigation, in order to keep random variables to a minimum. The molds were prepared from a sand mixture which was similar to that used for steels in the Shipyard Foundry, and which had the following formulation:

| | |
|---|---|
| Silica Sand, AFS Fineness 53-65, lb | 630 |
| Western Bentonite, lb | 24 |
| Dextrin, lb | 11 |
| Washing Soda, oz | 2 |
| Moisture, % by wt. | 3.5 |

The dry components were mulled for 4 min. The water was added through the machine spout while mulling, and mulling was continued for an additional 2 min. Preformed sprues, gates and risers were used in all cases, as shown in Figs. 6 and 7. Each mold was dried for 8 hr at a temperature of 450 F in a gas-fired mold drying oven, coated with zircon wash to minimize mold-metal reactions and redried for an additional 5 min prior to use. A 600 lb oil-fired furnace was used in melting.

A slightly oxidizing atmosphere was maintained by careful manual control of the air to oil mixture throughout the melting cycle. The tapping and pouring temperature ranges were 2170 to 2200 and 2050 to 2100 F, respectively, as checked with a calibrated immersion chromel-alumel pyrometer. A commercial neutral flux consisting primarily of $3NaF{\cdot}AlF_3$ plus NaF was added prior to tapping in amounts of one lb/100 lb of charge and thoroughly mixed. The additions for obtaining graded increases in aluminum were made in the furnace just prior to tapping, with an allowance of from 5 to 10 min for homogenization of the melt.

As noted previously for the preliminary work, no external deoxidant was used since it is considered unnecessary for alloys at the level of aluminum involved in this alloy type. A hot top compound was used to

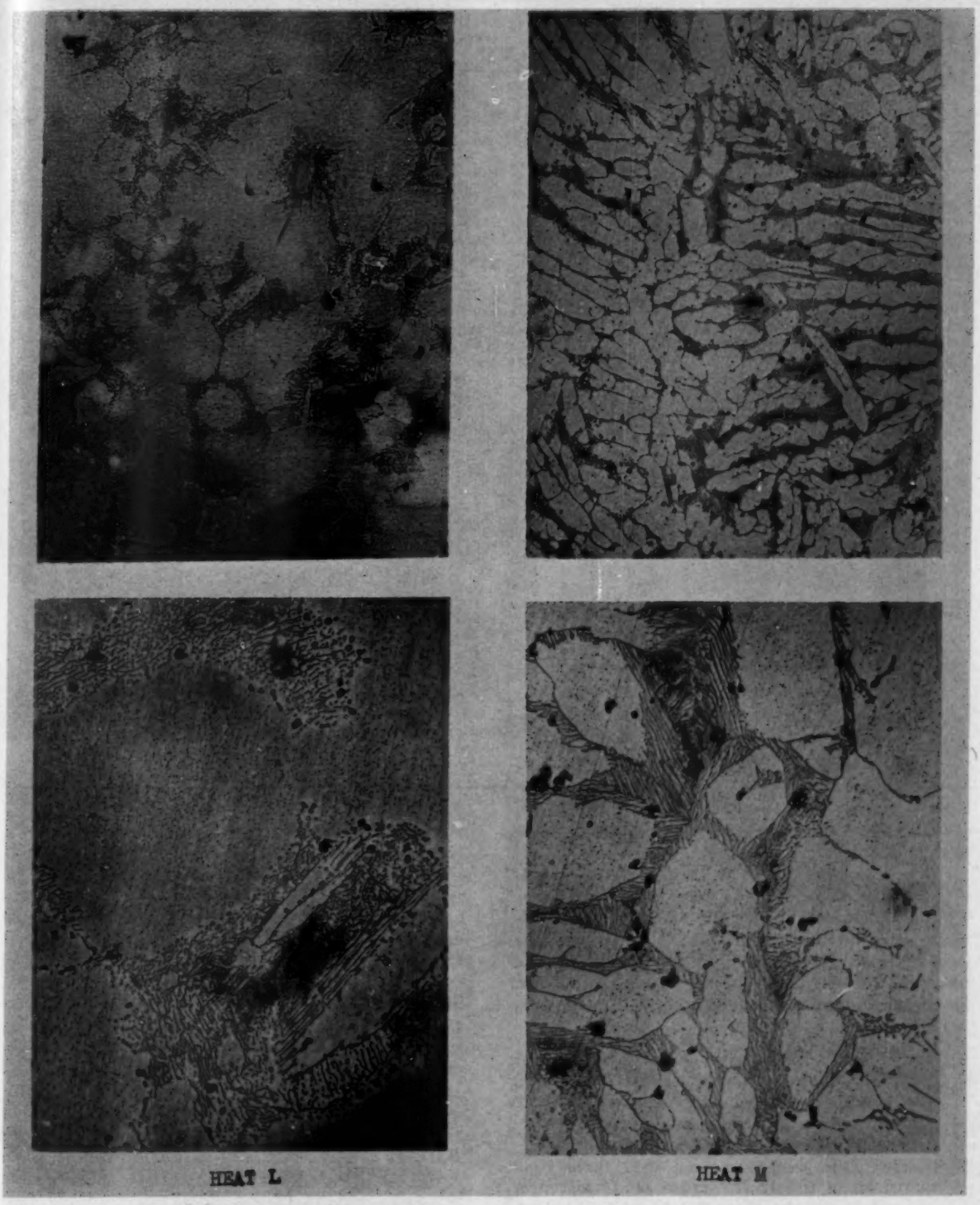



Fig. 5 — Photomicrographs of weldable, composition-adjusted yard foundry cast Ni-Al bronze heats. Items L and M, Table 1. $NH_4OH$ and $H_2O_2$ etch. *Top* — 120 ×. *Bottom* — 500 ×.

TABLE 2 — ALUMINUM CONTENT EFFECT ON MECHANICAL PROPERTIES OF SEPARATELY CAST Ni-Al BRONZE BARS

| Heat No. | Analysis | Composition, % Cu | Al | Ni* | Fe* | Mn* | Pb* | Yield Point, psi | Tensile Strength, psi | Elong., % in 2 in. | Bhn Avg., 3000 kg Load |
|---|---|---|---|---|---|---|---|---|---|---|---|
| | | | | | | | | 30,400 | 75,000 | 33.0 | |
| | Expected | 84.4 | 6.0 | 5.0 | 4.0 | 0.50 | — | 30,200 | 73,600 | 33.5 | |
| 1 | | | | | | | | 31,400 | 74,800 | 34.0 | 121 |
| | Actual | 84.9 | 5.9 | 4.5 | 4.1 | 0.44 | 0.02 | 33,500 | 75,500 | 31.0 | |
| | | | | | | | | 31,400 Avg. | 74,800 Avg. | 32.9 Avg. | |
| | | | | | | | | 34,200 | 83,600 | 26.0 | |
| | Expected | 83.3 | 7.0 | 5.0 | 4.0 | 0.50 | — | 33,600 | 81,600 | 26.0 | |
| 2 | | | | | | | | 32,400 | 77,500 | 28.0 | 131 |
| | Actual | 83.9 | 7.0 | 4.5 | 4.1 | 0.41 | 0.02 | 38,800 | 75,100 | 27.0 | |
| | | | | | | | | 34,800 Avg. | 79,500 Avg. | 26.9 Avg. | |
| | | | | | | | | 44,600 | 82,800 | 18.5 | |
| | Expected | 82.4 | 8.0 | 5.1 | 4.0 | 0.50 | — | 44,000 | 86,500 | 22.0 | |
| 3 | | | | | | | | 47,000 | 81,500 | 13.0 | 143 |
| | Actual | 82.6 | 8.3 | 4.7 | 4.2 | 0.39 | 0.04 | 40,500 | 89,600 | 22.0 | |
| | | | | | | | | 43,000 Avg. | 85,000 Avg. | 20.8 Avg. | |
| | | | | | | | | 45,200 | 78,200 | 8.5 | |
| | Expected | 81.3 | 9.1 | 5.1 | 4.1 | 0.50 | — | 42,000 | 94,300 | 23.5 | |
| 4 | | | | | | | | 42,600 | 93,900 | 21.1 | 156 |
| | Actual | 81.4 | 9.2 | 4.8 | 4.4 | 0.37 | 0.00 | 43,400 | 93,400 | 21.0 | |
| | | | | | | | | 42,700 Avg. | 93,700 Avg. | 21.8 Avg. | |
| | | | | | | | | 46,400 | 94,400 | 15.0 | |
| | Expected | 80.7 | 9.8 | 5.0 | 3.9 | 0.49 | — | 48,000 | 99,000 | 15.0 | |
| 5 | | | | | | | | 47,600 | 92,400 | 13.0 | 170 |
| | Actual | 80.9 | 10.0 | 5.0 | 4.2 | 0.37 | 0.02 | 46,700 | 95,000 | 15.0 | |
| | | | | | | | | 47,200 Avg. | 95,200 Avg. | 14.5 Avg. | |
| | | | | | | | | 55,500 | 97,500 | 3.5 | |
| | Expected | 79.6 | 11.0 | 5.0 | 4.0 | 0.48 | — | 56,500 | 94,000 | 3.0 | |
| 6 | | | | | | | | 59,500 | 93,600 | 2.5 | 217 |
| | Actual | 79.8 | 11.1 | 4.9 | 4.3 | 0.36 | 0.01 | 57,000 | 92,500 | 3.0 | |
| | | | | | | | | 56,300 Avg. | 94,000 Avg. | 3.0 Avg. | |
| | | | | | | | | 69,000 | 95,500 | 1.5 | |
| | Expected | 78.8 | 11.8 | 4.9 | 3.8 | 0.48 | — | 67,500 | 78,500 | 1.0 | |
| 7 | | | | | | | | 71,000 | 98,200 | 1.0 | 241 |
| | Actual | 79.1 | 11.9 | 4.8 | 4.3 | 0.35 | 0.01 | 67,800 | 74,600 | 0.5 | |
| | | | | | | | | 70,000 Avg. | 96,900 Avg. | 1.0 Avg. | |

* Spectographic analysis.

cover the wedge casting risers. Castings were permitted to remain in molds overnight in all cases.

For each composition considered, two wedges and two conventional keel blocks were cast. Figures 8 and 9 are photographs showing the design of the wedge used, including dimensions of the sprue, runners, gates and risers. The keel block was prepared in accordance with Fig. 1 of the applicable specification for nickel-aluminum bronze castings.[10] The test material was used as follows:

A) *Keel Blocks.* These provided a total of four, 0.505 in. unthreaded tensile bars for determining mechanical properties. Representative samples for microscopic examinations were prepared from end portions of the tensile bars.

B) *Wedges.* One wedge was cut into three equal parts, along the 12 in. length, and marked A, B, and C for the nearest, middle and furthest portion with relation to the down sprue, respectively. The second wedge was used for groove weld specimens.

Portion A was used for a surface weld-bead test as a preliminary method for screening out as-cast material with extremely poor weldability. Figure 10 is a sketch of this portion of the wedge showing the direction in which the beads were applied. It will be noted that this procedure was designed to take into account (a) differences in extent of heat affected areas, (b) section thickness and (c) the nature of the surface, i.e., whether it was as-cast or machine cut.

The weld rod used was stranded, 5/64-in. diameter, aluminum-bronze wire, designed to give an overlay bead of 300 Bhn. The inert gas (argon) process was used, and the welding was performed at approximately 60 volts DC and 240 amps. The completed samples were inspected by dye penetrant method for cracks on the surface and on cross-section parallel to the base, at 1¼-in. intervals, from the thin to the thick edges of the wedge.

Portion B was saved for subsequent use in hot bending tests, which will be reported in the next phase of this work.

Portion C was used to prepare Charpy V-notch specimens. A sketch of the wedge showing the location of the specimens and the orientation of the notch is shown in Fig. 11. Charpy V-notch specimens were tested at three temperatures:

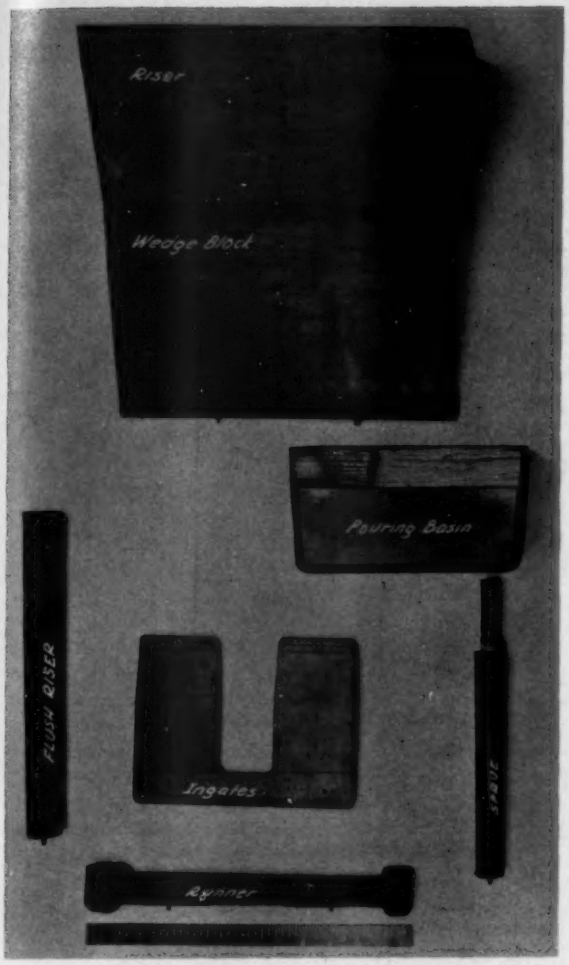



Fig. 6 — Components of preformed gating and risering system used for wedge casting.

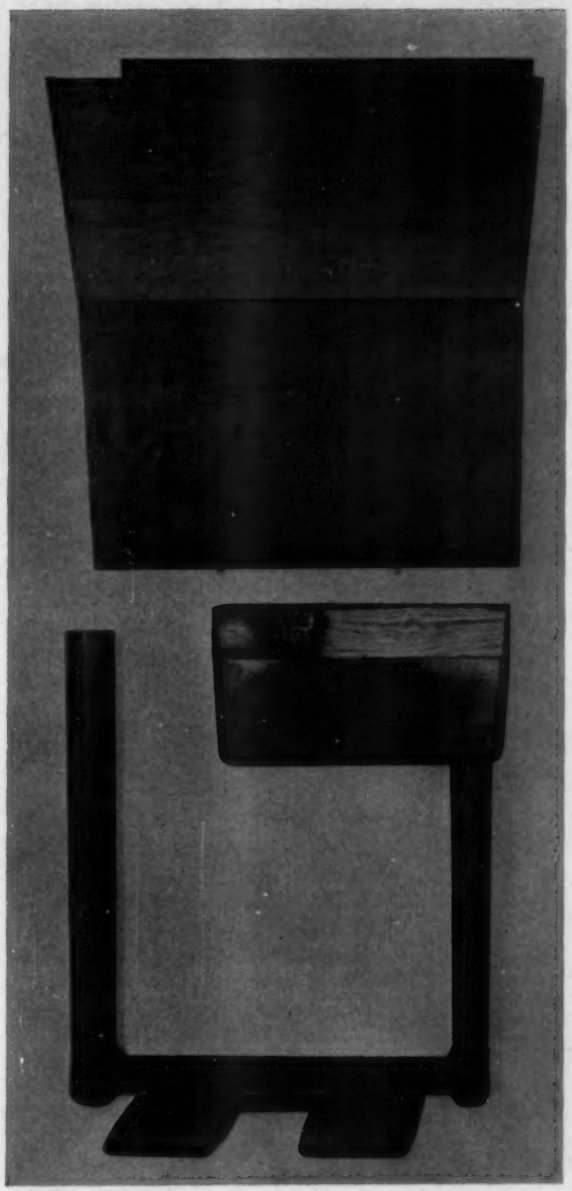


Fig. 7 — Assembled view of preformed gating and risering system used for wedge casting.



Fig. 8 — Front view of wedge casting.

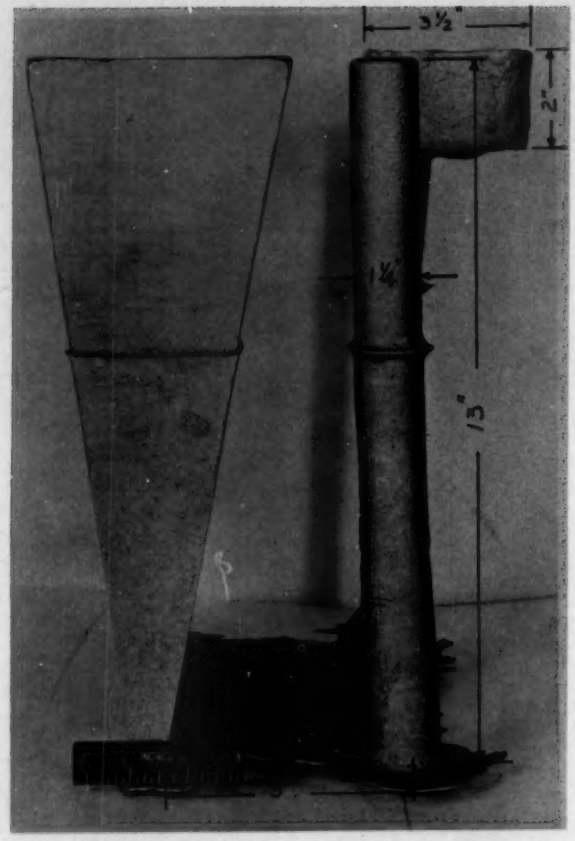



Fig. 9 — Side view of wedge casting and gating arrangement.

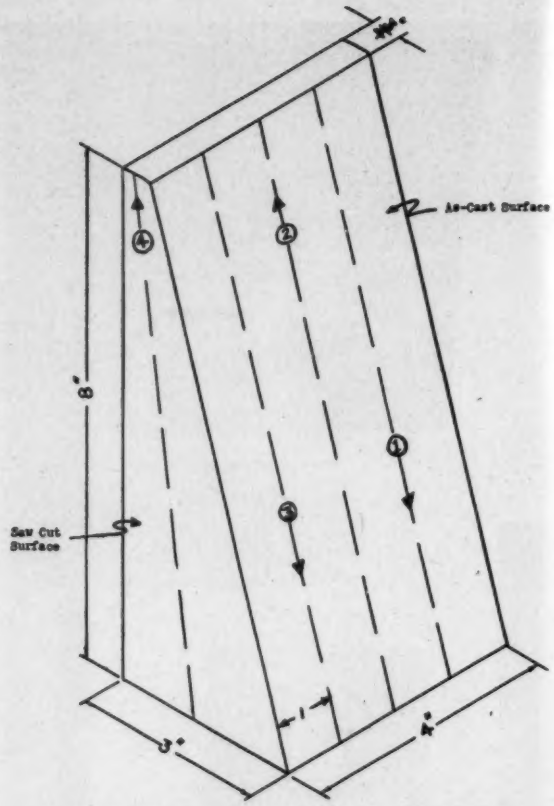



Fig. 10 — Sketch of position *A* of a wedge block casting used in surface weld bead tests. Labled lines show order and direction of bead application. All surfaces, except as otherwise noted, are as-cast.

| Specimen No. or Location | Test Temp., F |
|---|---|
| 1, 3, 7 | 0 |
| 0, 4, 9 | 75 |
| 2, 5, 6, 8 | 180 |

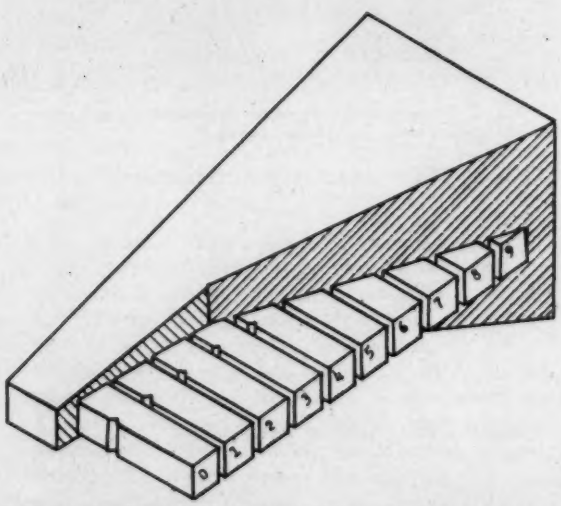


Fig. 11 — Sketch of position *C* of a wedge block casting used for Charpy V-notch specimens, showing relationship of notch and numerical identification to section thickness gradient.

It is to be noted that the specimens for test at a particular temperature were selected from staggered locations to take into account the variation in wedge thickness. The specimens were broken in a 220 ft-lb impact testing machine under a striking velocity of 18.1 fps.

### *Groove Welding*

Figure 12 is a sketch of the groove weld specimen made from the second wedge. One of the machined wedges prior to welding is shown in Fig. 13. It will be noted that the specimen provides for the influence of mass effect as well as progressively increased restraint as welding proceeds. The degree of restraint is considerably greater than involved in other related work reported in the literature, and is of the order of magnitude encountered in actual repair of propellers. This follows from the design of the wedge, in which the thickness variation is of the order of magnitude found in propellers, although the gradient is appreciable steeper.

The applicable welding details are as described for the weld-bead test. In the groove weld, no preheat was used, and the beads were deposited in a continuous manner except that deposition was interrupted when the base metal temperature was above 200 F. The completed weld specimens were cooled in still air and inspected by dye penetrant and radiography to evaluate soundness. They were next cut to provide

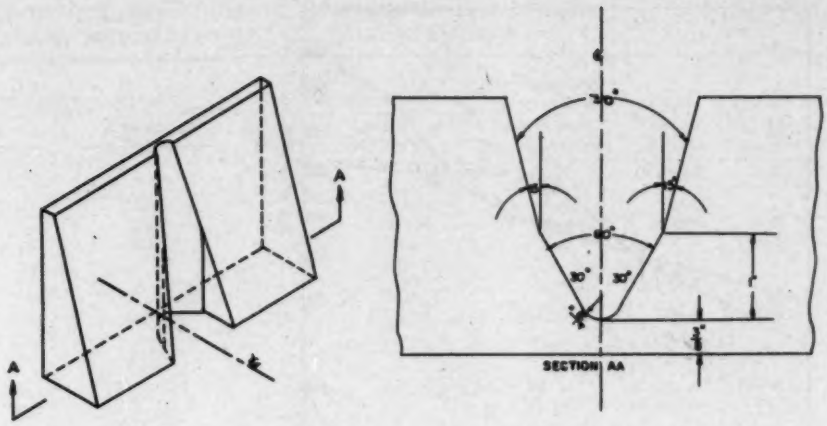



Fig. 12 — Sketch showing design of the groove used for groove weld test blocks.

flat tensile and bend specimens, as shown in Fig. 14a. To obtain these specimens, the reinforcement portion of the base metal was first cut off. In this manner, each specimen was cut so as to contain all weld metal in the section under test.

The tensile and bend specimens were prepared, as shown in Fig. 14b, in accordance with the applicable drawing of the current Federal Test Method Specification.[11] X-ray diffraction phase studies were made on samples of several heats to supplement metallographic examination, and to determine significant differences in phases and amounts which could be associated with degree of weldability. The details of the method employed will be submitted in the next progress report.

Subsequent to this work, three additional 1000 lb heats were cast using a foundry practice similar to that described previously, with the exception that a different flexing agent was used. The latter was a proprietary mix tried for the first time on this alloy type. The basic composition of these heats was that indicated in the work described below as being optimum insofar as major elements are concerned, i.e., 10 per cent Al, 4 per cent Fe, 5 per cent Ni and 0.5 per cent Mn.

One heat was used as a control, i.e., with no additions of Si and Pb. The second heat was subdivided into four portions with silicon at four levels ranging from 0 to 0.18 per cent. The third heat was also subdivided into four portions, and made to vary in Pb content from 0 to 0.10 per cent. For each composition, one wedge and one keel block was cast. Each wedge was prepared for groove welding as previously described.

Experience gained in the preliminary work on weld repair of Ni-Al bronze propellers showed that wherever castings contained gross internal defects cracking occurred on welding. The presence of even fine stringers of dross at the surface gave similarly bad results on welding. To assist in subsequent interpretation of weldability data, it was therefore decided to use radiography (radium) and dye penetrant tests to check the overall soundness of the wedge castings before and after groove welding.

It was not anticipated that radium radiography of the relatively thick section would reveal more than major discontinuities in the unsectioned wedges (cast or welded). On the other hand, it was expected that penetrant inspection would be helpful in highlighting dross and fine cracks coming to the surface on the welded wedges as a whole, and in the course of examination of sections prepared for more detailed studies.



Fig. 13 — Groove weld test specimen.

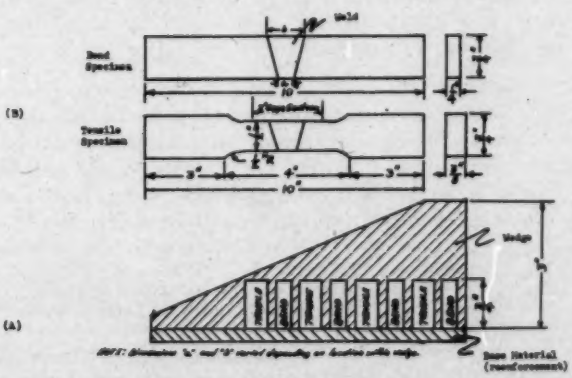



Fig. 14 — Sketch showing location and dimensions of tension and bend specimens taken from groove welded test block.

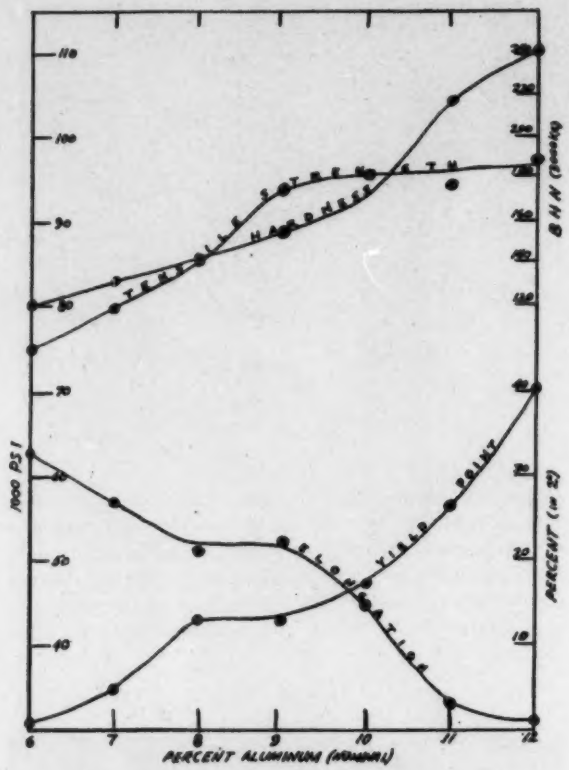



**Fig. 15 — Aluminum content effect on mechanical properties of as-cast nickel aluminum bronze (5 per cent Ni, 4 per cent Fe and 0.5 per cent Mn). Yield point determined by divider method using 2 in. gage.**

## RESULTS

Figure 15 is a plot of mechanical properties (tensile and hardness) vs. aluminum content for separately cast test coupons, based on the data given in Table 2. In general, as the aluminum content increases, the strength and hardness tend to increase while the elongation decreases. It is to be noted, however, that the plots of tensile strength, yield point and elongation show sharp changes in slope between approximately 8 to 10 per cent Al.

The relationship between hardness and tensile strength is shown in Fig. 16. It will be noted that there is a linear increase in strength up to a Bhn (at 3000 kg) of approximately 165, after which fur-

**TABLE 3 — V-NOTCH CHARPY VALUES VS. ALUMINUM CONTENT IN AS-CAST Ni-Al BRONZE**

| Heat No. | Al, % | 0 F Sample No. | 0 F Ft-Lb | 75 F Sample No. | 75 F Ft-Lb | 180 F Sample No. | 180 F Ft-Lb |
|---|---|---|---|---|---|---|---|
| | | 1 | 49 | 0 | 47 | 2 | 44 |
| 1 | 6 | 3 | 51 | 4 | 45 | 5 | 49 |
| | | 7 | 53 | 9 | 52 | 6 | 50 |
| | | | | | | 8 | 51 |
| | Avg. | | 51 | | 48 | | 49 |
| | | 1 | 43 | 0 | 42 | 2 | 40 |
| 2 | 7 | 3 | 43 | 4 | 42 | 5 | 46 |
| | | 7 | 49 | 9 | 47 | 6 | 48 |
| | | | | | | 8 | 47 |
| | Avg. | | 45 | | 44 | | 45 |
| | | 1 | 23 | 0 | 25 | 2 | 25 |
| 3 | 8 | 3 | 24 | 4 | 26 | 5 | 24 |
| | | 7 | 23 | 9 | 25 | 6 | 26 |
| | | | | | | 8 | 26 |
| | Avg. | | 23 | | 25 | | 25 |
| | | 1 | 18 | 0 | 19 | 2 | 20 |
| 4 | 9 | 3 | 18 | 4 | 19 | 5 | 20 |
| | | 7 | 17 | 9 | 27 | 6 | 17 |
| | | | | | | 8 | 19 |
| | Avg. | | 17 | | 22 | | 19 |
| | | 1 | 18 | 0 | 19 | 2 | 21 |
| 5 | 10 | 3 | 17 | 4 | 18 | 5 | 19 |
| | | 7 | 17 | 9 | 17 | 6 | 19 |
| | | | | | | 8 | 19 |
| | Avg. | | 17 | | 18 | | 20 |
| | | 1 | 12 | 0 | 13 | 2 | 13 |
| 6 | 11 | 3 | 8 | 4 | 10 | 5 | 10 |
| | | 7 | 10 | 9 | 9 | 6 | — |
| | | | | | | 8 | 9 |
| | Avg. | | 10 | | 11 | | 11 |
| | | 1 | 2 | 0 | 3 | 2 | 2 |
| 7 | 12 | 3 | 2 | 4 | 3 | 5 | 2 |
| | | 7 | 3 | 9 | 3 | 6 | 4 |
| | | | | | | 8 | 3 |
| | Avg. | | 2 | | 3 | | 3 |

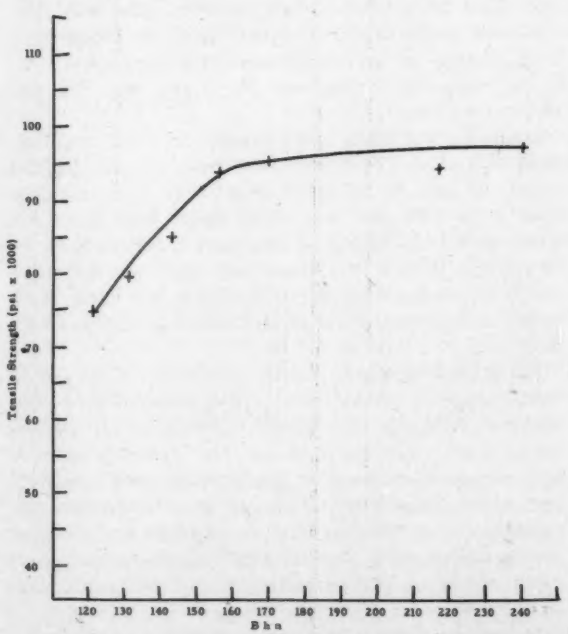



**Fig. 16 — Brinell hardness number vs. tensile strength for as-cast Ni-Al bronze (Base on data of Table 2 and Fig. 15).**

TABLE 4 — EFFECT OF ALUMINUM ON MECHANICAL PROPERTIES OF AS-CAST WELDED NICKEL ALUMINUM BRONZE

| Heat No. | Al, % | Tension Tests on Welded Specimens: Y.S.,[5] psi | T.S., psi | Elong., % on 2 in.[2] | Location of Bar Fracture | Bend Tests on Welded Specimens: Bend degrees[4] | Location of Bar Fracture | Dye Penetrant Inspection of Tensile and Bend Specimens | Weld Bead Test | Overall Weldability |
|---|---|---|---|---|---|---|---|---|---|---|
| 1 | 6[3] | — | — | — | — | — | — | Cracked | Cracked | NG |
| 2 | 7 | —<br>15,800<br>15,000[1]<br>17,100 | 12,700<br>16,300<br>15,500[1]<br>18,700 | —<br>—<br>—<br>— | Heat Affected Zone | 13.0 | Heat Affected Zone | Cracked | Cracked | NG |
| 3 | 8 | 37,600<br>35,300[1]<br>36,000 | 41,700<br>37,400[1]<br>36,200 | 3.0<br>3.0<br>2.0 | Heat Affected Zone | 33.0 | Heat Affected Zone | Cracked | No Cracks | NG |
| 4 | 9 | 45,600<br>47,400 | 63,700<br>55,700 | 3.0<br>1.5 | Heat Affected Zone | 47.0 | Weld | No Cracks | No Cracks | OK |
| 5 | 10 | 42,600<br>42,500<br>43,400 | 70,400<br>74,100<br>75,600 | 6.0<br>6.5<br>6.0 | Weld<br>Weld<br>Base Metal | 58.0 | Weld | No Cracks | No Cracks | Optimum |
| 6 | 11 | 46,000<br>47,400<br>48,700 | 70,600<br>76,400<br>74,000 | 3.0<br>2.0<br>3.0 | Weld<br>Base Metal<br>Weld | 26.6<br>27.6 | Weld<br>Fusion Zone | No Cracks | No Cracks | OK |
| 7 | 12[6] | — | — | — | — | — | — | — | No Cracks | NG |

NOTES: 1. Low values due to unsound specimens.
2. Values not reliable due to difficulty in properly meshing fractures.
3. No specimens could be obtained due to excessive cracking along fusion zone during welding.
4. The angle of bend is considered as the deflection from the horizontal when fracture occurred.
5. 0.5 per cent Extension under load (2 in. gage length).
6. No specimens could be obtained due to excessive cracking of low ductility, high hardness base metal during welding.

ther increases in hardness yields practically no significant increase in strength. This suggests that up to a hardness of about 165 Bhn the tensile strength of the alloy may be estimated from the hardness, regardless of composition.

The effect of aluminum content on the microstructure of nickel-aluminum bronze is shown in Fig. 17. At the lower levels of Al, the structure appears to consist essentially of a single phase matrix with small amounts of precipitated intermetallic constituent. As the Al content increases, a two phase polygonal structure develops, and at levels of over about 10.5 per cent the structure assumes a martensitic form (similar to that found in steels).

Table 3 lists Charpy V-notch data as a function of Al content, for temperatures of test and sample position in the wedge casting as noted. The effect of the original section thickness, as determined by location of the Charpy V-notch specimen, for several levels of aluminum, is shown in Fig. 18. It is to be noted that the aluminum content has a marked effect on the absorbed energy, but for any given level of aluminum specimen location or test temperature have no significant effect on toughness.

### *Aluminum Content Effect*

The effect of aluminum content on Charpy V-notch properties, regardless of sample location or temperature of test, is shown in Fig. 19. From this plot it can be seen that as the aluminum content increases there is a general decrease in energy value. Between 8 and 10 per cent aluminum there appears to be marked change in slope of the curve, which shows itself as a plateau at about 20 ft-lb, followed by a linear decline in energy absorbed above 10 per cent aluminum.

Table 4 contains a summary of weld-bead tests, and tensile and bend data for specimens machined from groove welded wedge castings at various levels of aluminum content. It should be noted that neither radiography nor dye penetrant inspection before welding revealed significant discontinuities in this series of wedges. The results of dye penetrant inspection of specimens cut from the wedges and comments on overall weldability are also included. The results show that the surface weld bead test is less discriminating than the groove weld test.

It will be noted that even the wedge materials which are considered weldable have somewhat lower tensile properties (as welded) than the corresponding separately cast tensile bars (Table 2). Among the weldable compositions shown the specimens representing the 9 and 10 per cent Al levels have the best ductility as evidenced by the angles of bend.

Figure 20 shows microstructures of welded samples with Al varying from 6 to 11 per cent, at original magnifications of 50 and 500 ×. At the lower aluminum content levels, namely from 6 to 8 per cent, it will be noted that cracks are visible in the base metal and appear to be intergranular. No cracks are visible in the samples containing 9 per cent aluminum and above for which, in addition, a better "knitting" of the structure is indicated in the fusion zone.

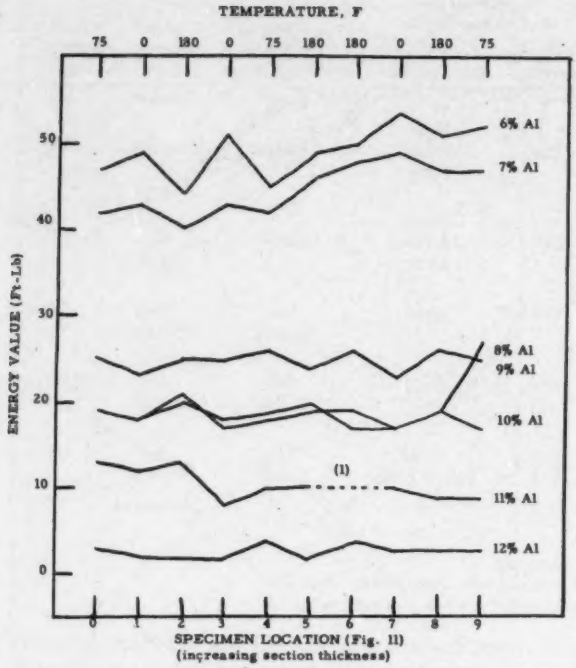



**Fig. 18 — Test temperature and section size effect on the Charpy V-notch properties of as-cast Ni-Al bronze (note: (1) — defective specimen, not tested).**

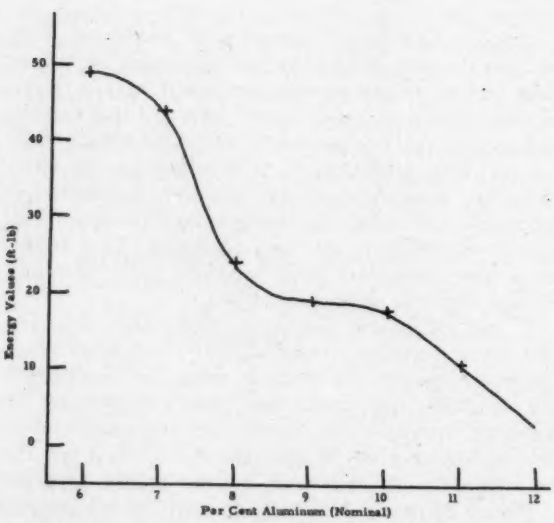



**Fig. 19 — Aluminum content effect on Charpy V-notch properties of as-cast Ni-Al bronze (points indicated are averages).**

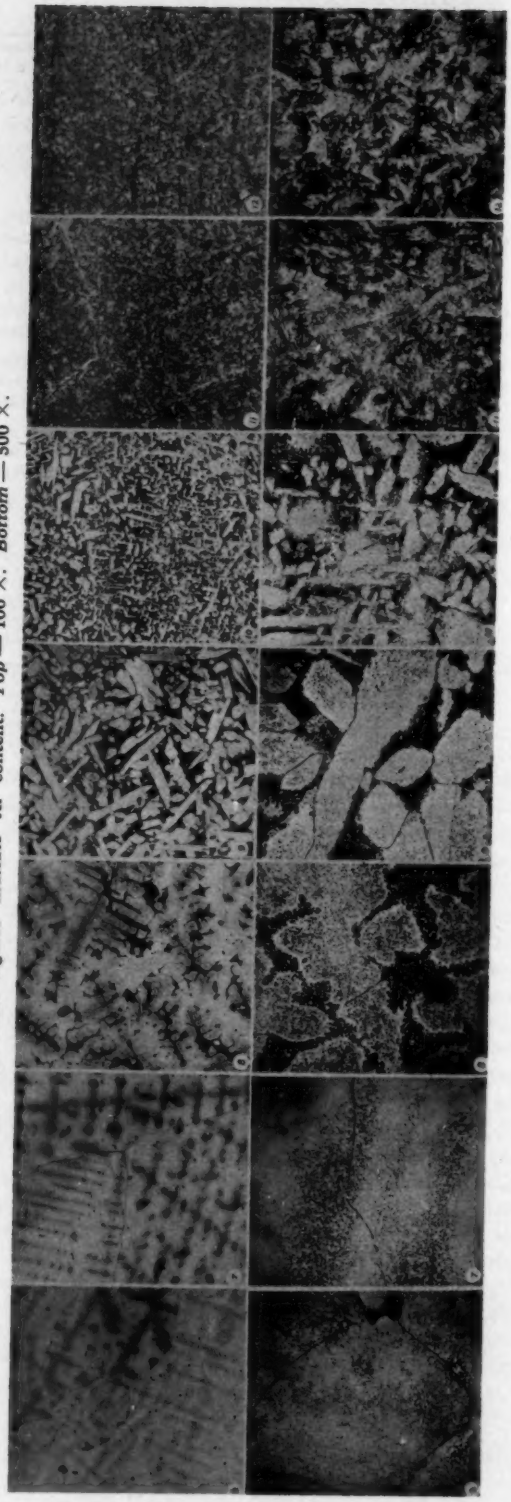


**Fig. 17 — Typical as-cast Ni-Al bronze microstructure with increasing aluminum content. Inset figures indicate Al content. *Top* — 100 ×. *Bottom* — 500 ×.**

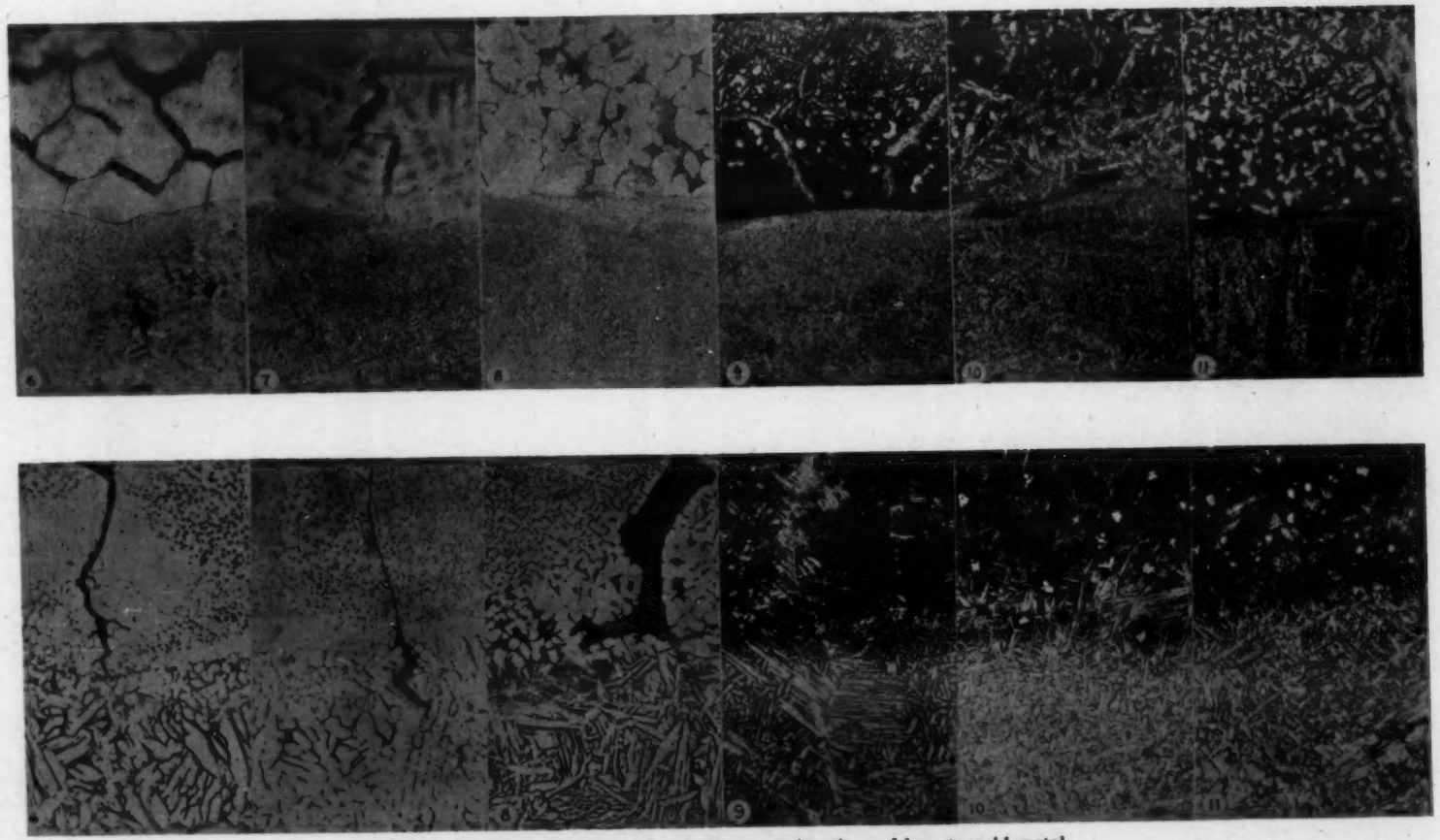


Fig. 20 — Representative microstructures at junctions of base-to-weld metal for as-cast Ni-Al bronze compositions varying in Al content. Per cent Al shown in inset figures. *Top* — 50 ×. *Bottom* — 500 ×. Reduced ½ photographically. Weld metal is at lower half in each photomicrograph.

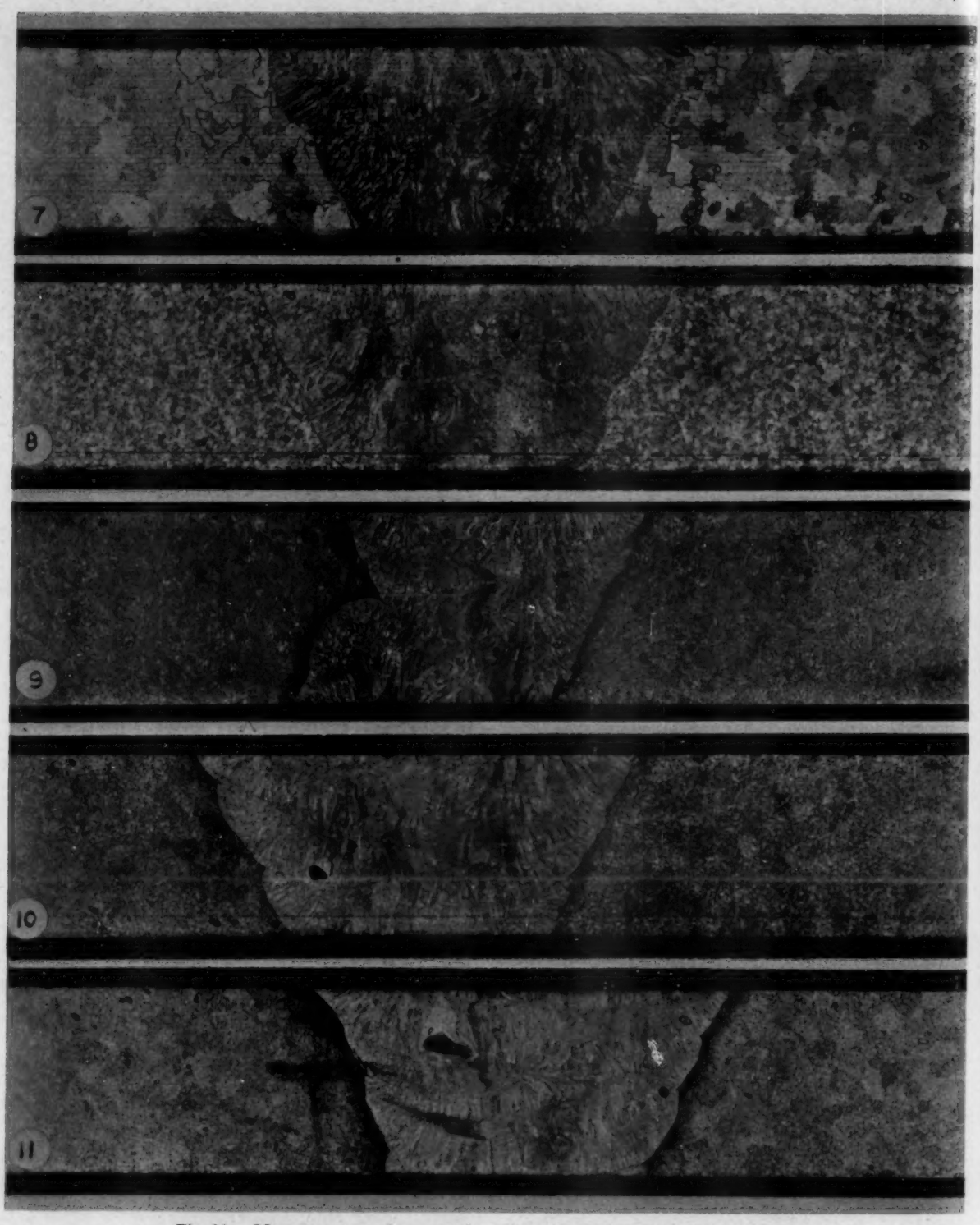



Fig. 21 — Macrostructures of gage sections of tensile specimens taken from groove welded castings (per cent Al shown by inset figures) enlarged 2½ times.

Macrostructures of the gage sections of some of the welded tensile specimens are shown in Fig. 21 for specimens varying in Al content from 7 to 11 per cent, as indicated on the macros. It is to be observed that the 7 per cent Al sample shows numerous cracks in the base metal. Similar finer cracks, but less extensive, were observed in the base metal adjacent to the fusion zone in the 8 per cent Al sample, but these are not readily visible in the photomacrograph. For actual evaluation, dye penetrant inspection was used to facilitate crack detection in the macrosections, and trends were further checked by microscopic examination.

Most specimens contain gas pockets in the weld metal in varying degrees of severity. It is interesting to note that the grain size of the 8 per cent Al alloy is significantly smaller than that of the others. The macro for the specimen representing the 9 per cent Al alloy shows an irregular weld area, which was due to unsymmetrical backchipping and rewelding. The latter procedure was sometimes resorted to in the preparation of the groove weld specimens, in line with accepted production techniques, when the thin base metal backing material cracked because of mechanical restraint.

### *Pb and Si Effect*

The effects of additions of Pb and Si to a weldable composition on subsequent weldability are shown in Table 5. Radiography of the wedges used in these tests showed excessive dross, which is considered to be due to the use of a previously untried flux. The weldability of this test material was therefore judged solely on the basis of dye penetrant indications and radiography applied to the wedges after welding as a whole. These data suggest that weldable structures are possible with Pb and Si as high as 0.10 and 0.18 per cent, respectively. These indications were confirmed by welders who reported that they encountered no trouble in the course of welding at these levels of so-called injurious elements.

TABLE 5 — Pb AND Si EFFECT ON WELDABILITY OF AS-CAST NICKEL ALUMINUM BRONZE (GROOVE WELDED SPECIMENS)

| Heat No. | | Lead | Silicon | Remarks |
|---|---|---|---|---|
| 13 | | — | — | Weldable |
| | A | — | 0.04 | Weldable |
| | B | — | 0.08 | Weldable[1] |
| 14 | | | | |
| | C | — | 0.12 | Weldable[1] |
| | D | — | 0.18 | Weldable |
| | A | — | — | Weldable[1] |
| | B | 0.06 | — | Weldable |
| 15 | | | | |
| | C | 0.09 | — | Weldable |
| | D | 0.10 | — | Weldable |

Avg. Analysis

| | |
|---|---|
| Cu | 80.5 |
| Al | 10.2 |
| Ni | 4.7 |
| Fe | 3.6 |
| Mn | 0.47 |

1. Dye Penetrant revealed small surface cracks in the thin section of base metal back-up material.

## DISCUSSION

Development of as-cast Ni-Al bronze compositions which yield satisfactory mechanical properties and consistent weldability requires determination of (a) the effect of major constituents on the mechanical properties, (b) the effect of composition on the microstructure, (c) the relationship of microstructure to weldability and (d) the effects of interaction, if any.

This portion of the investigation was limited to the study of the effects of varying amounts of aluminum (keeping the other specified elements at a substantially constant level within requirements of a reference[10]) on the mechanical properties, the resultant microstructures, and weldability. Lead and silicon were also considered, since impurities determine the cost and the acceptability of ingot material, and since their effects on weldability have raised many questions in the field.

This initial approach was taken as a result of previous work,[3,7] which had indicated that aluminum content was of prime importance in this alloy. Studies of the effects of the other major elements, i.e., nickel and iron, were planned as the next phase, using the optimum Al composition developed in the earlier work. Wedge block castings incorporating variations in Ni and Fe have been produced, but the balance of the work is not scheduled for continuation at present.

### *As-Cast Microstructures*

The photomicrographs, shown in Fig. 17, may be divided into four distinct groups on the basis of the relative change in appearance of the structures, as the aluminum content increases from 6 to 12 per cent. These groups are:

| Group | Aluminum, % |
|---|---|
| I | 6 to 7 |
| II | 8 |
| III | 9 to 10 |
| IV | 11 to 12 |

**Group I.** The microstructures, for the 6 and 7 per cent aluminum levels are basically similar and contain a matrix composed of large alpha grains. The alpha phase is identified by x-ray crystallography as a random face-centered cubic lattice structure. Within the $\alpha$ grains and in the grain boundaries, there appear small amounts of a $\kappa$ phase which is characterized by x-ray crystallography, as an ordered body-centered cubic lattice structure. This phase is composed of complexes of nickel, iron and aluminum and provides strength to the alpha matrix which is ductile and lacking in strength.

The combination of the two phases alpha and kappa in varying amounts and in orientation, will accordingly, yield a wide range of mechanical properties. At this level of aluminum, the kappa within the grains precipitates in a dendritic pattern with individual particles in globular form as revealed at higher magnification (Fig. 17). Between grains, the kappa is arranged in rod-like stringers which follow the grain boundary contours.

**Group II.** At 8 per cent aluminum, the microstructure shows an appreciable increase in the amount of kappa with a corresponding decrease in the primary alpha. Within the grains, the dendritic pattern of kappa is still evident at low magnification. The kappa in the dendritic skeleton appears largely as a precipitate of polygonal-shaped agglomorates within pure alpha. The kappa in these agglomorates is in the form of globules. The thin, rod-like kappa in the grain boundaries has greatly diminished in amount. The kappa within the alpha grains has increased in amount and is still in the form of a fine globular dispersion.

**Group III.** At 9 per cent aluminum, the appearance of the microstructure changes radically. The alpha matrix is now in the form of elongated finger-like grains in a Widmanstatten pattern. The kappa phase is still present in its fine spheroidal condition within the alpha grains. The grain boundaries now consist of transformation products of high temperature beta, similar in appearance to pearlitic structure in steel. The transformation products from the beta phase are secondary alpha with kappa in lamellar form (eutectoid). The rod-like formation of the kappa constituent coming from the primary alpha phase is still present, although in smaller amounts of shorter stringers.

In addition, the appearance of light gray spheroids was now observed. These spheroids are tentatively considered to be delta phase, which has been identified in samples containing 9 per cent Al by x-ray crystallographic work at the Laboratory. It is to be noted that this structure is not readily identified metallographically, and then only by tedious and differential etching techniques described in the literature.[3,12] Additional work planned to correlate more conclusively the optical and x-ray diffraction results on this point has not been started.

The microstructure of the 10 per cent aluminum alloy is basically similar to the 9 per cent alloy structure. The increase in aluminum content is accompanied by a decrease in the amount of alpha, and a corresponding increase of the eutectoid composition. The Widmanstatten pattern is still prevalent, and the finger-like grains of alpha are smaller and more randomly scattered. The primary alpha grains contain some precipitated kappa. Secondary alpha precipitate appears at the grain boundaries. In addition, the alpha plus kappa lamellar structure has increased in amount and the globules (or rosettes) of delta are now larger.

**Group IV.** At the 11 and 12 per cent aluminum levels, the microstructure changes from an essentially alpha matrix (Group I) to a matrix of beta transformation products and retained high temperature beta constituent. In contrast to the microstructures of the lower aluminum level alloys, the 11 and 12 per cent aluminum types show little alpha phase, most of which is secondary alpha located at the grain boundaries. The amount of eutectoid composition appears to be at a maximum in the 11 per cent alloy, whereas the 12 per cent alloy has more retained beta which incidentally, is a hard and brittle constituent. Both alloy type structures contain rosettes or cubicles of delta. The overall structure resembles martensite.

### *Mechanical Properties vs. Microstructure*

A study of the microstructure and mechanical properties indicates that there is a rational relationship between the two. The discussion which follows applies to microstructures and mechanical properties of separately cast tensile bars.

At low levels of aluminum, as the aluminum content increases the amount of kappa constituent increases, thereby increasing hardness and strength and reducing elongation. The curves (Fig. 15) for these properties vs. aluminum content bear out this explanation for the 7 and 8 per cent aluminum alloys. Between 8 and 9 per cent aluminum, the tensile strength and elongation curves exhibit significant changes in slope. On increasing the aluminum content beyond 8 per cent, the tensile strength continues to increase more rapidly but the elongation remains approximately constant, instead of dropping as might be expected from the usual relationship between strength and ductility.

This condition can be explained by careful study of the microstructure. It is considered to be probably due to the distribution and orientation of the kappa phase. In the 8 per cent Al alloy, the kappa is formed from the super-saturated primary alpha phase whereas in the 9 per cent Al alloy the kappa originates from the beta which transforms into lamellar plates of alpha plus kappa eutectoid. The ductility is not affected by the increase in kappa for the 9 per cent aluminum alloy (compared to the increase in tensile and hardness), because in this alloy the distribution of the phases is more random and hence more favorable to ductility. Above 9 per cent aluminum, the elongation falls off rapidly due to the increased retention of the hard and brittle beta constituent.

Figure 15 indicates a continued increase in hardness with a leveling off of tensile strength as the aluminum content increases from 9 to 12 per cent. That this increase in hardness takes place not only at the expense of ductility but of toughness as well, is shown by a plot of Charpy V-notch energy vs. per cent aluminum, Fig. 19. The approach to a practically constant tensile strength above 9 per cent aluminum is probably due to the formation of secondary alpha at the grain boundaries, which affect the strength more than the increasing amount of harder constituent within grains. The continued increase in hardness can be explained by the fact that resistance to indentation is a direct function of the hard material within the area of contact involved, which is increasing with aluminum content.

Aluminum content has a pronounced effect on the Charpy V-notch properties of nickel-aluminum bronze. The trend indicated for the Charpy data, Fig. 19, closely parallels that indicated by the plot for elongation, Fig. 15. In both plots, there appears to be a leveling off of the curves near the 9 per cent alumi-

num level. Figure 19 suggests nickel-aluminum bronze is relatively notch-insensitive to variations in Al content in the range of 8.5 to 10 per cent. While variation of aluminum within the range noted produces changes in geometrical size of constituents, these changes affect the elongation and Charpy value little.

Beyond 10 per cent aluminum, the toughness declines rapidly due to an increase in the brittle beta constituent. From Fig. 18 it can be seen that for any given aluminum content the toughness is practically independent of section size, from which the specimen was taken, or temperature at which specimen was tested, i.e., ½ to 3½-in. and 0 to 180 F, respectively.

### *Microstructure and Weldability*

This discussion applies to specimens sectioned from groove welded wedge castings, which, according to Table 4, are more discriminating than the surface weld-bead test with regard to weldability. From Fig. 20, it can be seen that intergranular cracking of the base metal occurs for Group I (6 and 7 per cent Al) and Group II (8 per cent Al) alloys, the latter having less extensive cracking. It is interesting to note that cracking takes place along the thin rod-like kappa constituent, which is precipitated by the supersaturated alpha phase at the grain boundaries.

The tendency for cracking to occur in the grain boundaries is enhanced by the fact that the kappa phase is (a) more brittle than alpha, (b) essentially continuous and (c) oriented with respect to other phases so as to provide paths of least resistance to stress concentrations. At high aluminum contents, 9 to 11 per cent, the photomicrographs show a sound and complete "knitting" of the weld metal to the base metal. With increasing Al content, there is a "transition" in the region between 8 and 9 per cent, at which the base metal will bond properly with the weld metal and the susceptibility of cracking is lessened. The significantly smaller grain size observed for the 8 per cent Al macrospecimen (Fig. 21) compared to all others, appears to confirm that such a "transition" exists.

We therefore see that the low Al, nonweldable, alloys contain primarily alpha and kappa, the latter precipitating in thin, essentially continuous stringers at grain boundaries from a supersaturated alpha solution. On the other hand, the higher Al, weldable, alloys contain alpha plus kappa (beta transformation products) with retained beta and some delta.

While the alpha plus kappa alloys are superior to the alpha-kappa (eutectoid) plus beta alloys, as far as room temperature ductility is concerned, the situation is reversed at the temperatures of welding. In other words, the alpha plus kappa alloys are hot short and have a tendency to crack while the alpha-kappa (eutectoid) plus beta alloys show good hot ductility, which is necessary for successful welding. Under ideal and practically unattainable conditions it would be desirable to have an all beta structure at elevated temperatures, and an all alpha plus kappa structure at room temperature.[13]

For practical purposes, it follows then that for consistent weldability a composition must be selected which has enough beta at welding temperatures for hot ductility, and enough alpha plus kappa on cooling to room temperature for cold ductility. The composition which should be used as a compromise falls in the range of 9 to 10 per cent Al, because at the higher concentrations of Al, an excess amount of beta is retained thereby affecting ductility and toughness. For aluminum below 9 per cent, the amount of beta appears to be insufficient for successful welding.

The data on mechanical properties (representing the casting in the welded condition, Table 4) and consideration of its ultimate use as a propeller alloy, indicate that the 10 per cent aluminum alloy is optimum with respect to weldability, strength and ductility. A point to be noted is that the tensile strength of the 10 per cent aluminum welded (as-cast) specimen, is at least 75 per cent of the separately cast bar. This is an acceptable percentage in the foundry field, because it is generally recognized that the separately cast test bar properties are not wholly reflected in the casting.

### *Impurity Content vs. Weldability*

It will be noted from Table 5 that wedge block castings containing the "optimum" Al content (approximately 10 per cent) welded satisfactorily despite the fact that Si and Pb were as high as 0.18 and 0.10 per cent, respectively. This emphasizes the previously indicated importance of having the proper level of aluminum. These results were obtained in spite of the fact that radiography revealed a large amount of dross in the as-cast wedge blocks, which was probably due to the use of a previously untried flux.

Fortunately, in the machining of the groove, a major portion of the dross was removed leaving only small amounts in the back-up strip portion of the wedge block casting. It is in this portion that some fine cracks were revealed by dye penetrant inspection which were, however, not due to welding but rather to unsound and drossy metal. The excessive dross in the Si and Pb treated wedge test castings makes it impossible to correlate the effects of these impurity elements on the mechanical properties at this time.

New heats have been cast to provide sample material to check effects of impurity content on mechanical properties. Nevertheless, even though the welding tests were purely qualitative in nature, it is felt that the results tend to show that Pb and Si in the amounts indicated are not injurious to weldability.

The preliminary work,[7] which was the basis of this investigation, was concerned with inconsistent weldability of cast propeller patches with apparently similar chemical compositions. The results of the current investigation tend to confirm and explain facts disclosed by the earlier work. In this connection these points are considered salient:

1) The microstructures which are shown by the preliminary work to be "good" microstructures (high Al content) are similar in appearance to those which are considered weldable on the basis

of results of this investigation. This is evident from a comparison of the microstructure of Fig. 2a (from a reference,[7] and the microstructures for the 9 and 10 per cent Al samples, Fig. 17. Similarly, there is a comparison of nonweldable microstructures (low Al content), Fig. 2c (from a reference[7]) and those for the 6 and 7 per cent Al samples, Fig. 17.

2) Some microstructures of the material studied in the preliminary work, such as Figs. 2b and 2c, show markedly different structures although they apparently have similar chemical compositions. Based upon the current investigation, and especially on the study of microstructures, it is felt that Fig. 2b is more representative of 8 per cent aluminum while Fig. 2c suggests an aluminum content below 8 per cent. Therefore, the value of aluminum given in Table 1 (Item F1) appears to be doubtful, which may be due to the fact that it was obtained by difference. This difficulty was avoided in the current investigation by determining aluminum contents directly.
3) Based upon a limited, qualitative investigation of the effects of silicon and lead contents on weldability, it is felt that these elements, in the percentages as given in the preliminary work were not a cause for cracking.
4) It was felt on the basis of preliminary investigation, that improved weldability would generally be expected to accompany increased elongation due to low Al content. The current work shows that room temperature ductility is not a criterion of resistance to hot shortness at welding temperatures.

## CONCLUSIONS

The results of this phase of the investigation on as-cast Ni-Al bronze lead to:

1) The general effect of an increase in aluminum content is to increase strength and hardness and decrease the ductility and toughness.
2) Increasing aluminum content results in a change in microstructure from that consisting primarily of alpha plus kappa to structures containing alpha plus kappa (eutectoid), beta and delta.
3) An aluminum content of approximately 9 to 11 per cent imparts good hot ductility properties necessary for welding. Between 8 and 9 per cent aluminum, there is a transition where enough beta is formed for successful welding.
4) The optimum composition, taking into account weldability, strength, ductility and toughness appears to be at about 9.5 per cent aluminum when the Ni, Fe and Mn were held at 5, 4 and 0.5 per cent, respectively.
5) Elements such as silicon and lead appear to have no effect on the weldability characteristics of the optimum composition in amounts up to 0.18 and 0.10 per cent, respectively.
6) Microstructures can be used as an index to approximate the aluminum content and, hence, the relative weldability.
7) Room temperature ductility is a fair index of toughness properties of the alloy.
8) At high aluminum content levels, where retained beta is prevalent, an increase in hardness is accompanied by a loss in toughness and ductility.
9) Variations in aluminum content of the basic composition in the approximate range of 8.5 to 10 per cent results in virtually no effect on Charpy V-notch properties.
10) For the section sizes (½ to 3½-in. thickness) and temperatures of test (0 to 180 F) studied, Charpy V-notch properties show no significant variations.

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# SYSTEMATIC APPROACH TO SAND DESIGN AND CONTROL

## *Progress report 3 – The mulling effect*

by G. J. Vingas and A. H. Zrimsek


### ABSTRACT

**The third in a series of reports presenting basic data describing the effect of a number of variables on foundry sands. This report deals with the effect of mulling time and method of additions to a sand-clay-water system in relation to green compression strength, green shear strength, green tensile strength, ratio of green compression strength to green shear strength, density, dry compression strength and dry shear strength. The clays investigated are western and southern bentonite and fireclay.**


### INTRODUCTION

The two previous reports of this series, reports one and 2, showed conclusively that mulling must be considered a major variable in sand design and control. It is a major variable to the extent that it often overshadows clay type and clay content in new sand-clay-water systems. In actual foundry practice, however, castings of approximately equal quality are being produced in sands mulled for extensive periods, ones not mulled at all and all points in between. To help clarify the importance of mulling, this separate study on mulling as a major variable was initiated dealing with the effects of mulling on the physical properties of simple sand-clay-water systems.

### EXPERIMENT

As in previous reports, the base sand used is that described in the table. In the experiment, bentonite contents of 4.75, 7.45 and 10 per cent and fireclay contents of 10 and 15 per cent were used. Mulling of the various sand-clay-water mixtures was performed in three mullers of the following description:

1. 12 in. diameter laboratory muller with 1500 gram capacity, 46 rpm, 2 – 6 in. diameter x 1½-in. width wheels. Batch size for this experiment – 1250 grams.
2. 18 in. diameter laboratory muller with 5000 gram capacity, 40 rpm, 2 – 9 in. diameter x 2½-in. width wheels. Batch size for this experiment – 4500 grams.
3. 40 in. diameter production muller with 200 lb capacity, 48 rpm, 2 – 16 in. diameter x 3-in. width wheels. Batch size for this experiment – 150 lb.

Two methods of addition were employed in the experiment. In the first, sand and clay were dry mulled for 15 sec, after which water was added and the mixture wet mulled for periods of 2, 4, 6 or 12 min. In the second method, water was added to sand and distributed by a 15 sec mull, after which clay addition was made and mulling continued for 2, 4, 6 or 12 min periods.

With the variation in mullers, mulling time, clay content and type, methods of addition and water content, a total of 200 different mixtures were produced in the experiment for study.

The sands produced were discharged into airtight polyethylene bags and tested immediately for green compression, green shear and tensile strengths as well as rammed density at 1, 3, 5, 7 and 10 rams. Specimens for dry compression and dry shear were prepared at the same ramming energies and oven dried at 220 - 230 F for not less than 5 hr.

**SCREEN ANALYSIS OF BASE SAND USED**

| U.S. Standard Sieve No. | Retained, % |
|---|---|
| 20 | 0.0 |
| 30 | 2.6 |
| 40 | 19.3 |
| 50 | 30.1 |
| 70 | 24.4 |
| 100 | 15.1 |
| 140 | 5.6 |
| 200 | 2.4 |
| Pan | 0.6 |

### RESULTS

The results of the experiment are summarized graphically in Figs. 1 through 16. As in previous reports, graphical representation of data is employed. The large volume of data collected prohibits presentation in tabular form. Graphical presentation also illustrates more clearly the principles involved.

### DISCUSSION

Literature dealing with foundry sands seldom includes consideration of mulling as a major variable. Mulling is referred to simply as adequate, implying

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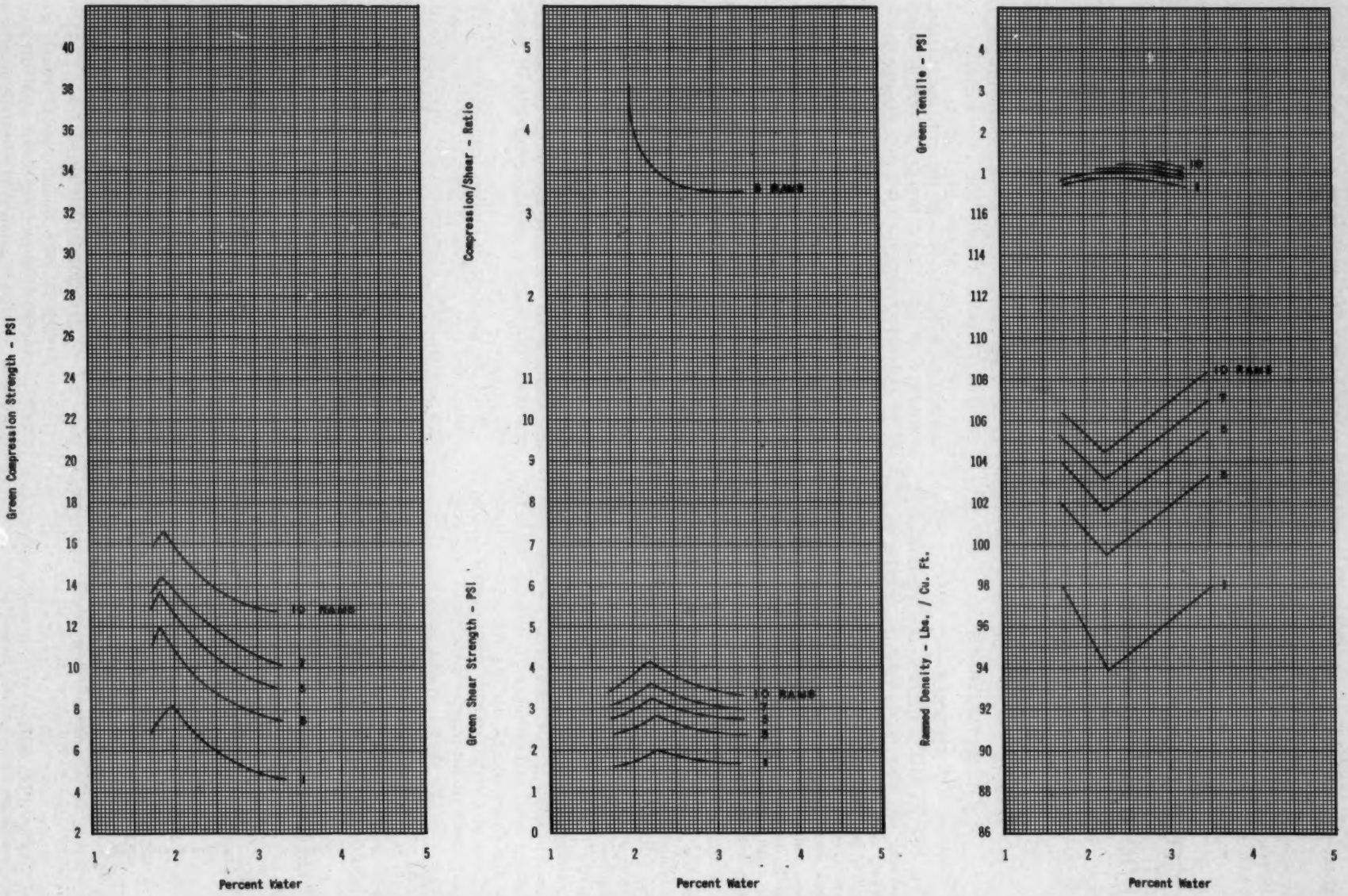



Fig. 1 — Water content and ramming energy effect on physical properties of 4.75 per cent western bentonite bonded Portage silica sand mulled 4 min. Method of addition: clay followed by water.

bentonite bonded Portage silica sand mulled 4 min. Method of addition: clay followed by water.

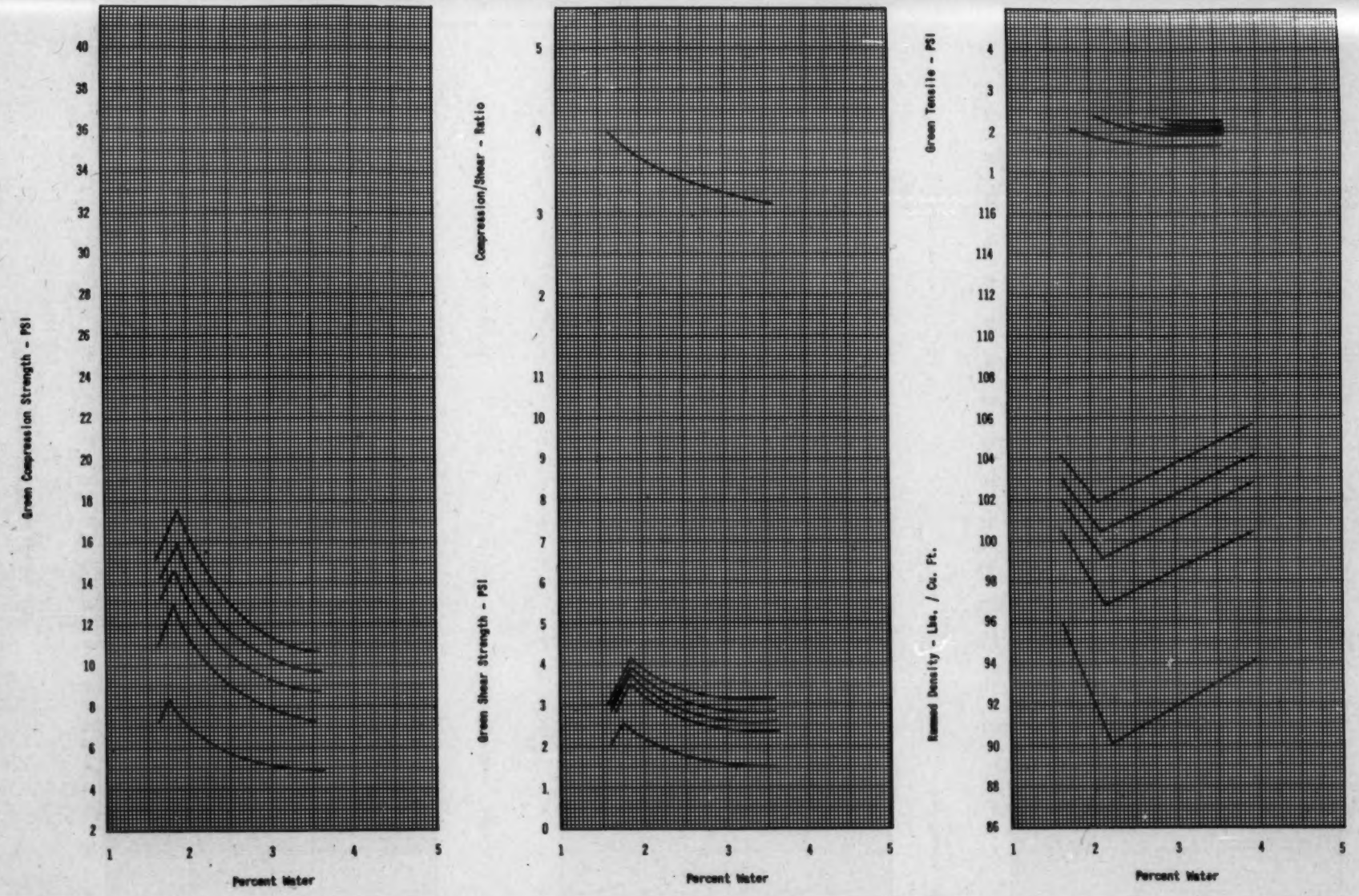



Fig. 2 — Water content and ramming energy effect on physical properties of 4.75 per cent western bentonite bonded Portage silica sand mulled 6 min. Method of addition: clay followed by water.

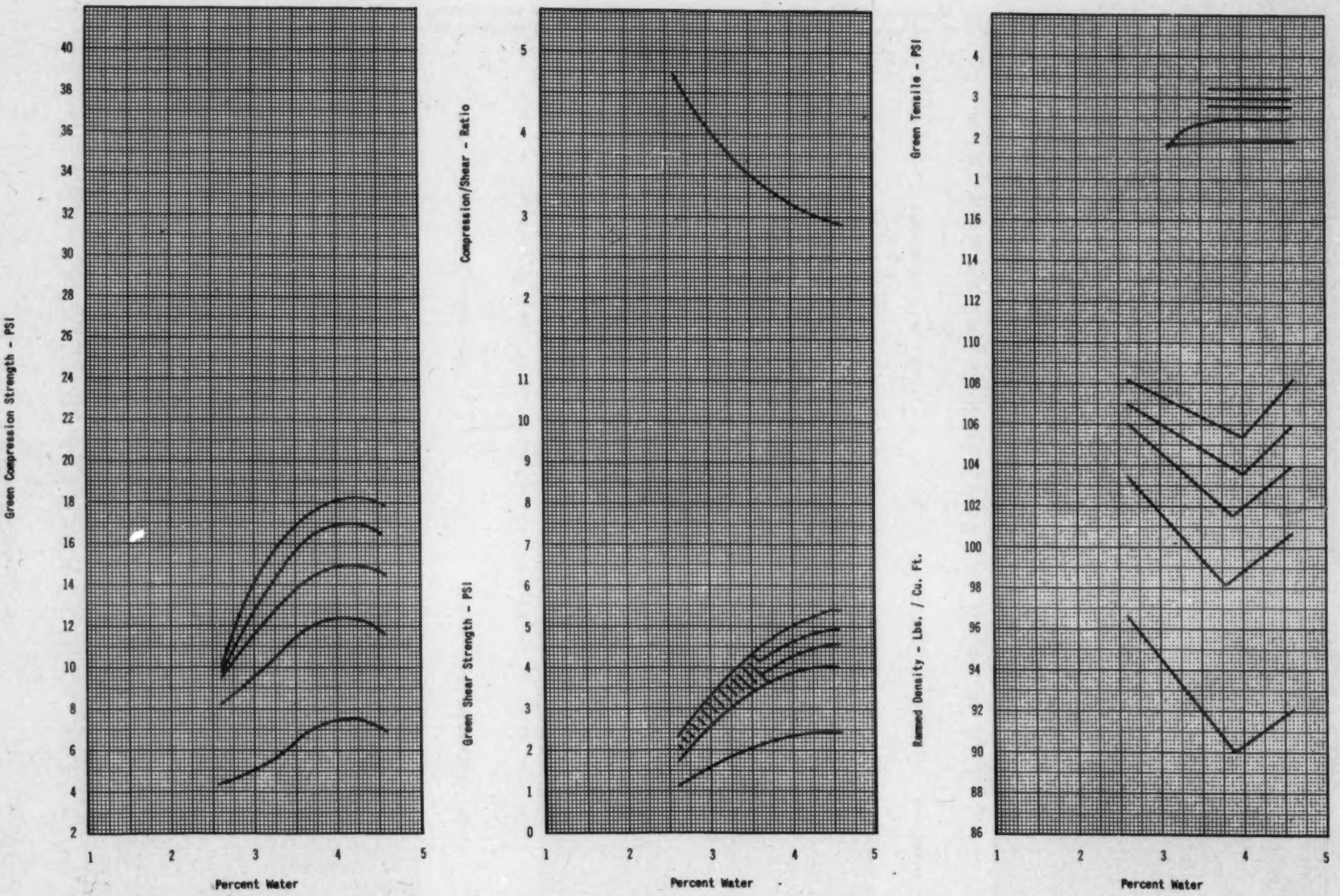



Fig. 3 — Water content and ramming energy effect on physical properties of 7.45 per cent western bentonite bonded Portage silica sand mulled 2 min. Method of addition: clay followed by water.

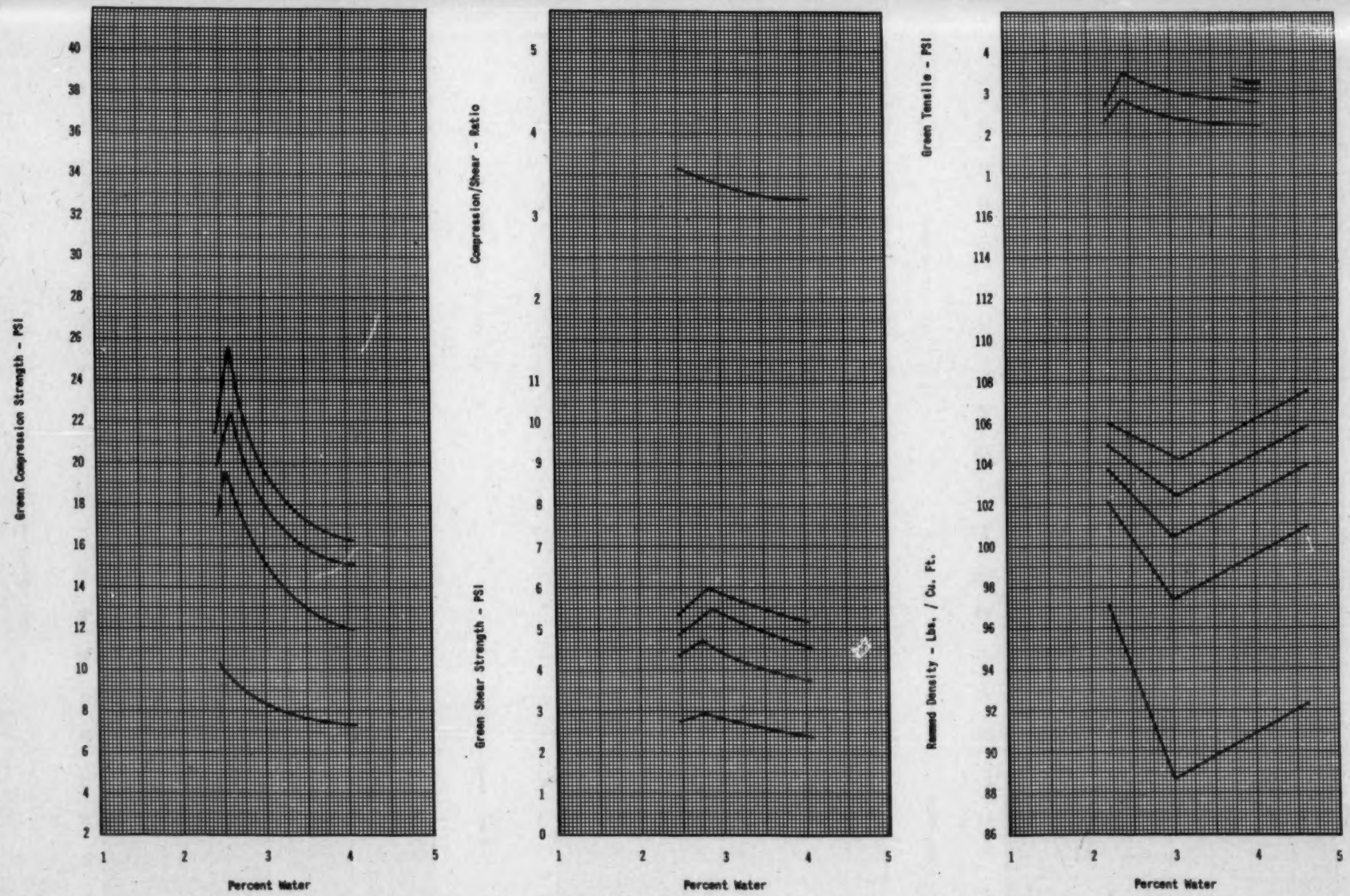



Fig. 4 — Water content and ramming energy effect on physical properties of 7.45 per cent western bentonite bonded Portage silica sand mulled 6 min. Method of addition: clay followed by water.

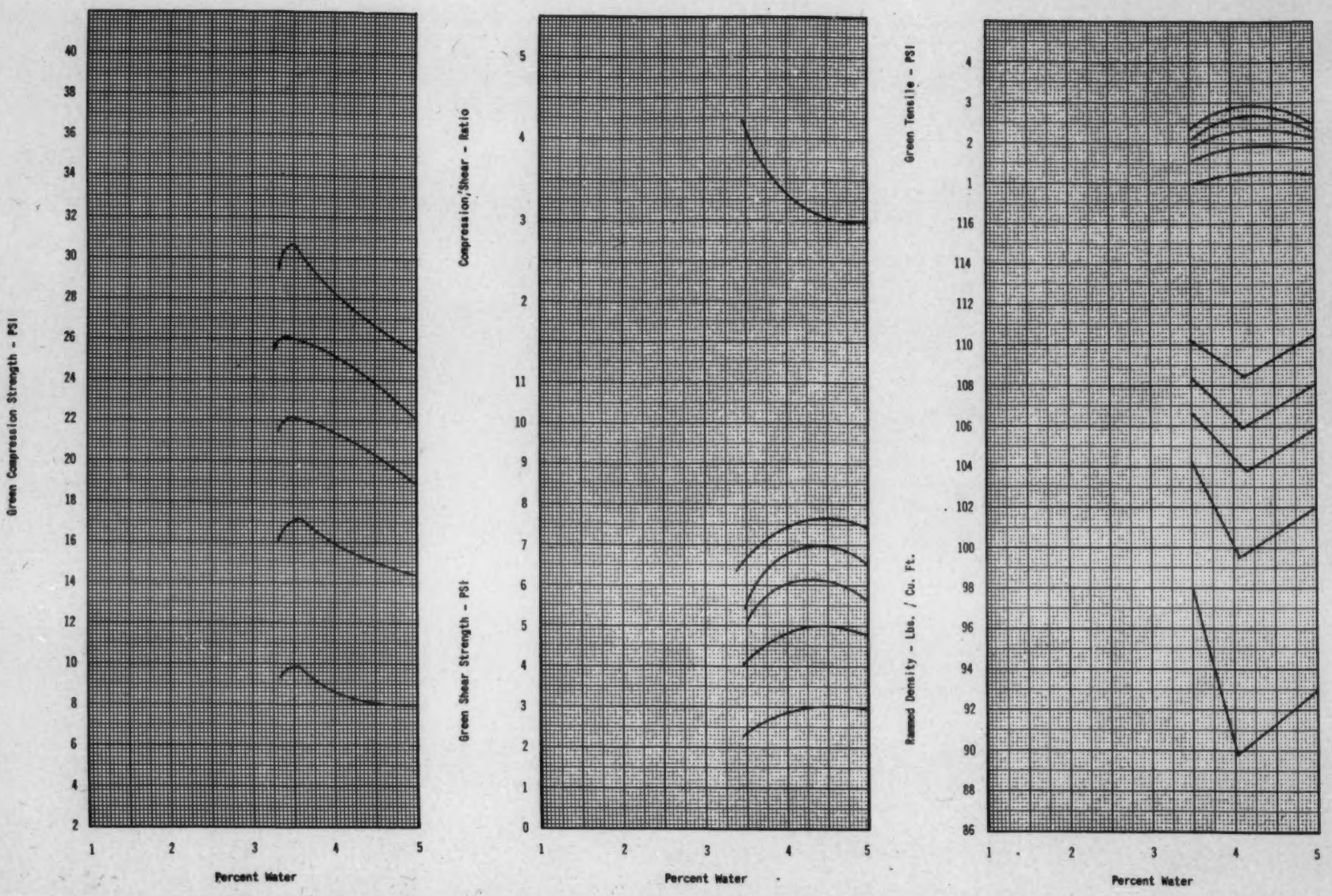



Fig. 5 — Water content and ramming energy effect on physical properties of 10 per cent western bentonite bonded Portage silica sand mulled 4 min. Method of addition: clay followed by water.

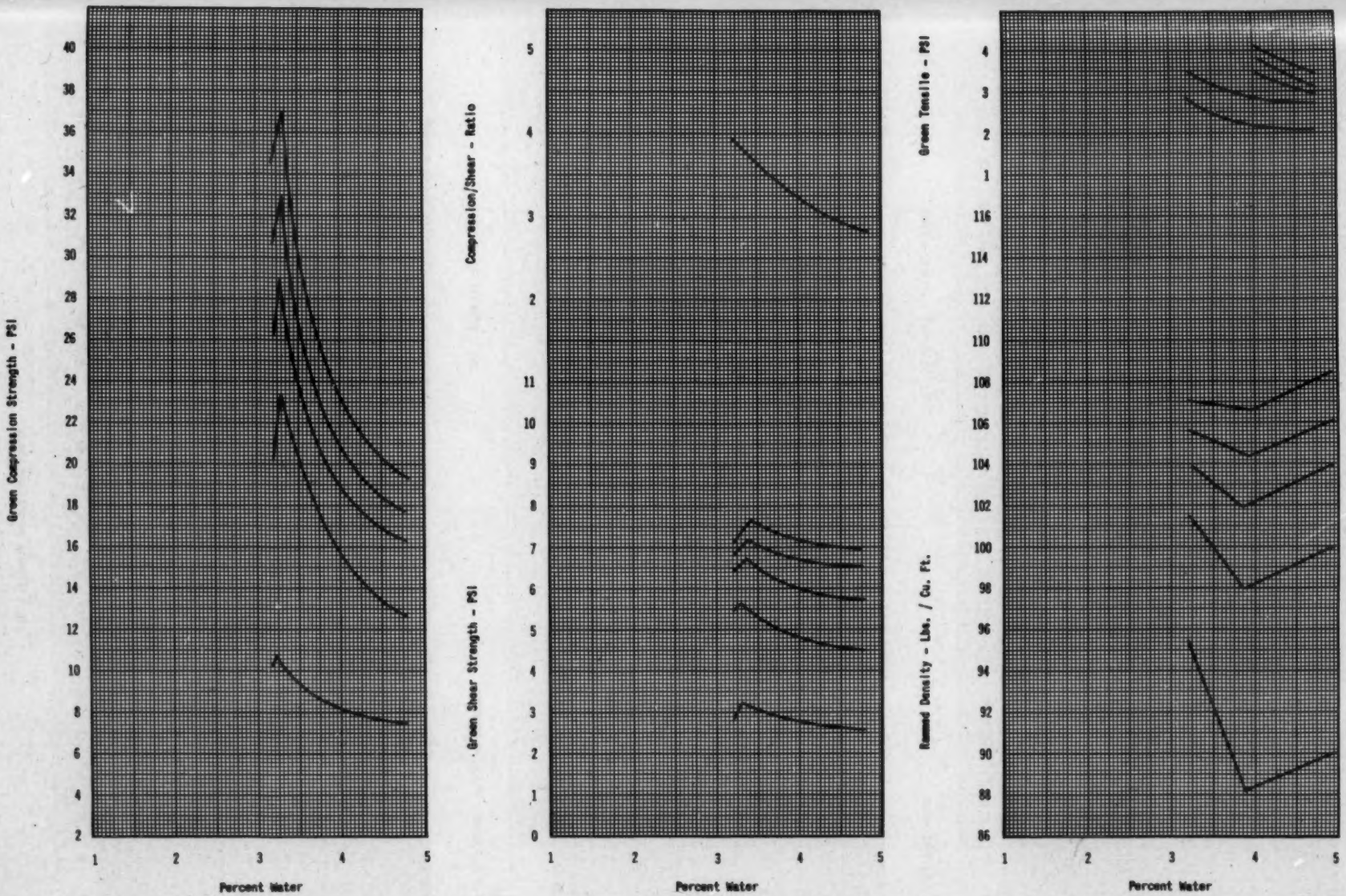



Fig. 6 — Water content and ramming energy effect on physical properties of 10 per cent western bentonite bonded Portage silica sand mulled 6 min. Method of addition: clay followed by water.

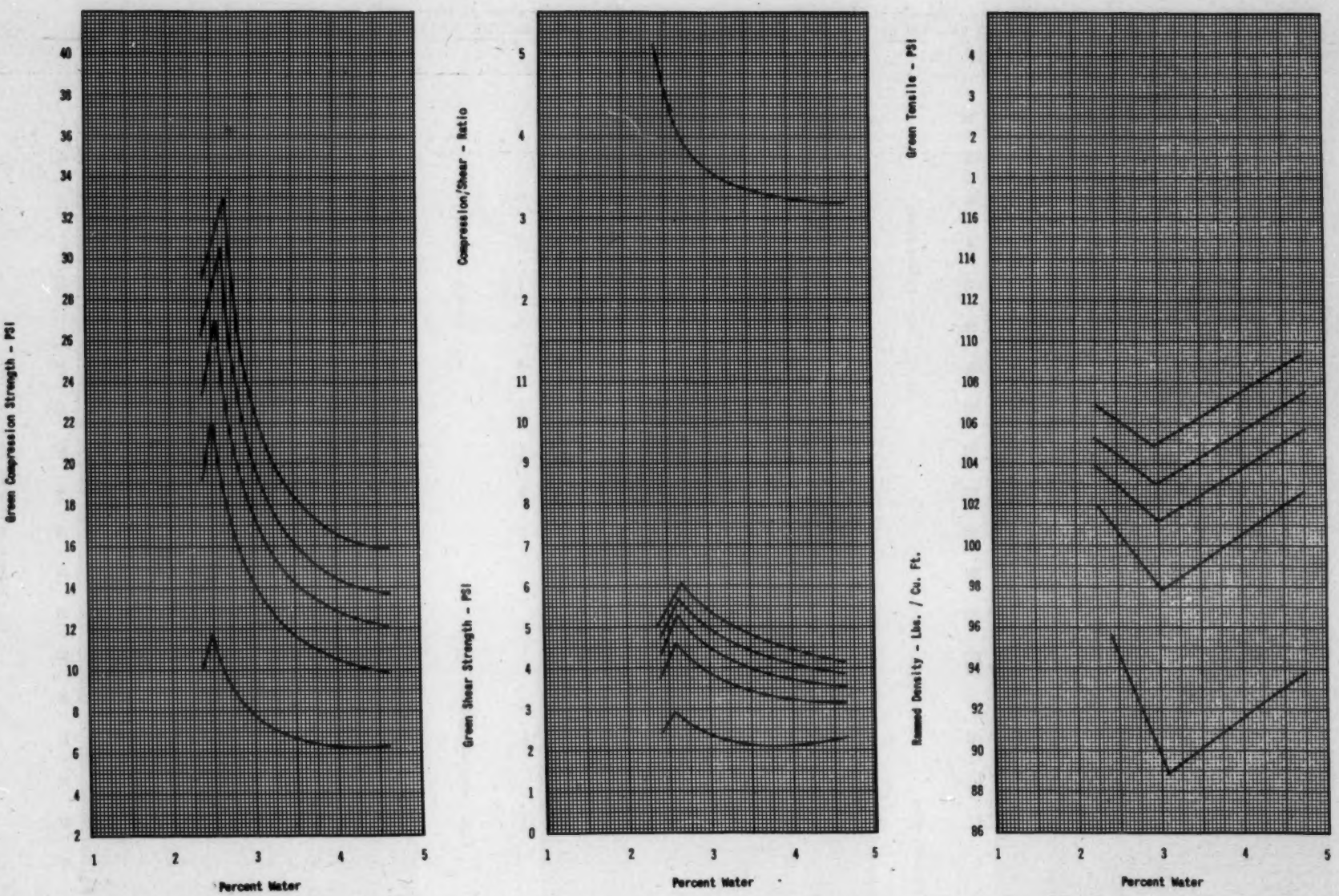



Fig. 7 — Water content and ramming energy effect on physical properties of 7.45 per cent western bentonite bonded Portage silica sand mulled 6 min. Method of addition: water followed by clay.

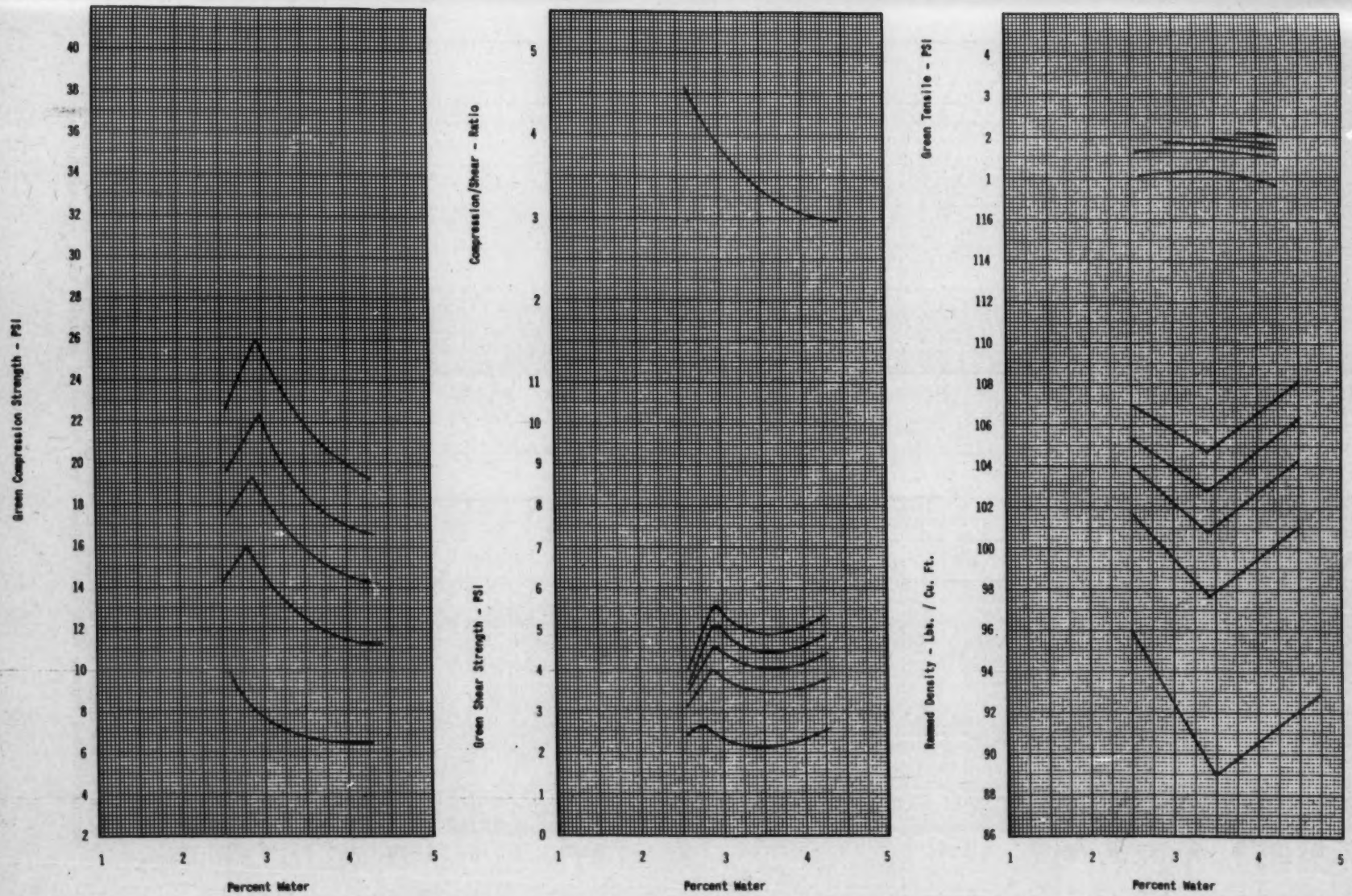



Fig. 8 — Water content and ramming energy effect on physical properties of 7.45 per cent western bentonite bonded Portage silica sand mulled 2 min. Method of addition: water followed by clay.

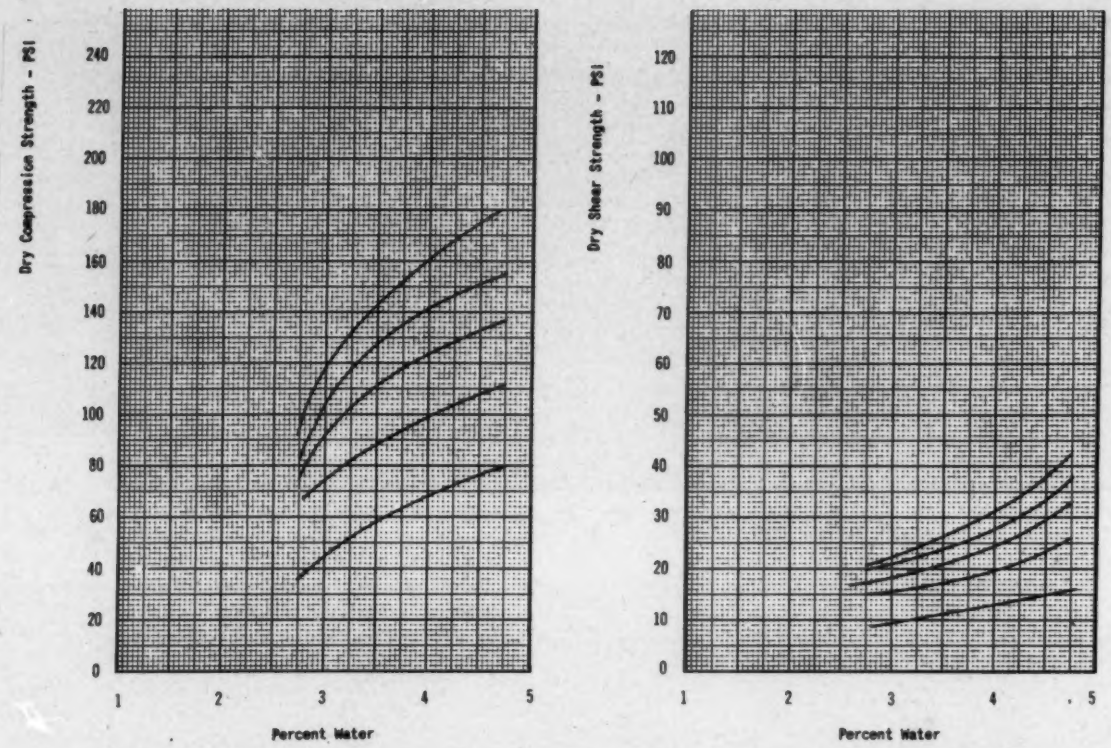



Fig. 9 — Water content and ramming energy effect on physical properties of 7.45 per cent southern bentonite bonded Portage silica sand mulled 4 min. Method of addition: water followed by clay.

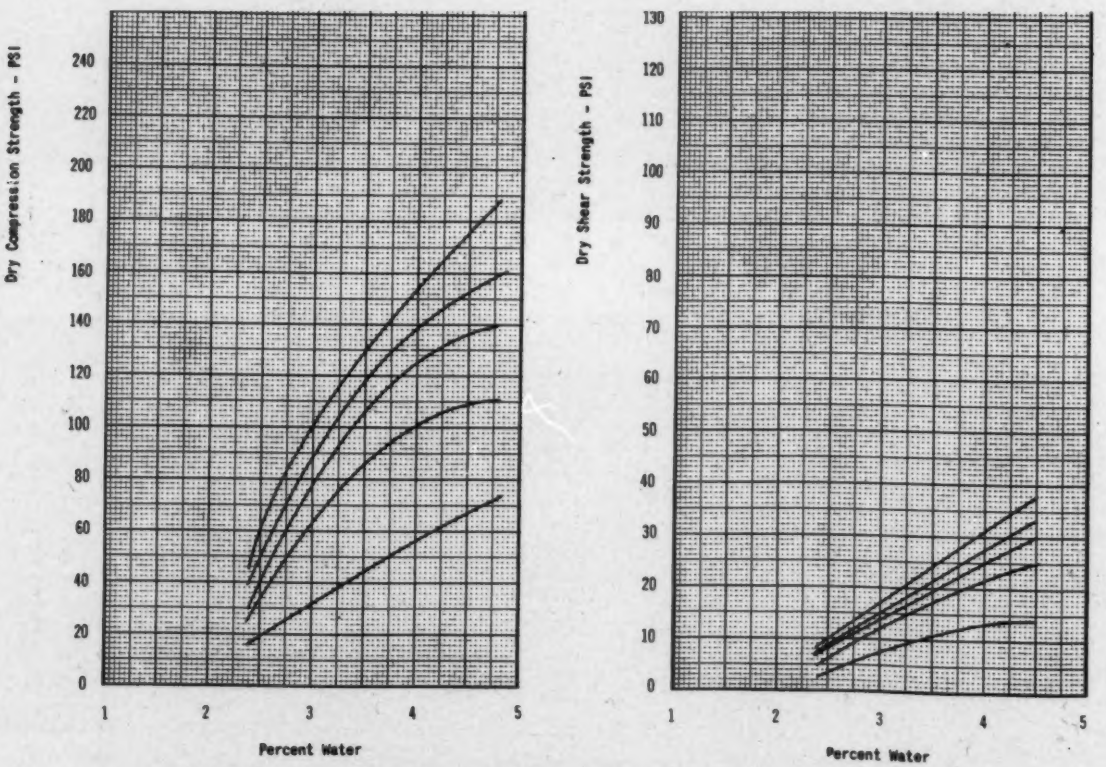



Fig. 10 — Water content and ramming energy effect on physical properties of 7.45 per cent southern bentonite bonded Portage silica sand mulled 2 min. Method of addition: water followed by clay.

that it is long or complete, with the assumption that extensive mulling is good mulling. This progress report will show that mulling is relative, that there is no absolute and that equipment as well as clay per cent, clay type and water content are variables which must be considered.

Previous reports have shown the influence of mulling time on the physical properties of sand-clay-water systems. In general, as the water content of the sand mixture is increased, the effects of varied mulling decrease. At the higher water contents, it is actually possible for overmulling to occur, and sands can actually experience a drop in green and dry properties. It was also noted that western bentonite sands were influenced to a considerably greater extent by variations in mulling than were fireclay and southern bentonite sands, in that order.

Since western bentonite bonded sands are affected by mulling variation to a greater degree than either fireclay or southern bentonite sands, the majority of data to be covered will deal with western bentonite sands.

### *Green Strength Development*

A thorough study of Figs. 1 through 6 shows that as western bentonite content is increased the mulling time required for development of maximum green

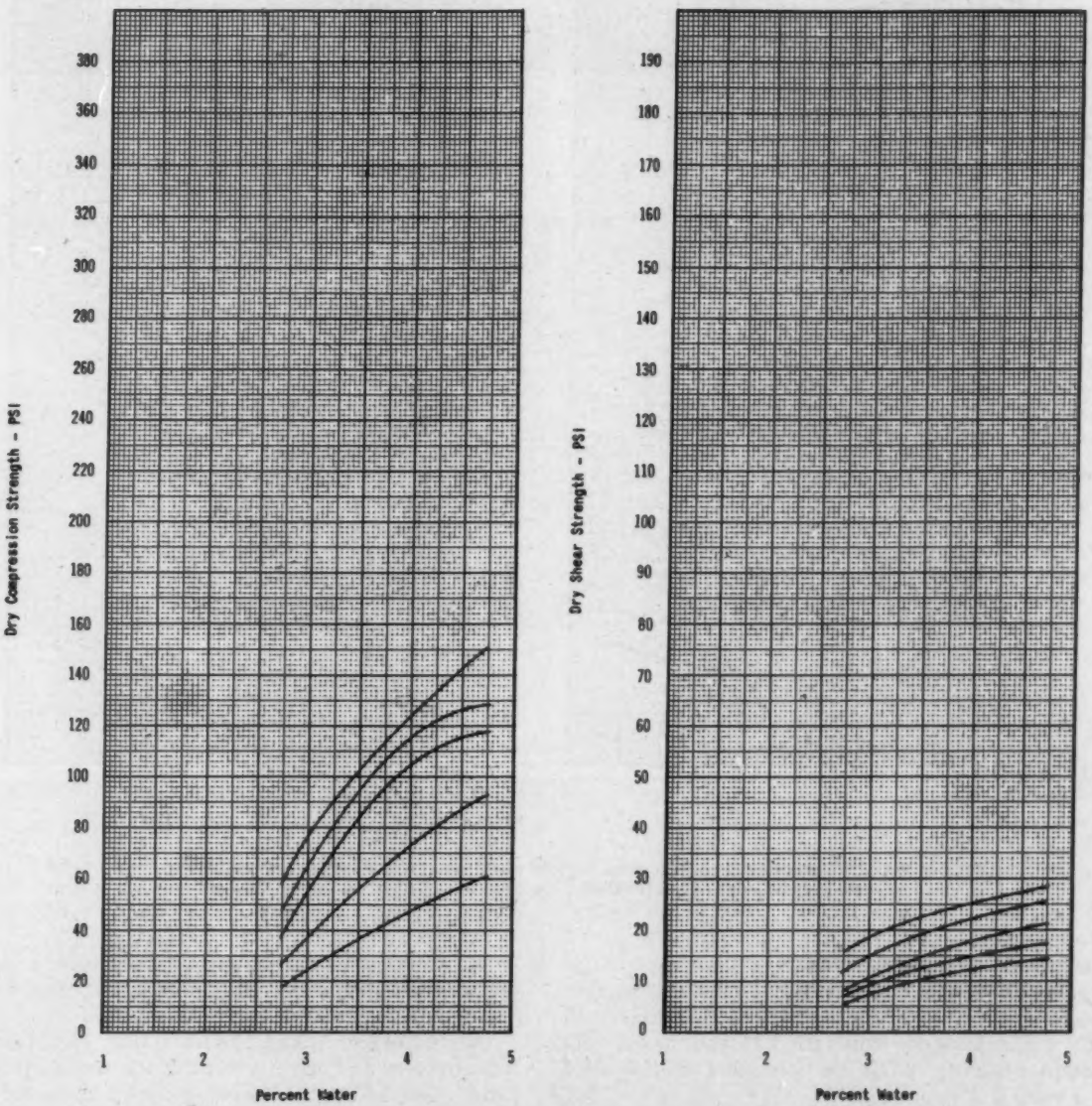



**Fig. 11 — Water content and ramming energy effect on physical properties of 15 per cent fireclay bonded Portage silica sand mulled 2 min. Method of addition: water followed by clay.**

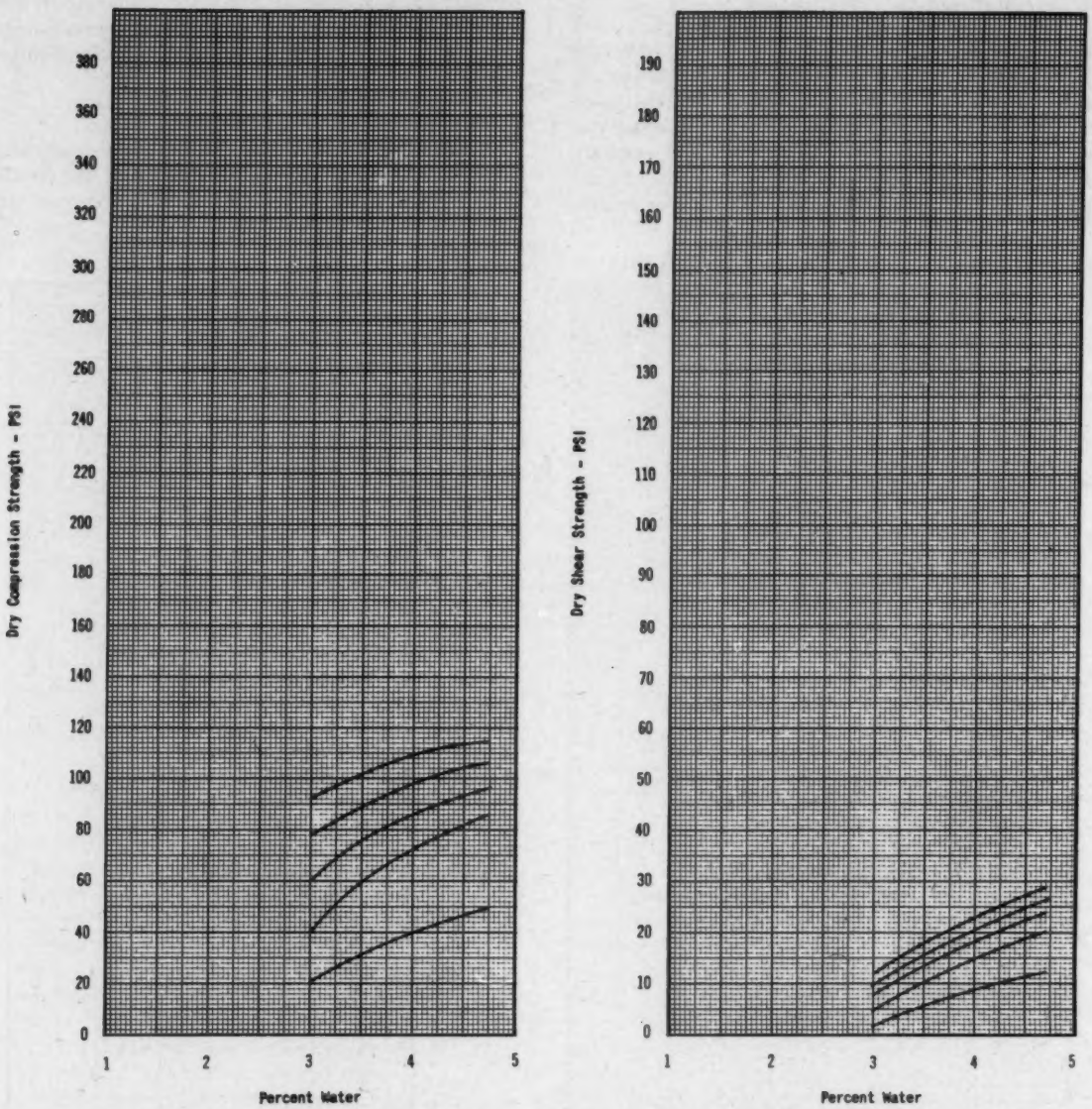



**Fig. 12 — Water content and ramming energy effect on physical properties of 15 per cent fireclay bonded Portage silica sand mulled 6 min. Method of addition: water followed by clay.**

strength at water contents near peak strength is increased. The system bonded with 4.75 per cent western has come close to reaching its full potential with only 4 min mulling, while the 7.45 and 10 per cent western bentonite sands are somewhat undermulled at 4 min.

Data dealing with the development of dry properties, though not presented here, substantiate the conclusion that mulling requirements increase with increasing clay content. It is also apparent that at high water levels the clay content becomes a minor factor, and that at all clay levels green compression actually dropped with extensive mulling.

A comparison of Figs. 3 and 4 with Figs. 7 and 8 illustrates the effect of reversing the method of clay and water additions. Figures 3 and 4 show that development of physical properties is slow for sands prepared by adding water to a mixture of sand and clay. In contrast, sands prepared by adding clay to pre-wetted sand develop much faster, as can be seen

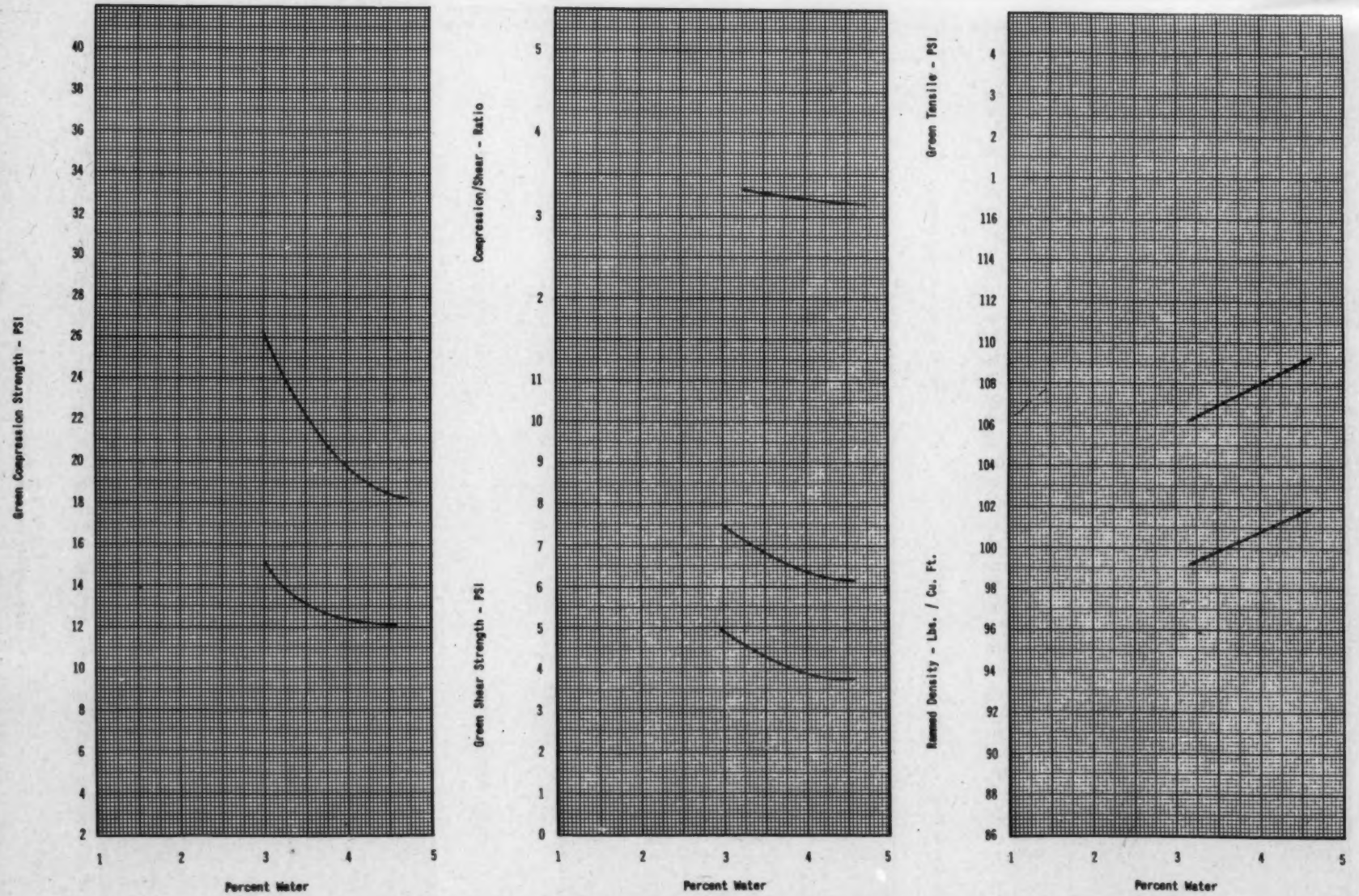



Fig. 13 — Water content and ramming energy effect on physical properties of 7.45 per cent western bentonite bonded Portage silica sand mulled 2 min in muller with 12 in. diameter. Method of addition: clay followed by water.

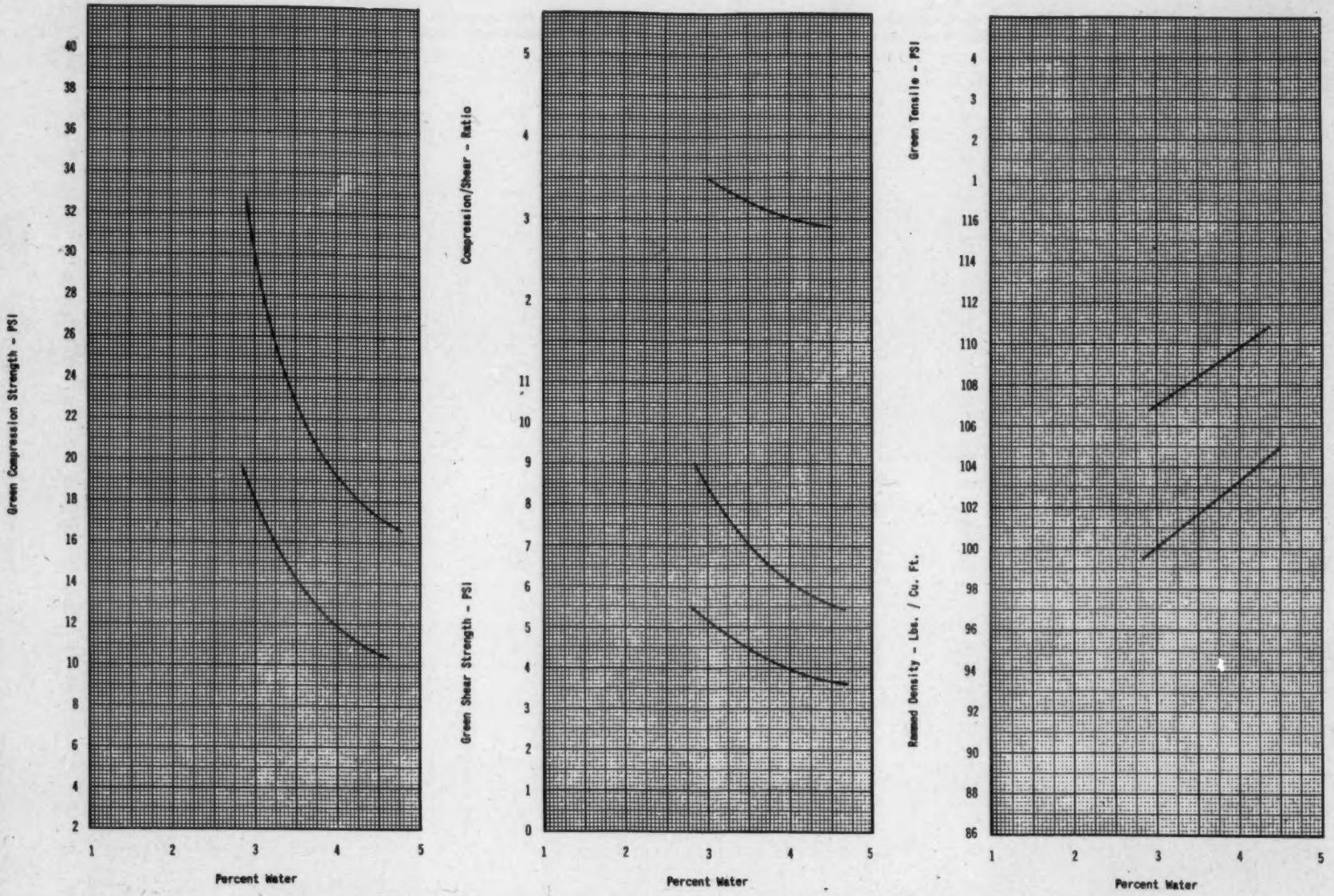



Fig. 14 — Water content and ramming energy effect on physical properties of 7.45 per cent western bentonite bonded Portage silica sand mulled 6 min in muller with 12 in. diameter. Method of addition: clay followed by water.

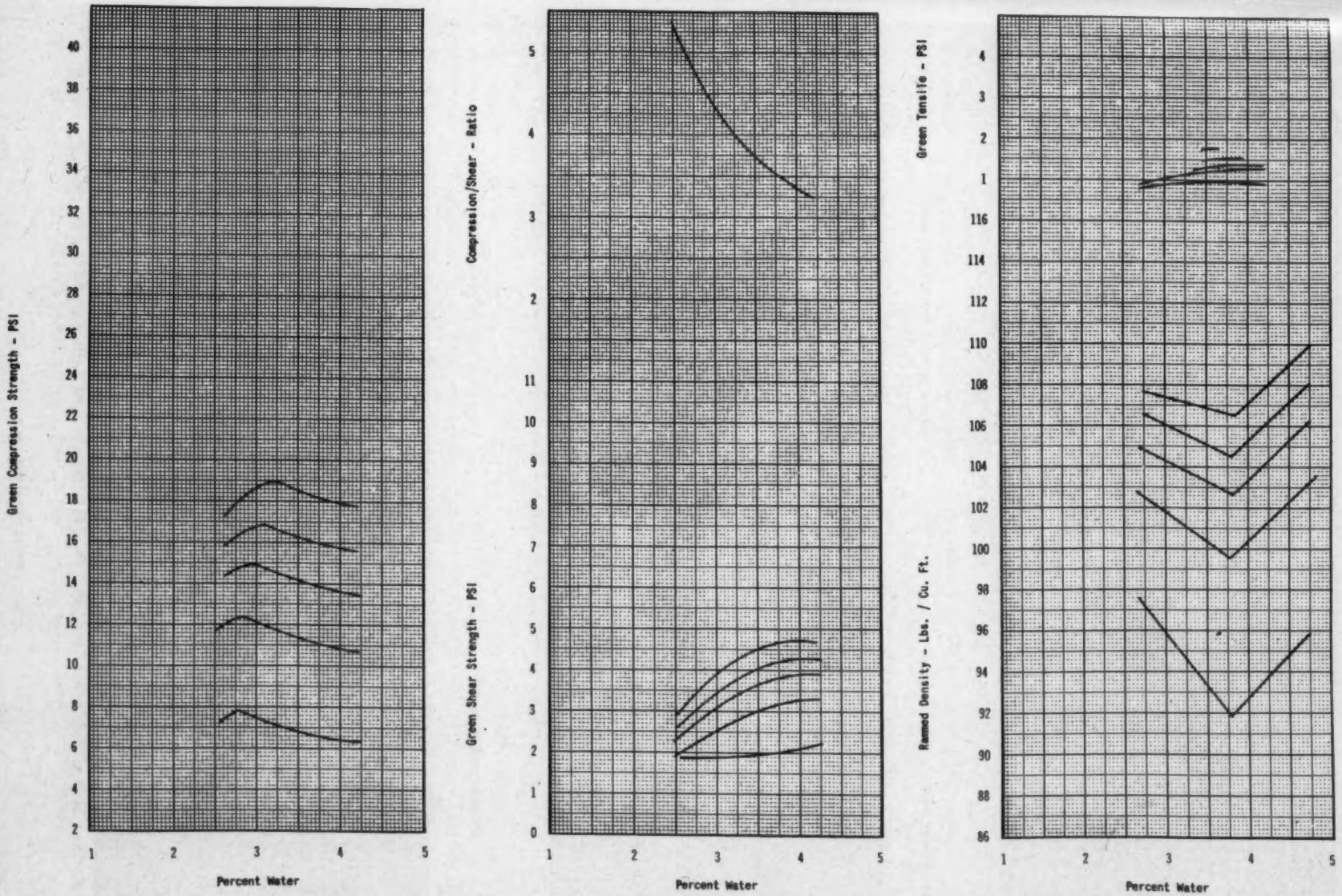



Fig. 15 — Water content and ramming energy effect on physical properties of 7.45 per cent western bentonite bonded Portage silica sand mulled 2 min in muller with 40 in. diameter. Method of addition: clay followed by water.

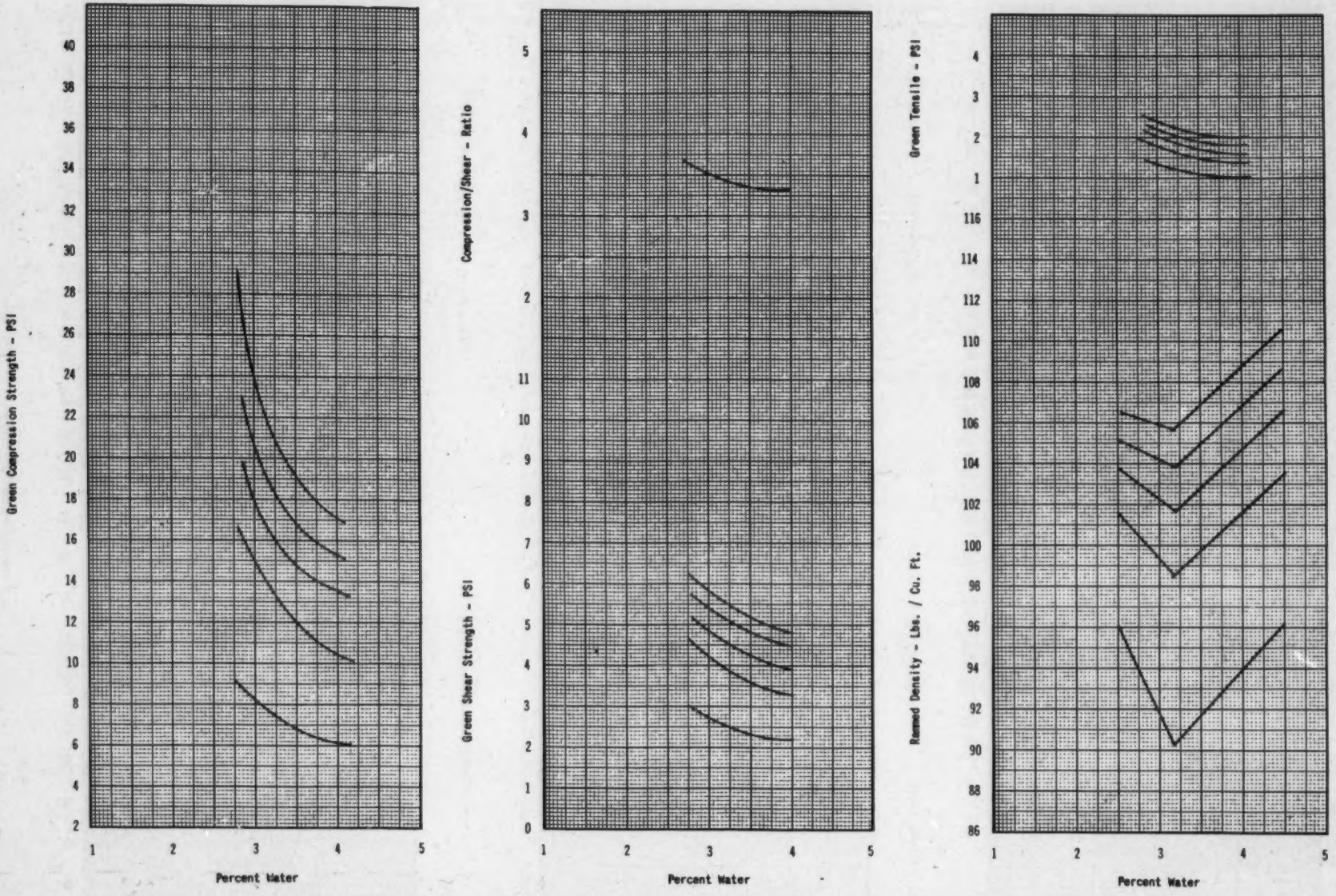



Fig. 16 — Water content and ramming energy effect on physical properties of 7.45 per cent western bentonite bonded Portage silica sand mulled 6 min in muller with 40 in. diameter. Method of addition: clay followed by water.

in Figs. 7 and 8. This method of additions allowed development of properties in only 2 min mulling, which were not too far removed from those obtained with 6 min mulling when water is added to sand-clay mixtures.

It can also be seen by comparing Figs. 4 and 8 that the mulling efficiency at 6 min mulling is substantially greater when water additions preceed clay ones. At high water contents, however, the effects of overmulling become more apparent through use of the water-first method. Although green properties dropped with improved mulling at high water levels, dry strengths continued to rise as mulling efficiency improved.

Southern bentonite and fireclay bonded sands reacted quite differently to changes in the method of addition. Southern bentonite and fireclay sands develop green properties quite readily regardless of the method of addition, the water-first method being slightly superior. These southern and fireclay sands experienced a drop in green as well as dry properties with increased mulling time. The use of the water-first method of additions also resulted in a substantial drop in dry strength. The cause of what appears to be overmulling is not understood at this time. The only comment which can be made is that it does occur.

### *Method of Addition Effect*

There can be no doubt that the method of addition has important bearing on the mulling characteristics and final physical properties of clay bonded sands. To further assess the importance of mulling as a variable, three mullers, as described in the experiment section of this report, were used to prepare sands for evaluation of mulling characteristics and potential.

To accentuate the differences in mullers, the 7.45 per cent western bentonite-sand-water system, with water additions following clay additions, was used. As expected, the data collected showed that both the muller characteristics and potential of the three mullers are different. In comparing Figs. 3 and 4 with Figs. 13 through 16, it becomes apparent that the 12 in. diameter laboratory muller not only develops physical properties more rapidly, but the potential at 6 min mulling is also greater than the 18 in. diameter muller.

The 40 in. diameter muller developed higher physicals much earlier in the mulling cycle, but its potential at 6 min mulling is not much different than the 18 in. muller. As previously mentioned, green properties actually dropped with increased mulling time or efficiency at high water levels. As a consequence, sands containing high water that were mulled in the 12 in. muller for 6 min had the lowest strength, and in keeping with the observations made earlier relative to dry properties, had the highest dry properties.

The 3-ram dry compression at 4.3 per cent water was 190 psi, substantially higher than the 152 psi obtained with both the 18 and 40 in. mullers. This also seems to confirm the popular opinion that the dry strength potential of western is higher than that of southern bentonite. It will be recalled that the bulk of data from earlier reports indicated that the dry strength potential of western and southern was equal. This again emphasizes the importance of mulling as a major variable. When we consider that the three mullers studied are essentially the same type, it is not illogical to expect even greater differences if the study were extended to other type mullers.

### *Data Application*

Application of the principles described by the data of this report can aid the foundryman in choosing the most economical combination of mulling equipment, clay type and content and working water range. While efficient mulling equipment would give greater flexibility, the data clearly show that through the application of principles, sands of adequate quality can be produced by low efficiency mixers and cutters.

Where mixers or cutters are employed, adequate physicals can be consistently obtained by employing high clay and high water levels. Consistent results cannot be obtained with this type equipment when low water levels are employed, nor can adequate physicals be obtained except through use of high clay content.

The more efficient mullers, on the other hand, possess a high degree of flexibility. Through a combination of low clay, low water and long mulling, sands of medium physicals can be obtained. High physicals can be attained through use of high clay, low water, and long mulling. Where facilities are overtaxed, large volumes of sand can be produced through short mulling cycles and a combination of high clay and water levels.

The data clearly show that it is not possible to produce sands of consistent quality at low water levels when low mulling efficiency is employed. Since high strength is obtained only at lower water, it is not possible to obtain consistent sands of high strength except through efficient mulling. There is sufficient difference in the mulling characteristics of western bentonite, southern bentonite and fireclay to indicate that the choice of clay to be used can depend on the prevailing mulling conditions.

## CONCLUSIONS

Dry sands are influenced by mulling to a greater extent than wet sands. Mulling efficiency decreases with increased clay at low water contents.

Western bentonite sands respond to mulling at a slower rate than fireclay and southern bentonite sands, in that order.

At high water levels, sands bonded with any of the three clay types experience a drop in green physical with extended mulling. Fireclay and southern bentonite sands also experience a drop in dry physicals.

Addition of water to sand followed by clay rather than clay addition followed by water is the more efficient method of additions.

There is a significant difference in the mulling performance of various mullers.

## ACKNOWLEDGMENT

The writers wish to express their appreciation for permission of Magnet Cove Barium Corp. to gather and evaluate the data for this series of progress reports and publish the findings in written form.

# AIRCRAFT – MISSILE STEEL CASTINGS

## *Challenge or compromise*

by S. A. McCarthy


### ABSTRACT

**During the past and present years of jet aircraft and missile progress foundrymen have become more and more adept in the knowledge to the needs, as well as the rewards, of working directly with the design engineers in creating steel castings to meet the requirements of the producer and consumer. The intention of this mutual objective is that steel casting design must make efficient use of the practical metals and alloy compositions to result in good castability in the foundry. The system of good steel casting designs must be a satisfactory engineering approach to obtain these two desired end products — 1) good castability characteristics and 2) equal stress distribution while in application.**


### INTRODUCTION

Foundrymen and design engineers of steel castings are aware that each has basically similar problems in castings, either of commercial or aircraft quality application. However, there must be a separation in the preliminary planning of the design which will produce the best possible casting at the lowest possible cost. This division must be understood—that aircraft and missile steel castings are not processed through the foundry in the same "tote" boxes, but approached on an individual basis other than for commercial or mechanical requirements. This is the result of the high standards demanded by the military services that are mandatory in the production of precision quality castings for high performance aircraft application.

### DESIGN COORDINATION

When developing successful designs to be processed as precision steel castings, assurance of the serviceability must be focused on engineering material that is dependent on quality and adequate laboratory and field testing information, if satisfactory production is ultimately to be achieved. These principles imposed are generally located in foundry practice to involve adjustment, rather than radical changes in intended design. Every steel casting design undertaken for aircraft or missile application forms a challenge or compromise. The challenge usually attempts to obtain tolerance ranges in the as-cast condition to eliminate additional costly major or minor machining requirements. The compromise may include several concessions:

1) Some dimensioning may indicate excess metal to permit machining for closer inside limits than possible economically in the as-cast condition.
2) Complex or intricate cored areas or holes be solid metal where a high degree of close tolerance accuracy of center locations is required that would necessitate end mill or reaming operations be overly difficult.
3) Design modification to be of assistance to the foundry and facilitate the desirable aspects for elimination of added machining.

A great percentage of aircraft steel casting designs submitted are often referred to as "tool room" drawings, when dimensional tolerances indicated are less than $\pm$ 0.03 in. and draft angles two degrees maximum, with wall or sectional thickness less than 0.25 in. over all. Precision quality steel castings are the result of coordinated evaluation previous to final acceptance that includes stress analysis, material, weight, cost and procurement time.

When these factors are favorable, the design is released for production, and firm costs are sought by sealed bid quotation from three or more approved casting sources. Upon return of the quotation with the marked up "redline" print to Engineering, a complete re-evaluation is made of the requested concessions by the casting facility. This may involve changes in dimensional tolerances, radius, fillets, alloy preferences, etc. If affirmative or negative, the incorporated requests are made on the drawing to be released as acceptable for the design of the casting by the successful bidding facility.

### PRODUCTION CONCESSIONS

Occasionally it becomes necessary to deviate from a foundry approved casting design for unforeseen circumstances. This happens where the alloy characteristics of the configuration result in a high rejection

S. A. McCARTHY is Senior Production Design Engr., McDonnell Aircraft Corp., St. Louis, Mo.

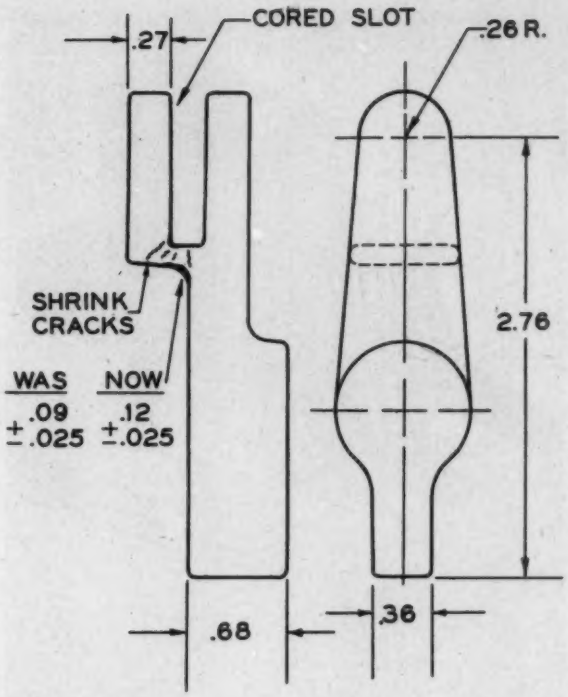



Fig. 1 — A shrink and crack condition of a small shut-off valve arm casting.

rate at final inspection. A minor change may be required to result in 100 per cent acceptable castings of uniform high quality.

The example (Fig. 1) will illustrate a shrink and crack condition of a small shut-off valve arm casting. This was corrected by increasing the radii from 0.09 in. to 0.12 in. ± 0.025 in. The shrink-crack would have been avoided if the cored slot had not been requested in the rough casting to reduce the machining time to a simple skin surface operation. The increased radius was the logical method. Rejects were reduced from 100 per cent to 3 per cent or less.

The cut-away view (Fig. 2) of an engine air duct casting shows where a fine hair crack developed along the 0.09 in. flange which resulted in 100 per cent rejects. A foundry request to increase web section thickness was not permissible as an aluminum skin is rivetted to the highly stressed stepped flange of the duct. However, a dimensional problem occurred on the contour (Fig. 3) after gate removal, where blending of the in-gate stub to the as-cast surface was not in tolerance to over grinding. This type of deviation from an acceptable design will require a compromise with Quality Control, Stress Engineering and the design engineer if acceptable or rejected.

| | |
|---|---|
| Raw Casting Cost, lb | $41.50 |
| Raw Casting Weight, lb | 3 |
| Machined Casting Weight, lb | 2.44 |
| Material | A.M.S. 5350 |

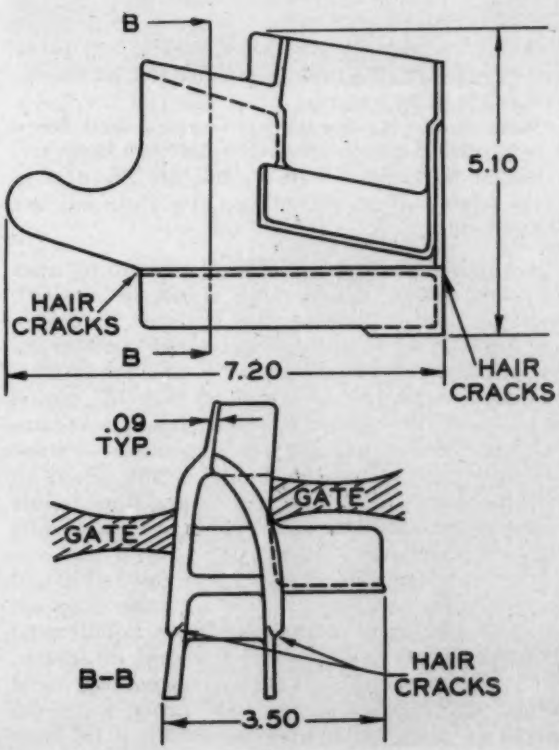
## SEVEN DESIGN RULES

The design engineers, through their contacts and intimate workings with the foundryman on a particular steel casting, are alerted that to produce the utmost in the casting seven "rules of thumb" are to be considered. In order to produce the best possible casting for smooth as-cast surfaces with minimum weight and costs, whenever the application of the part will permit:

1. Induce progressive solidification by tapering or gradually decreasing adjoining sections as the distance from the largest center or centers of mass become greater. This is to induce x-ray soundness in the junctions and surfaces.
2. Avoid isolated heavy sections in thin walls which cannot be easily gated or fed by separate risers.

Fig. 2 — A cut-away view of an engine duct casting showing fine hair crack developed along the 0.09 in. flange.

3. Do not add excessive clean-up stock on functional surfaces. This will vary from 0.015 in. to 0.030 in. on castings one to 3 in. in their maximum dimensions, to 0.090 in. to 0.125 in. on larger castings.
4. Fillet design is critical. Excessive fillets cause sections which cannot be fed, and a centerline type



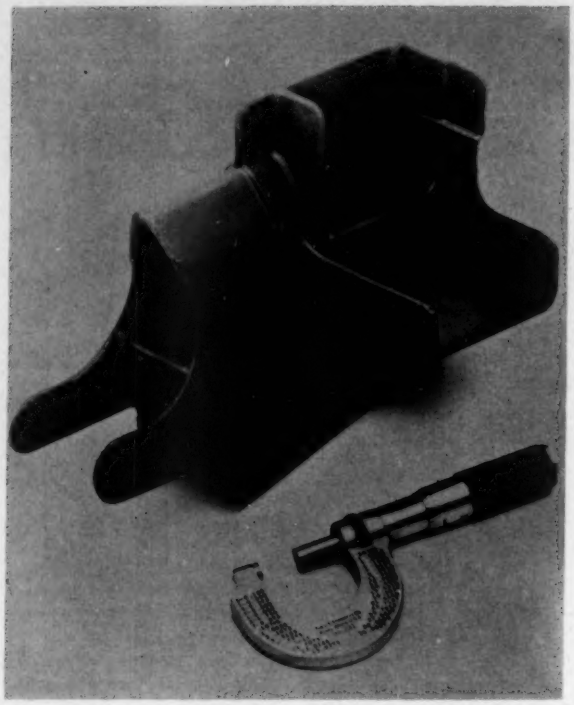


Fig. 3 — Dimensional problem on the contour after gate removal where blending of the ingate stub to the as-cast surface was not in tolerance to over grinding.

porosity will result. Sharp corners ($\frac{1}{64}$-in. R or less) will cause cracks and surface shrinkage.

5. Plain flat surfaces or unbroken gradually curved surfaces, if proportionately larger than adjacent bosses, etc., will have inadequate surface and distortion due to inherent problems while in process. These should be broken by gussets in adjoining corners, ribs and minor corrugations in the surface.
6. Wall thicknesses should be held to a minimum of 0.12 in. on small castings. However, 0.60 in. to 0.090 in. can be held if possible to be fed by adjacent heavy sections.
7. Small holes $\frac{3}{16}$-in. diameter or less, and blind holes with depth in excess of twice their diameter, should be avoided. Cast threads are practical in the ferrous alloys providing Class 2 threads are acceptable.

It is realized that the ideal in a steel casting from the standpoint of design often cannot be attained, owing to factors limiting the engineer to certain conditions of form, contour and metal distribution. But a recognition of effects, due to and controllable by design, represents advantages to both the casting facility and the engineer. Design should not carry the desire for lightness to such a point as to insist on sections throughout the casting, or any part of it, so thin as to either cause serious loss from failure to run or greatly increased cost due to impractically high temperatures.

This latter condition increases both cost and hazard of production, and also produces a casting expensive to finish and often to machine. When liquid metal solidifies, it contracts about 3 per cent in volume. Since metal solidifies progressively from the mold surface to the center of the mold cavity, a contraction cavity will result unless the section is fed from some reservoir containing liquid metal. If the section is unable to draw liquid metal from other sources because these inlets have completely solidified, then a defect will appear in the unfed section.

Castings with defects of this type, under high frequency vibration stress aided by stress centralization, may develop cracks extending from the cavity to the casting face, causing ultimate rupture. If the casting is operating under pressure leaks may develop. The problem of removing the possibilities of defects due to contraction involves the elimination of hot spots in the casting. These isolated hot spots are sources of great trouble in the foundry, and anything that the casting designer can do to eliminate this condition makes for the production of higher quality castings.

## RISER PLACEMENT

One of the most difficult problems in designing to prevent contraction cavities is to know when and where feed heads may be placed, since their position is quite largely a matter of personal judgement based upon experience. The casting may be molded contrary to any calculations that may be made if the foundryman is not consulted; the casting may be better designed with the supposition in mind that none of the metal junctions or hot spots can be fed from outside risers. Assuming such procedure, experimentation must be carried on with adjoining sections to determine the extent of the contraction cavity with various changes in design.

This may be further clarified in Fig. 4, where the first production casting attempt by dry sand was not successful because of the shrink and misrun condition in the 0.25 in. and 0.16 in. wall and web section. After repeated relocation of the risers and blind risers in addition to some modification to the design, a complete turnabout was made and the dry sand mold discontinued. A coordinated redesign of the casting was made for the frozen mercury investment process, where the shrink and misrun problems were eliminated with closer dimensional and tolerance control. The raw casting weight savings of the two methods totaled 20 lb (dry sand casting weight 210 lb — frozen mercury process casting weight 190 lb).

Figure 5 shows the drag half of the casting, with the stiffeners on the 0.16 in. web sections from the

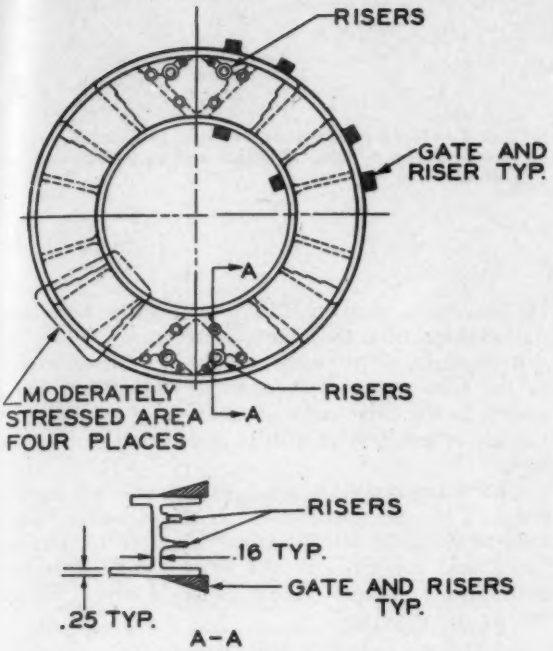



Fig. 4 — The first production casting in dry sand was not successful because of shrink and misrun condition in the 0.25 in. and 0.16 in. wall and web section.

outer diameter ring to the inner diameter ring, acting as additional gating to feed the 0.16 in. web areas that remained as an integral part of the casting. Combination gates and risers are clearly visible on the cope side. Figure 6, cope view, shows the riser locations on the external diameter of ring, with hidden risers located on bosses still intact. Figure 7 is a more direct view of the cope half of the casting with all risers and gating locations clearly visible. This design is an example of approximately the maximum in as-cast weight to date for a missile application steel casting.

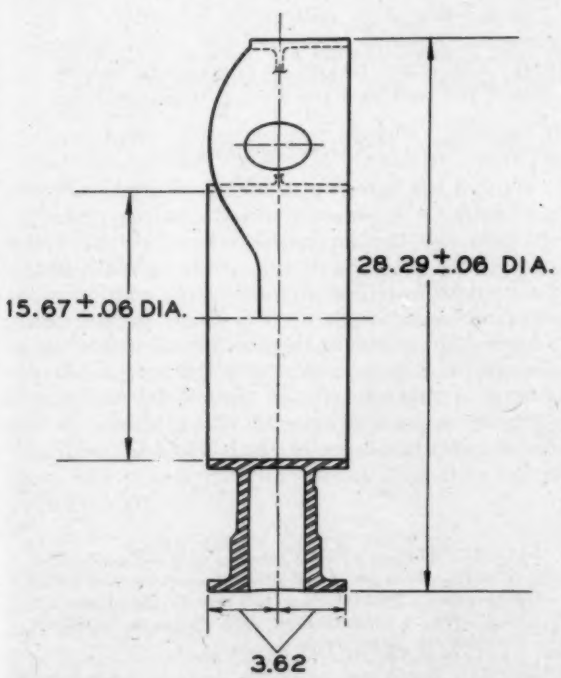



Fig. 4a — Cross-section of "Moderately stressed area, four places" in Fig. 4.

## WEIGHT CONTROL IMPORTANCE

The design engineer of airframes, missiles and aerospace vehicles are motivated for a particular design concept, where the lighter and smaller the

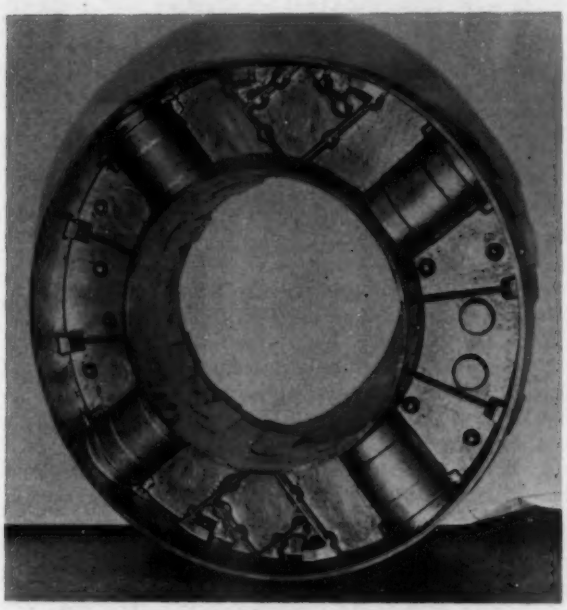


Fig. 5 — Drag half of casting with stiffeners on the 0.16 in. web section from the outer ring to the inner diameter ring which act as additional gating to feed the 0.16 in. web areas that are integral parts of the casting.



Fig. 6 — Cope view of casting showing riser locations on the external diameter of the ring with hidden risers located on bosses still intact.

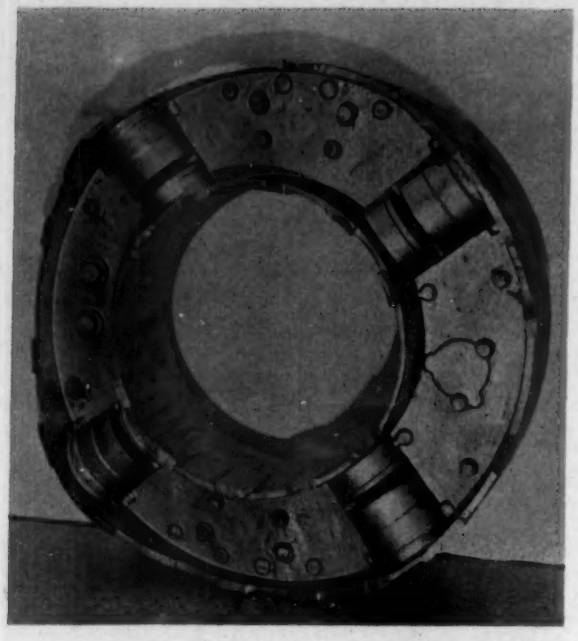


Fig. 7 — More direct view of cope half of casting, shown in Fig. 6, with all risers and gating locations clearly visible.

vehicle the less power will be necessary to acquire the same speed, range and altitude. It is the usual approach to create a hypothetical figure for cost controls of parasitic weight accumulation in a structure. For example, in an airframe one lb of increased weight results in a gross increase of approximately 4 lb, with at least 10 lb gross increase in the high performance vehicles anticipated, at a cost of $10 to $100/lb of structure.

In missiles, this is increased to approximately $400 to $500/lb of structure, and in aerospace vehicles, $4,000 to $5,000/lb of structures. It has been estimated that every pound of added weight to a missile at the end of burn-out, requires about 40 lb of extra fuel for that particular stage for the same initial fuel load every extra pound reduces the range approximately one mile. It may be our thinking that fuel is inexpensive, and adding additional fuel would be less costly than reduction of ounces here and ounces there, from the casting or structure.

Reduction of parasitic weight is recommended as the most economical method when every extra pound in the final stage of multi-stage missiles may require as much as 1000 lb of added fuel in the first stage.

This information is intended for a better understanding of the importance, to the designer, when close dimensional tolerances are noted on the drawing for a steel casting, and the efforts undertaken for maintaining weight accumulation control in the casting and structure.

## REJECTION ANALYSIS

An analysis has been made of the steel castings received and placed on salvage because of various discrepancies over a two year period. These data are presented in Fig. 8. The rejection percentages shown are not to be accepted as total scrap or unusable, but to relay to the foundryman the additional handling and paper work necessary after the castings leave the foundry's shipping platform. The greater percentage is delayed by inspection for dimensional tolerances, not conforming to the drawing callout. This is followed by radiograph indications or interpretation, physical properties and minor surface defects. The actual returns to the casting facility vary from 3 per cent to 5 per cent as unacceptable.

## CONCLUSION

What is the correct approach to weight control in the castings? Apparently the design engineers are studying the developmental efforts of the higher strength to weight beryllium alloys and the 300,000 psi strength materials. In view of the considerations, assurance must be that every ounce in the casting designed must contain the anticipated values, by insistence on close tolerances. The aircraft industry in general is depending upon the casting facilities to assist in providing some of the solutions for the future, and the advent of Mach 3 and up vehicles.

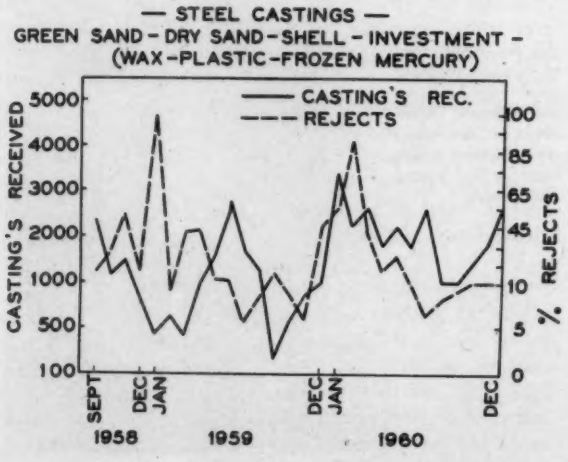



Fig. 8 — Analysis of steel castings received and placed on salvage because of various discrepancies over a two year period. The rejection percentages shown are not total scrap or unusable, but those requiring additional handling and paper work.

# HOT CORE BOX DESIGN AND ENGINEERING

by E. E. Harkess


## ABSTRACT

**A description of equipment and production processes in the foundries of the author's company is presented, dealing with the use of faster setting binders composed of various combinations of urea and formaldehydefurfural and single stage phenolic resins. Of prime importance in this process is good sand control. The proper measurement of ingredients and timing of their addition during controlled mulling cycle, plus care in sand handling to the machine help make a uniform product on the machine. The hot core box process seems to offer advantages as increased productivity and and material cost reduction.**


## INTRODUCTION

A large number of foundries currently use the shell process to produce molds and cores in the manufacture of some castings. Probably the most widely used thermosetting sand binders are the phenolic resins. Some plants are experimenting with the use of faster setting binders composed of various combinations of urea and formaldehydefurfural and single stage phenolic resins.

These faster setting binders seem to promise increased productivity. In addition, material cost reduction should encourage increased use of the economical "hot box" method of producing accurate cores. The hot box process offers many advantages, but the process finally selected should be based on a careful appraisal of all production factors involved.

Good sand control is of prime importance. Proper measurement of ingredients and timing of their addition during a carefully controlled mulling cycle, plus care in sand handling to the machine, will help greatly in making a uniform product on the machine. This discussion is limited to a description of equipment and production processes in foundries of the author's company.

## CAST IRON CORE BOXES

Our core boxes are made of cast iron, an economical, serviceable material with good heat retention, and relatively good stability when exposed to heat. The boxes are made of S.A.E.-120 cast iron stabilized at 1550 F for a minimum of 30 min, cooled to 1000 F in the furnace in one hr and finally air cooled to room temperature in still air that is 70 F minimum.

The walls surrounding the core cavity are made uniformly to 7/8-in. thickness, where practical, to help achieve uniform heating. Ribs are designed on the outside of the box to hold its shape when exposed to heat. Copper-base alloy inserts are sometimes used in phenolic resin boxes in critical areas where a faster heat transfer is required. Copper-base alloy inserts are not practical in furfural boxes, because of the corrosive action of the ingredients. Hardened steel is used for locating pins, bushings and wear pads.

## HOT CORE BOX DESIGNING

In designing hot boxes, factors related to expansion and contraction must be taken into account to obtain accurate cores. The most important of these factors are:

1. Expansion of cast iron core box when heated to operating temperature (based on a 450 F operating temperature, a 0.003/in. expansion factor is incorporated).
2. Expansion or contraction (determined by materials used) of cores after removal from box.
3. Dimensional change of cores requiring core wash dip and oven redrying.

Another design factor considered is the proper alignment of upper and lower half core boxes. This is usually accomplished by providing three elongated bushings; one on, or near and parallel to the crosswise centerline, and two on and parallel to the lengthwise centerline. On large boxes four bushings are desired; two on and parallel to each centerline, spaced as far apart as is practical. These bushing arrangements minimize misalignment and eliminate binding on locating pins when boxes are heated.

Heat, which is obtained by gas or electricity, is constantly being lost during each operating cycle, and must be replaced as uniformly as possible at a rate that will constantly maintain the desired operating temperature. Box temperature control is especially important when using the faster setting binders because over or under cure is detrimental to core quality.

Electrically heated core boxes have special heater plates attached to the back surface of the box (Fig. 1).

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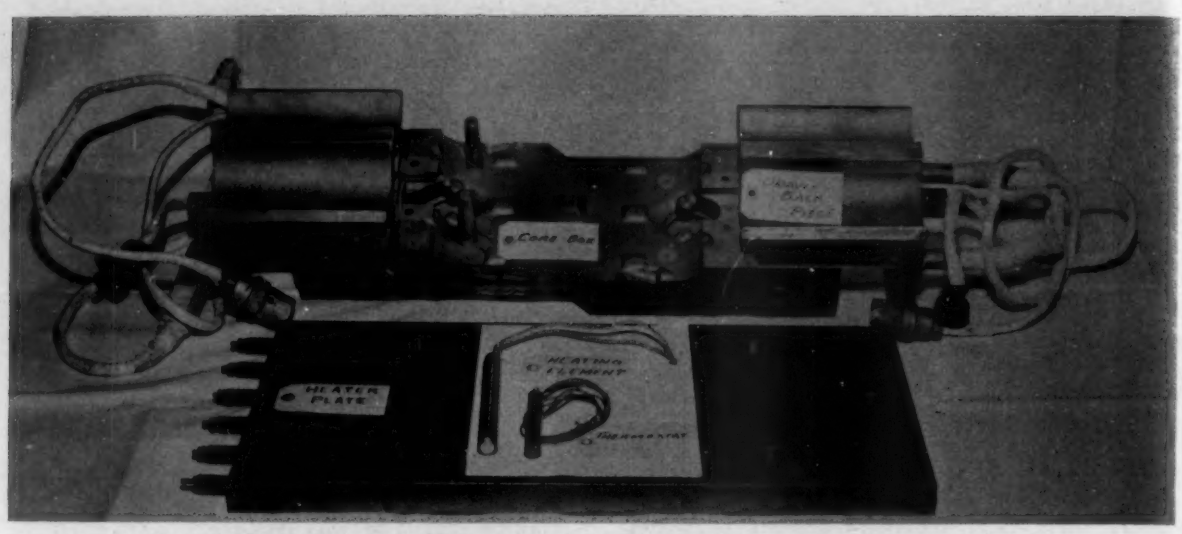



Fig. 1 — Electrically heated core boxes have heater plates attached to the back surface of the core box. Heating elements are cast into these heater plates with leads protruding from the ends.

Heating elements are cast into these heater plates with leads protruding from the ends. Complicated boxes often require cartridge-type heaters adjacent to thin sections or areas excessively cooled during blowing. All electrical elements are thermostatically controlled.

Loose pieces in the core box are sometimes necessary, but they should be avoided because of poor heat transfer. Manual or mechanically operated draw-backs should be substituted whenever possible. Aluminum-bronze "T" slot slides have proved successful as guides for the draw-back pieces.

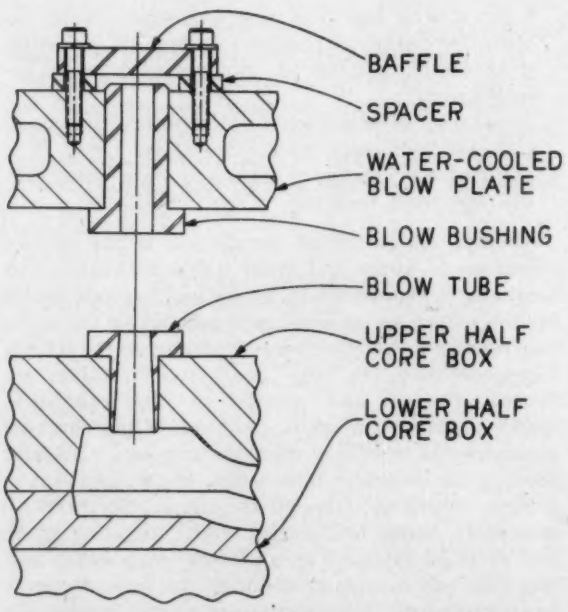



Fig. 2 — Blow plates are water cooled to minimize heat transfer from the box to the sand in the blow holes and in the sand reservoir. Water flows through the water jacket continually while the machine is in operation.

## SAND MIXTURES

Core sand mixes containing fast setting binders have a degree of green strength, and can be blown and vented similar to oil sand mixes. Vent screens may be used with tempered sand, but are undesirable for dry mixes which rapidly plug vent screens with free resin abraded from the sand grains. Special provisions are necessary for the dry, free flowing phenolic mixes. Milled vent slots (0.005 in. deep and 1.00 in. wide) are used on partings where possible. Large boxes are vented by providing a gap of 0.012 to 0.017 in. between all partings by means of C.R.S. inserts.

Blow plates are water cooled to minimize heat transfer from the box to the sand in the blow holes and in the sand reservoir (Fig. 2). Water flows through the water jacket continually while the machine is in operation. Blow holes can be the same size for tempered sands as for oil sand mixes. Dry mixes are provided with a baffle over the top of the blow hole in the reservoir, so sand will only drain out of the blow hole until it reaches its angle of repose at the edges of the hole.

Blow tubes in the upper half boxes are tapered to facilitate the ejection of cores. The large end of the blow tubes are located at the core shape. Cores, except those with a shallow draw, are ejected from the box by an ejection mechanism forcing cores from the box by means of pins sliding through blow holes. To reduce core damage, the end of the blow tube is chamfered to insure the blow plug breaking below the surface. This technique is always used where blow holes are on core prints.

Steel discs surround blow holes on the top of upper half core boxes to reduce the area of box contacting the blow plate for minumum heat transfer. When required, additional bearing pads are installed to prevent distortion of core boxes by clamping pressures.



Fig. 3 — A five station rotary machine is used in production of the barrel slab shell core for a V-8 cylinder block at one plant. This machine has a hot upper and lower box at each station.

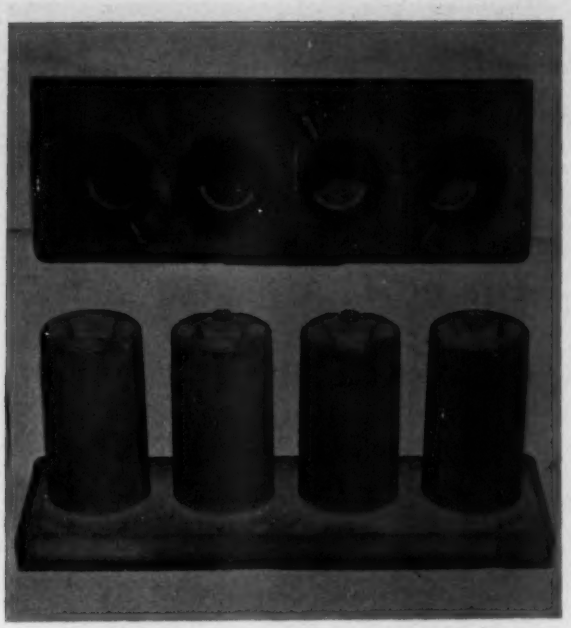


Fig. 4 — After removing the core, the operator blows all sand particles from the boxes and sprays the box with core release.

## PROCESS OPERATION

The barrel slab shell core for the V-8 cylinder block is made at the author's company's foundry on a five station rotary machine which has a hot upper and lower box at each station (Fig. 3). Starting at the blow station, the upper and lower half core boxes are raised against the water cooled blow plate, which is fastened to the sand reservoir. The core is blown, the boxes are lowered and cycled into the oven portion of the machine. The boxes are heated by means of gas jets evenly spaced around the inside of the insulated oven.

The boxes are heated through three of the five stations of the machine, and are cycled into the ejection station. At this station, the lower half box is retained in position by engaging cleats while the upper box is raised against pins attached to the ejection plate. With the upward movement of the box, the pins pressing against the blow tips eject the core onto a platform swung into place by the operator.

After removing the core (Fig. 4), the operator blows all sand particles from the boxes and sprays box with core release as required. The upper half box is then lowered into place, and the unit cycles into the blow station. Both the upper and lower half of the core box are hot. The core is always retained in the upper half for ejection purposes.

Barrel slab shell cores for V-8 blocks are made in another foundry of the author's company on a shuttle machine in this way—two units, or boxes, are mounted in a wheeled carriage that shuttles back and fourth between the blow and ejection stations (Fig. 5). This machine permits four cores to be made at each blow. After each blow the unit moves into the ejection station where the upper stripper mechanism engages the core at the blow tube locations. At this point the entire unit remains stationary while gas heat is ap plied to cure the core. The upper box is then lifted a short distance to strip the box clear of the core.

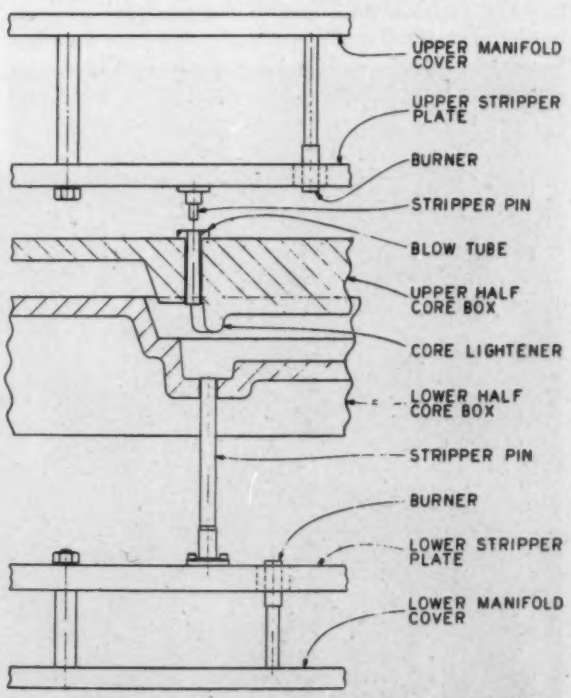



Fig. 5 — Barrel slab shell cores for V-8 blocks are made on a shuttle machine in another plant. Four cores can be made at each blow.

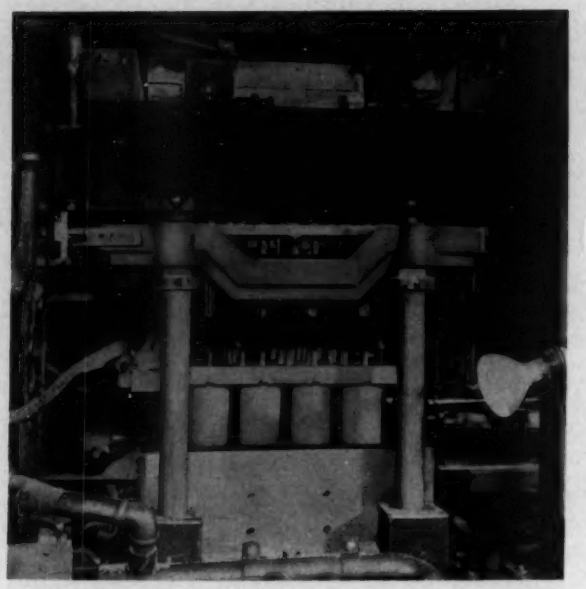


Fig. 6 — Strip mechanism and upper half of core box are raised a sufficient height for core ejection from the lower half box.

*Core Ejection*

The strip mechanism and upper half box are then raised a sufficient height for core ejection from the lower half (Fig. 6). Next, the ejection mechanism in the lower half box raises the cores to a height sufficient for the pick-off fingers to move into position.

The pick-off fingers raise the cores clear of the box and ejection pins, and then retract for pick off by the operator (Fig. 7). Air and spray lines attached to these fingers clean and spray the box prior to lifting the cores clear of the box. After the cores are removed, the upper half box is lowered into position. The unit is now ready to cycle into blow position. The upper box is heated by gas jets projecting through the stripper mechanism. The lower box is heated by gas jets projecting through the ejection plate.

Flywheel end cores for V-8 blocks are made in this foundry on the same type machine used for production of barrel slab cores. The cutaway view of various parts that make up the unit for this type shell box are shown in Figs. 8, 9 and 10.

The combustion chamber slab core for one automobile cylinder head is made in the author's company's foundry on a six-station rotary machine. The six station machine has a hot lower box at each station, and a cold upper at the blow station only.

*Core Box Blowing*

At the blow station, the hot lower box is raised against the water cooled upper half which is attached to a flat blow plate bolted to the sand reservoir (Fig. 11). The core is blown, the heated box is lowered and cycled into the oven portion of the machine where it is heated by gas jets through three stations.

In the fourth, or ejection, station the box is retained by engaging cleats (Fig. 12). The ejection piston located beneath the box raises the ejection plate and pins forcing the core clear of the box. The operator manually removes the cores from the two gang box. The ejection plate lowers, and the box cycles into the fifth, or pre-heat, station before going into the blow station (Fig. 13). The ejection pins have 0.005 in. clearance on the radius and are held 0.03 to 0.06 in. above the box surface.

In this foundry, crankcase cores for an automotive cylinder block are made in electrically heated core boxes on a one station roll-over type core blower (Fig. 14). The door piece swings into closed

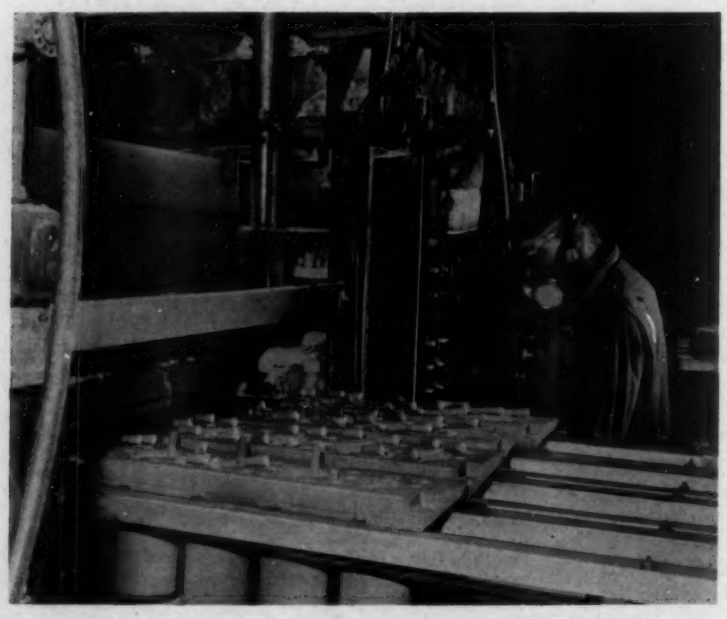
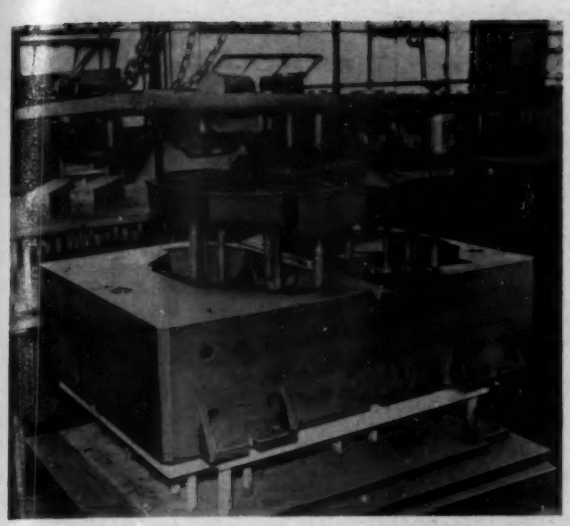
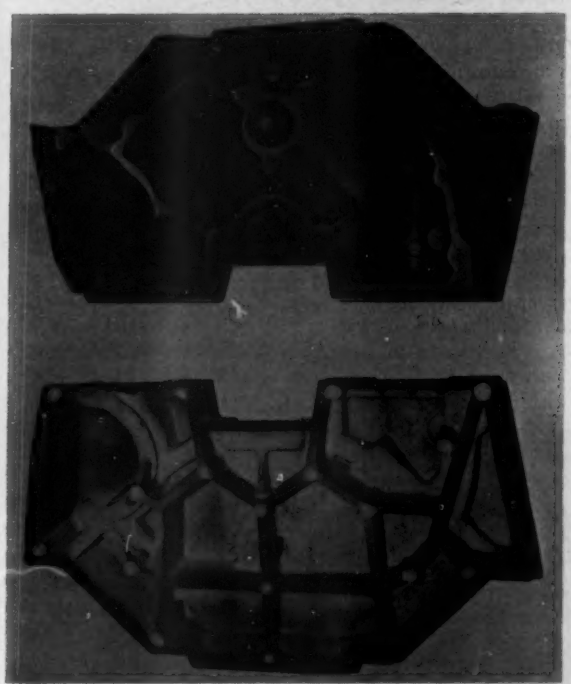


Fig. 7 — Pick-off fingers raise the cores clear of the box and ejection pins and then retract for pick off by the operator.







Figs. 8, 9 and 10 — Cutaway views of various parts that make up the unit for shell box on shuttle machine.

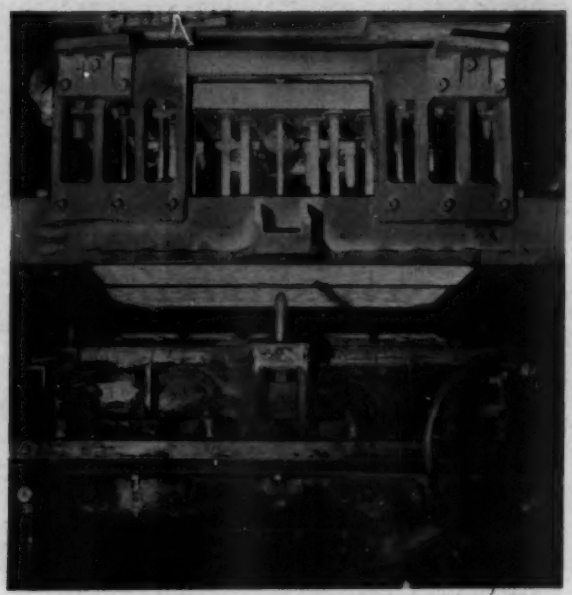


Fig. 11 — At the blow station, the hot lower box is raised against the water cooled upper half and the core is blown.

position; two retractable skirt pieces then move into place; the ram piece moves forward against the door piece; the two side pieces move into position; and the two top end barrel plugs cycle into proper location. The box is ready for blow.

The sand reservoir at the bottom of the machine has an aluminum blow plate attached with a mesh-lined heat resistant rubber gasket fastened to the box

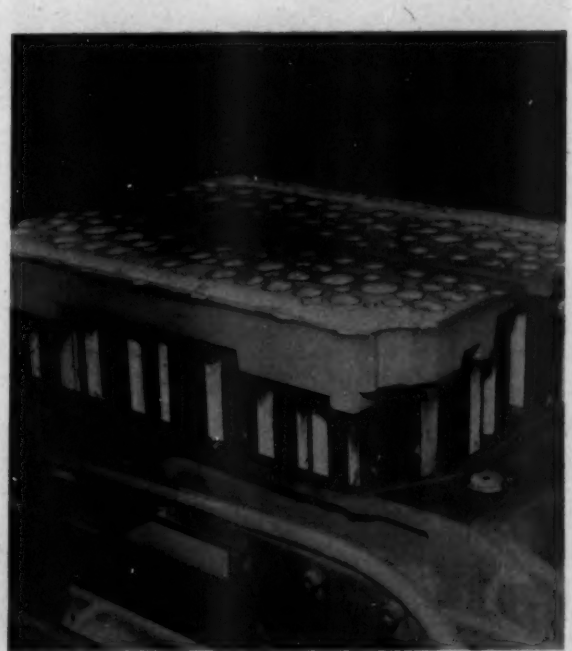


Fig. 12 — At the ejection station the box is retained by engaging cleats. The ejection piston raises the ejection plate and pins forcing the core clear of the box.

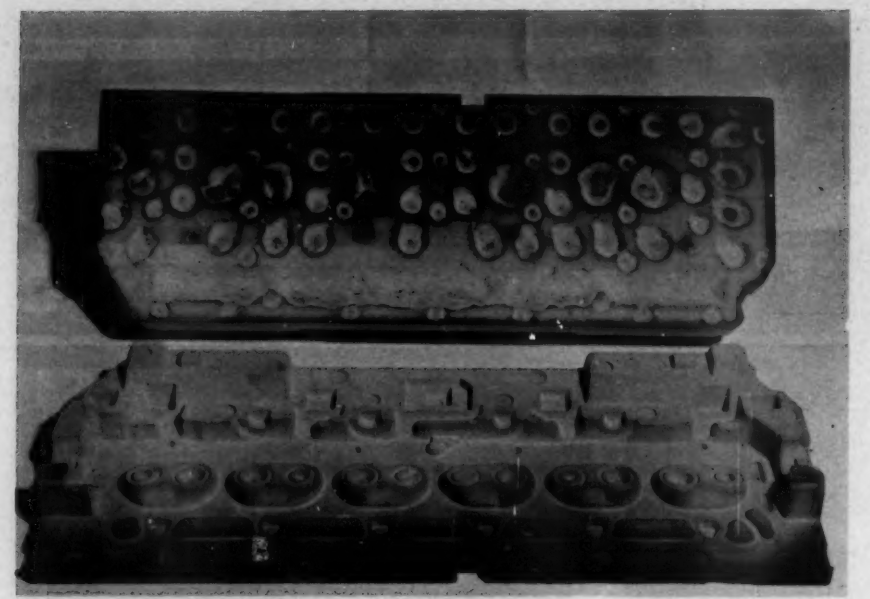


Fig. 13 — After the ejection plate lowers, the box is cycled into the pre-heat station before going into the blow station.

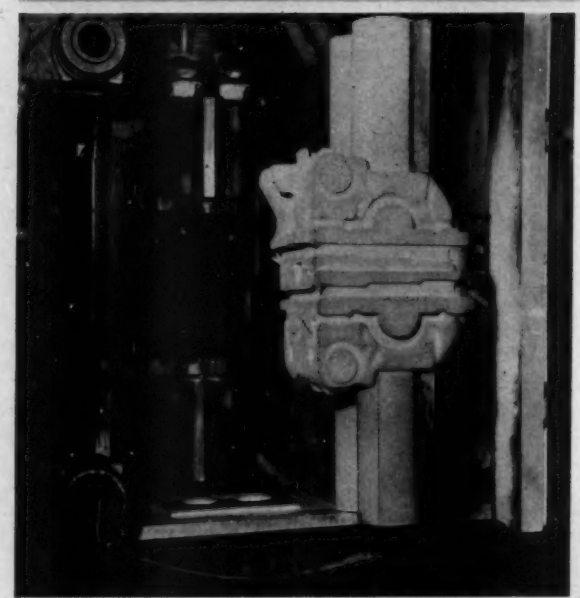


Fig. 14 — Crankcase cores are made in electrically heated core boxes on a one station roll-over type core blower in one foundry. The door piece swings into closed position, two retractable skirt pieces move into position, the ram piece moves against the door piece, two side pieces move into position and the two top end barrel plugs cycle into proper location so the box is ready for blow.

contact surface. This gasket forms a heat transfer barrier and provides a positive seal at blow. Sand reservoir and blow plate are raised against the bottom of the box, and the entire assembly rotates 180 degrees, in which position the core is blown and held for cure. The assembly returns to its original position after rocking back and fourth several times permitting uncured sand to fall back in the reservoir. Following cure time, the side pieces move back; the two skirt drawbacks retract; the ram piece retracts; and finally the door piece swings open with the core adhering to its surface. An air-operated ejection mechanism is used to facilitate removal of the core from the door piece. Figure 15 shows the core piece.

It may be necessary to get the assistance of the product engineer to make some cores practical to make by this method. Elimination of troublesome loose pieces, thin fingers of sand that may not ram well and simplification of box partings are ways in which he can help. Jobs on which the product engineer can be shown that economy and accuracy can be obtained by the use of the hot-box method will usually get his help where he can give it.



Fig. 15 — Crankcase core made in electrically heated core boxes on the one station roll-over type core blower.

# HYPEREUTECTIC ALUMINUM-SILICON CASTING ALLOYS PHOSPHORUS REFINEMENT

by F. L. Arnold and J. S. Prestley


## ABSTRACT

Optimum conditions and limitations of the phosphorus refinement process were investigated for a 16 per cent silicon casting alloy. Two types of phosphorus-containing inoculants were used, namely a commercially available nucleant containing 20 per cent red phosphorus, and a phosphor-copper alloy with 15 per cent phosphorus. Experimental melts of the alloy were cast into permanent mold and sand mold wedges giving a wide range of solidification rates. Metallographic specimens taken from an area of measured chill rate revealed the actual degree of refinement.

The optimum amount of phosphorus necessary for refinement was determined for both inoculants. Solidification rate was demonstrated to have a great influence on total primary silicon refinement, as well as some slight degree of modifying effect on the eutectic structure. Also investigated were the effects of remelting the refined alloy, temperature of the melt and holding time.

Metallographic examination of the refined silicon particles at high magnifications revealed tiny crystal shapes believed to be the phosphorus-containing nuclei. Observations supported the nucleation theory, commonly believed to be the mechanism of phosphorus refinement.


## INTRODUCTION

It has been known for some time that the addition of small quantities of sodium to hypoeutectic aluminum-silicon casting alloys greatly modifies the eutectic structure of these alloys and likewise enhances their mechanical properties. While the alloys had been well known before the modifying technique was patented by A. Pacz in 1921, this date actually marks the beginning of their commercial importance.[1]

The hypereutectic alloys have had a somewhat similar history. The first German patent was obtained in 1924, at which time considerable interest was aroused by the low coefficient of thermal expansion and good wear resistance of the high silicon alloys. However, the metallurgical difficulty in this case was that the large primary crystals of silicon adversely affected certain properties of the alloys. Considerable research was directed toward the problem of refining the silicon constituent, and eventually it was found that phosphorus performed this needed function when introduced into the molten metal. Thus the door was opened to the potential commercial use of the hypereutectic alloy system.

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## REFINEMENT THEORY

The mechanism of the refinement of primary silicon by phosphorus appears to be different from any of those generally attributed to sodium in eutectic modification. One theory is that sodium in hypoeutectic alloys reduces the interfacial tension between the eutectic alpha aluminum and the eutectic silicon constituents, increasing their interfacial angle and allowing the aluminum to envelop and arrest the growth of the eutectic silicon platelet.[2]

The theory of phosphorus refinement is that insoluble aluminum phosphide (AlP) particles are formed in the melt after addition of phosphorus, and these act as nuclei for primary silicon crystals.[3,4,6] Both have a diamond cubic crystal habit and a similar lattice constant (Si, $a_o = 5.43$ Å: AlP, $a_o = 5.45$ Å). Like sodium modification, phosphorus refinement must be accomplished in the molten metal during the process of solidification, since there is no way to refine large silicon crystals once they have been formed in a casting. Thus the process is one in which the foundry is vitally concerned.

Although the necessity for silicon refinement has been established there remains many practical questions, such as the optimum form and amount of the inoculant necessary and the effect on refinement of solidification rate, high melt temperature and several remeltings. The question of solidification rate is of particular importance because the choice of casting method may be influenced in part by required solidification rates.

## REFINEMENT NECESSITY

The hypereutectic aluminum-silicon alloys are presently being considered for applications, such as the automobile engine, in which they will be subjected to severe wear and elevated temperatures. In their contemplated use, the castings will often require considerable machining.

Figure 1 shows the effect of refinement on tensile strength of binary aluminum-silicon alloys containing 12 to 24 per cent silicon. The unrefined tensile strength drops sharply with increasing silicon to

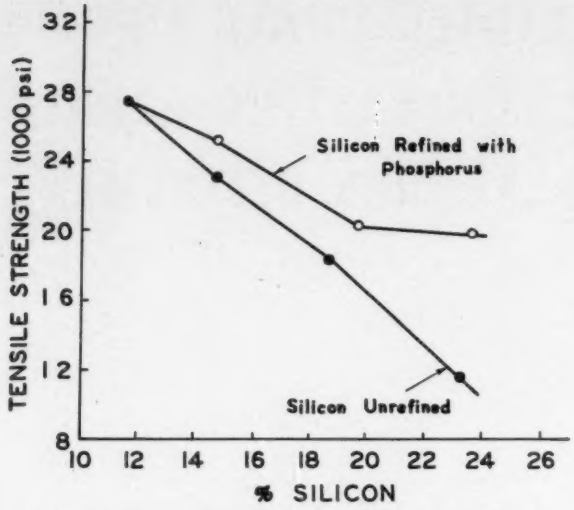



**Fig. 1 — Refinement effect on tensile strength of binary alloys containing 12-24 per cent silicon. As-cast permanent mold test bars.**

12,000 psi at 23 per cent silicon.[5] On the other hand, the tensile strength of phosphorus refined alloys drops to and levels off at 20,000 psi in the range of 18 to 23 per cent silicon.

The detrimental effect of unrefined primary silicon on machinability is relatively well known. The large crystals tend to break up under the machine tool, reducing tool life. Also, the hard silicon particles embedded in the soft aluminum matrix cause variations in resistance offered to the tool, setting up local heating and thermal expansion. This, in turn, leads to excessive local removal of metal during machining, causing variations in the dimensions of mass-produced precision parts.[6] Optimum wear resistance may require refined silicon particles for similar reasons.

## EXPERIMENTAL TECHNIQUE

The particular casting alloy used for refinement studies contained nominally 16 per cent silicon. Laboratory work has shown that other elements like copper, magnesium, manganese, iron and titanium, commonly found in hypereutectic silicon casting alloys, have no effect on primary silicon so far as amenability to phosphorus refinement is concerned.

Both a phosphor-copper alloy containing 15 per cent phosphorus, and a commercially available finely divided nucleant mixture[7] containing approximately 20 per cent red phosphorus, 10 per cent potassium titanium fluoride and 70 per cent potassium chloride were used as inoculants. Phosphorus pentachloride has been used as a refiner, but since it is expensive, produces objectionable fumes and offers no advantage over the inoculants just described, it was not considered in these investigations.

Ten-pound melts were prepared in a silicon carbide crucible heated in a gas-fired furnace. When phosphor-copper was used as the inoculant, the addition was made as shot as soon as a heel of molten metal was formed in the crucible. The metal temperature was then raised to 1500 F (815 C) to get the shot into solution. Later experiments indicated that it was not necessary to go to this temperature if the shot was crushed before addition. The alloy was then brought to 1400 F (760 C), fluxed with chlorine for 5 min and held 5 to 10 min before casting at 1400 F (760 C).

For inoculations of the commercial nucleant mixture, the addition was made at 1400 F (760 C) by plunging the inoculant, wrapped in foil, into the melt with a carbon "phosphorizer" having sufficient holes to allow intimate contact of the refining agent with the metal. The alloy was then fluxed and cast as described previously for the phosphor-copper inoculant.

In order to evaluate the effect of solidification rate of the casting, known from preliminary work to have an influence on silicon particle size, wedge-shaped permanent and baked oil sand molds were designed. After treatment, metal from each experimental melt was poured into these molds. Solidification time was determined with calibrated iron-constantan thermocouples inserted into the castings near the tip, at the center and near the top of the wedge. Solidification rates (between the liquidus and solidus temperature) varied from 61 F (16 C) to 1650 F (890 C)/min (Fig. 2). These three areas of known solidification rate were chosen for metallographic examination.

## RESULTS AND DISCUSSION

### *Optimum Phosphorus Addition — Phosphor-Copper*

One of the important conclusions to be derived from this work was the amount of the 15 per cent phosphor-copper alloy and the commercial inoculant necessary to effect optimum refinement. Optimum refinement, in this case, is defined as the finest primary crystal size obtainable within the range of inoculant additions studied.

In the permanent mold castings refined with phosphor-copper the addition of only 0.01 per cent phosphorus to the casting alloy yielded silicon particles only slightly refined, as compared with an alloy containing no phosphorus at all. However, a total addition of 0.05 per cent phosphorus produced optimum refinement, and further additions up to 0.17 per cent phosphorus provided no further reduction in particle size. Samples with phosphorus additions of between 0.01 and 0.05 per cent were then prepared, and it was found that 0.02 per cent was the minimum addition for which optimum refinement could be expected. Figure 3 shows the coarse silicon particles in an alloy containing no phosphorus, and the effect on silicon particle size of increasing additions of phosphorus.

In the more slowly solidified sand castings refined with phosphor-copper the same refinement trend was found with increasing phosphorus additions, except that the amount needed for optimum refinement was shifted from 0.02 to 0.05 per cent for the permanent mold castings to from 0.05 to 0.08 per cent for the sand castings.

The addition of phosphorus by phosphor-copper alloy was found to be quite simple and reliable, as compared with other types of inoculants. Since the addition can be made when the metal begins to melt, a separate and time-consuming step is avoided

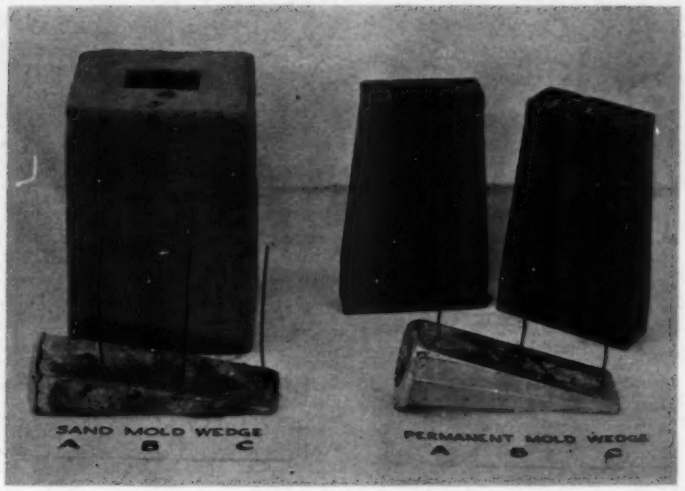



Fig. 2 — *Left* — Baked sand wedge mold and casting showing thermocouple locations. *Right* — Iron permanent mold and casting. The table lists solidification rates for respective thermocouple locations.

| Mold Type | Position | Solidification, F (C)/min |
|---|---|---|
| Sand | A | 61 ( 16) |
| Sand | B | 89 ( 32) |
| Sand | C | 114 ( 46) |
| Permanent | A | 286 (141) |
| Permanent | B | 474 (246) |
| Permanent | C | 1650 (899) |

at a later point. Although the use of phosphor-copper does add copper to the casting alloy, it may be held to a minimum of only 0.06 per cent for each 0.01 per cent phosphorus added. This may be tolerable in an alloy which has a relatively high copper content to begin with, and it is possible that a part of the nominal copper content of the alloy could be added in this way.

### *Optimum Phosphorus Addition — Commercial Inoculant Mixture*

Additions of from 0.20 to 1.25 per cent of a commercial refining inoculant containing 20 per cent red phosphorus revealed that an addition of 0.40 per cent was necessary for optimum refinement of both permanent mold and sand casting. The same refining trend was observed with this inoculant as with phosphor-copper, namely, that little refinement was obtained until a certain percentage of inoculant, (in this case 0.40 per cent) was added. With this amount, a fine silicon particle size was obtained which varied only slightly with additional amounts of phosphorus.

Figure 4 shows the effects of additions of increasing amounts of the commercial phosphorus-containing inoculant. This material has the advantage over phosphor-copper of not adding copper to the alloy, but it must be added to the melt in a separate operation after the metal is molten. Phosphor-copper may be added to the crucible at the beginning of the melting process.

### *Solidification Rate Effect*

The rate at which the 16 per cent silicon alloy solidified was found to have a marked effect on primary silicon particle size in alloys both with and without phosphorus additions. With no phosphorus, however, even at the extremely rapid solidification rate [1650 F (899 C)/sec] at the tip of the permanent mold wedge casting, refinement was considerably less than satisfactory. This indicates that some means of artificial nucleation is essential for refinement of commercial castings. Figure 5 compares the microstructures of sand [solidified at 89 F (32 C)/min] and permanent mold [solidified at 1650 F (899 C)/min] wedge sections cast without phosphorus additions.

In general, a rapid solidification rate, such as was obtained in the permanent mold castings, coupled with at least the minimum addition of phosphorus, as discussed previously, produced the best refinement. Figure 6 compares the refinement of sand cast [solidified at 89 F (32 C)/min] wedge section with that of a permanent mold wedge section solidified at 1650 F (899 C)/min, both having additions of 1.0 per cent of the commercial refining inoculant.

### *Remelting Effect*

The effect of remelting previously inoculated and cast metal was considered in these investigations because of the practical situation where scrap metal may be melted several times in a foundry. One remelt appeared to have no effect on the refined structure, while two to four remelts had a slight coarsening effect on the silicon particles. This was most noticeable in the more rapidly chilled section of the permanent mold wedge where refinement was excellent to begin with. Figure 7 illustrates these observations (remelted metal was fluxed with chlorine before casting).

### *Holding Time and Melt Temperature Effect*

Since in the foundry operation it is impossible to insure that inoculated molten metal will always be immediately poured into molds, the question of the effect of holding time (after inoculation) on refinement becomes important. Laboratory work with the wedge castings indicated that properly inoculated metal could be held for at least 1½-hr at pouring temperature with no great change in silicon refinement, although there existed a tendency towards slightly coarser particles.

When melts were inoculated at 1400 F (760 C) with the commercial nucleant powder, the subsequent heating to still higher temperatures revealed a trend toward slight coarsening in the range of 1500 F (815 C) to 1700 F (927 C). Thus, metal temperature is not critical for this inoculant at temperatures normally used in melting practice for this alloy. In these studies, the melt was cooled again to the

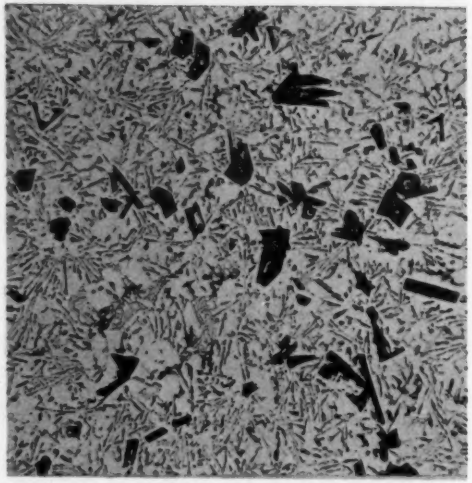


Fig. 3a — No phosphorus addition. Silicon particle coarse.

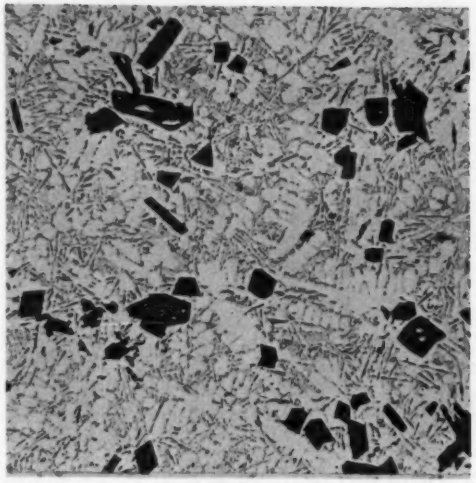


Fig. 3b — Addition of 0.01 per cent phosphorus. Silicon refined only slightly.

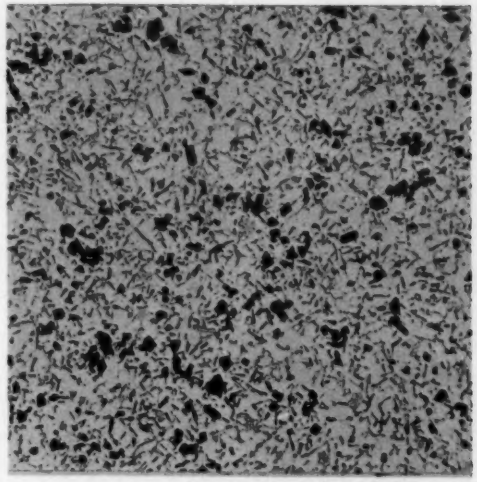


Fig. 3c — Addition of 0.02 per cent phosphorus. Optimum refinement.

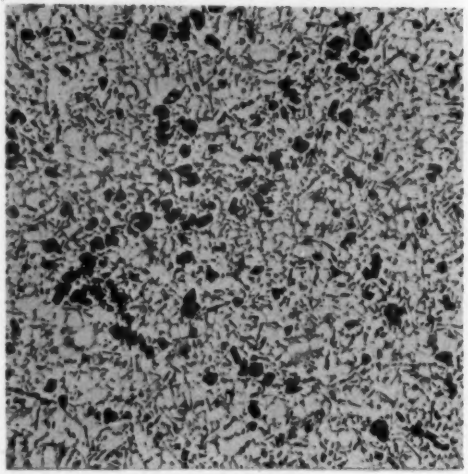


Fig. 3d — Addition of 0.05 per cent phosphorus. Optimum refinement.

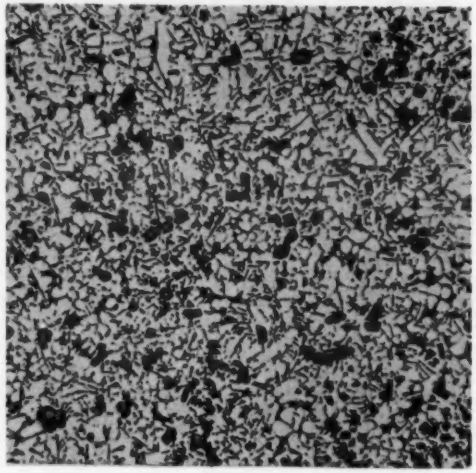


Fig. 3e — Addition of 0.08 per cent phosphorus. Optimum refinement.

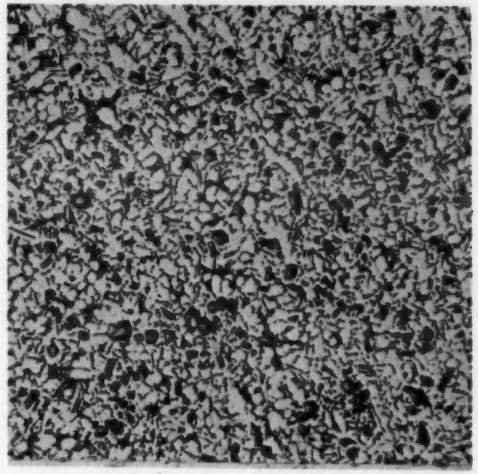


Fig. 3f — Addition of 0.17 per cent phosphorus. Optimum refinement.

Fig. 3 — P-Cu additions effect on primary silicon refinement. All samples taken from *B* section of permanent mold wedge, solidified at 474 F (246 C)/min. Unetched. 100 ×.



Fig. 4a — No phosphorus addition. Silicon particle coarse.

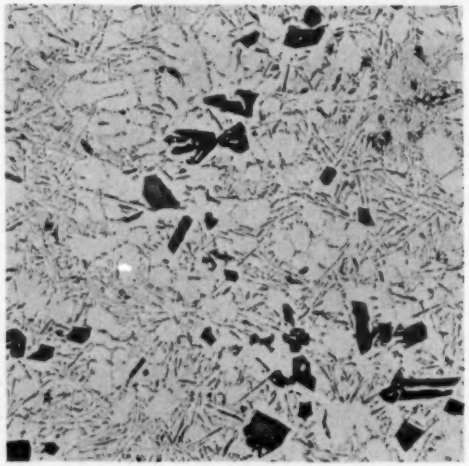


Fig. 4b — 0.20 per cent nucleant mixture. Silicon refined only slightly.

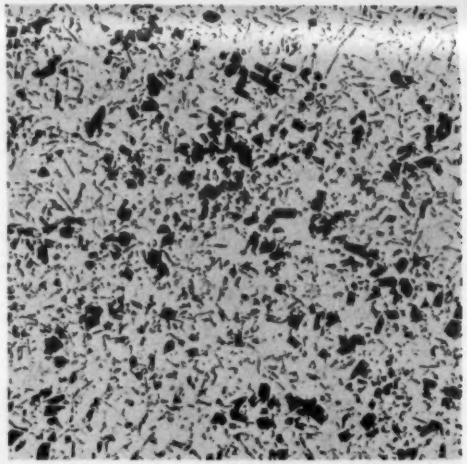


Fig. 4c — 0.40 per cent nucleant mixture. Silicon particles well refined.

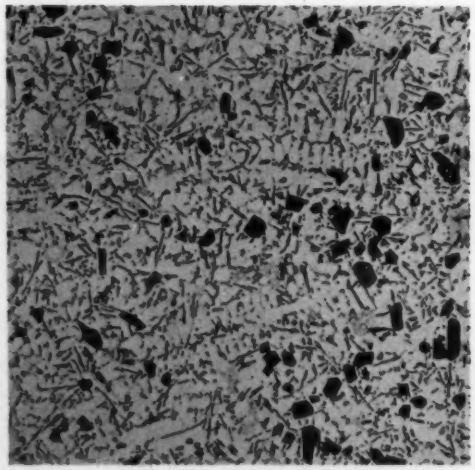


Fig. 4d — 0.60 per cent nucleant mixture. Silicon particles well refined.

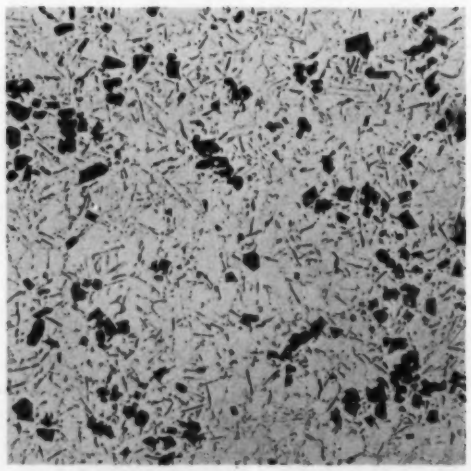


Fig. 4e — 0.80 per cent nuclean mixture. Silicon particles well refined.

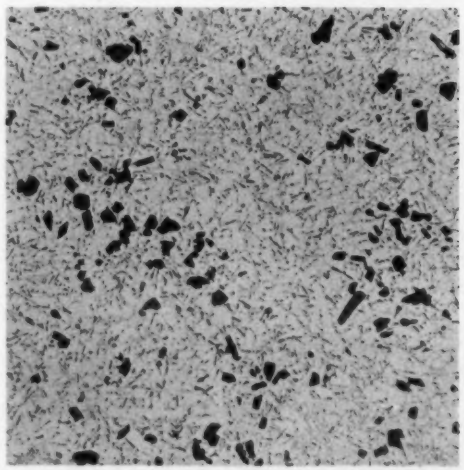


Fig. 4f — 1.0 per cent nucleant mixture. Silicon particles well refined.

Fig. 4 — Commercial nucleant mixture effect on primary silicon refinement. All samples taken from *B* section of permanent mold wedge, solidified at 474 F (246 C)/min. Unetched. 100 ×.

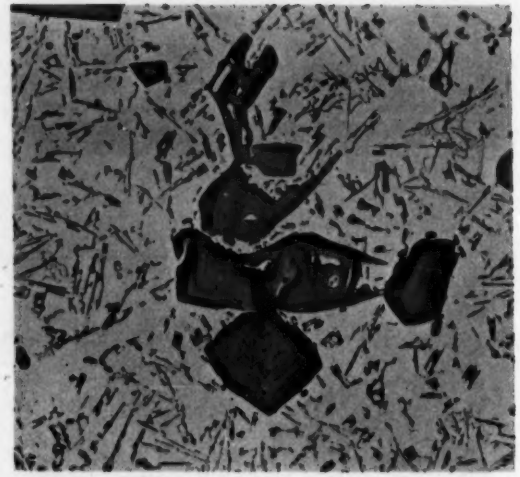


Fig. 5a — Solidification rate 89 F/mn in sand mold *B* section. Coarse silicon particles.

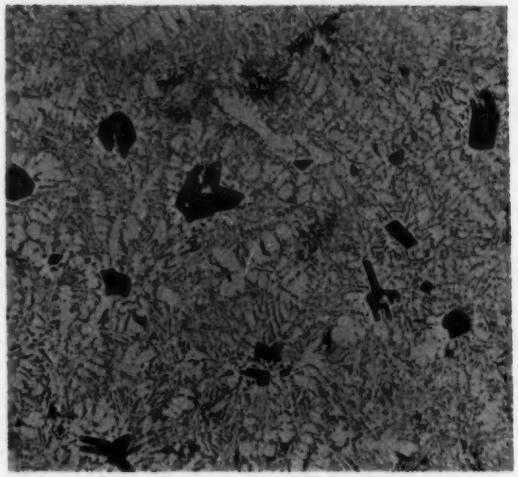


Fig. 5b — Solidification rate 1650 F (899 C)/min in permanent mold *C* section. Silicon finer, but still not well refined.

Fig. 5 — Solidification rate effect on alloy with no phosphorus addition. Unetched. 100 ×.

Fig. 6 — Solidification rate effect on alloy with addition of 1.0 per cent commercial nucleant mixture containing phosphorus. Unetched. 100 ×.

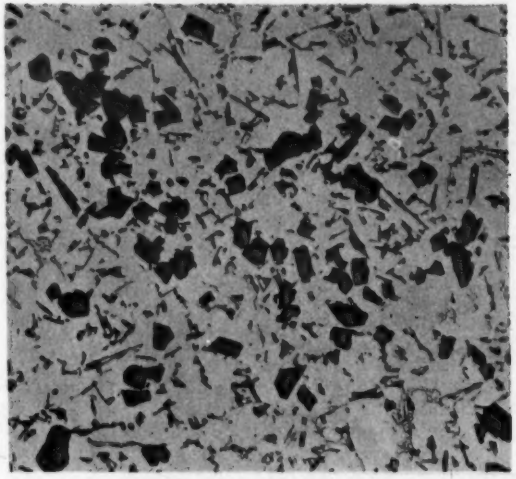


Fig. 6a — Solidification rate 89F (32 C)/min in sand mold *B* section. Silicon particles fairly well refined.

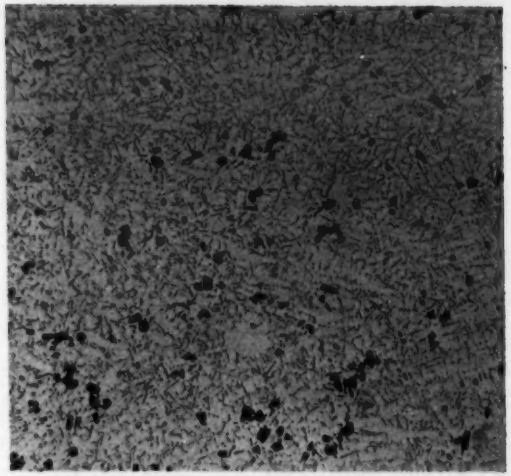


Fig. 6b — Solidification rate 1650 F (899 C)/min in permanent mold *C* section. Excellent refinement.

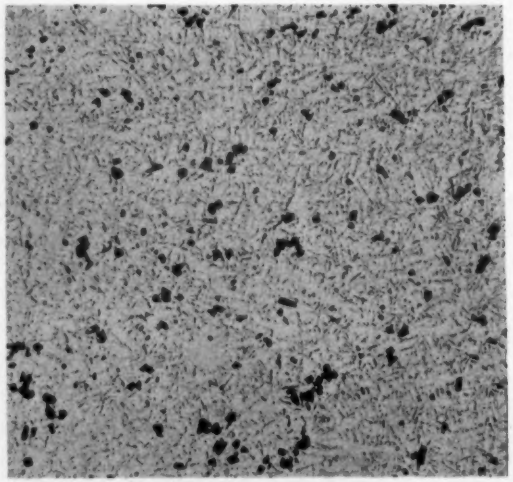


Fig. 7a — Original melt. Good refinement.

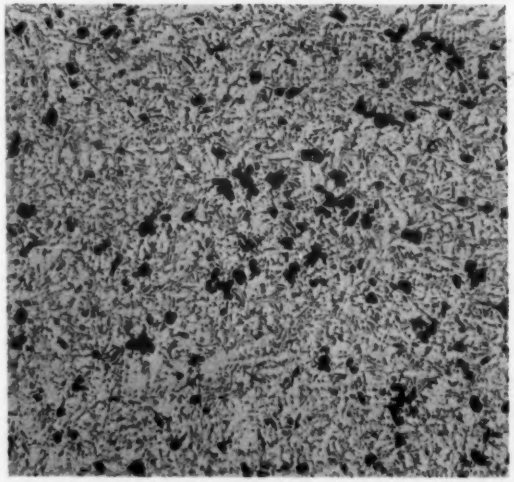


Fig. 7b — After four remelts silicon particles slightly coarser than original melt.

Fig. 7 — Four remelts effect on alloy refined with 1.0 per cent commercial nucleant mixture. Samples taken from *C* section of permanent mold wedge solidified at 1650 F (899 C)/min. Unetched. 100 ×.

Fig. 8 — Commercial nucleant mixture addition effect on eutectic structure. Samples taken from *C* section of sand mold wedge solidified at 61 F (16 C)/min. Unetched. 100 ×.

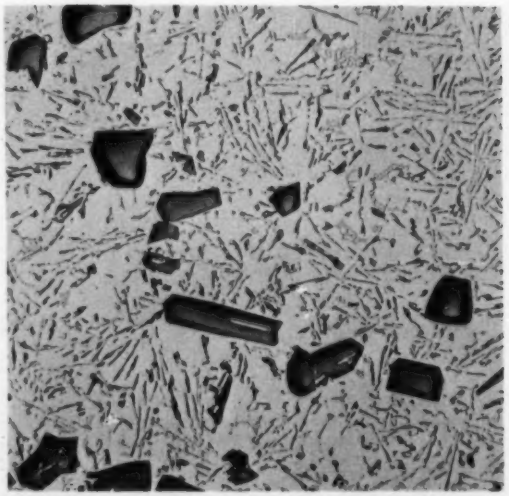


Fig. 8a — No phosphorus addition. Coarse, acicular eutectic structure.

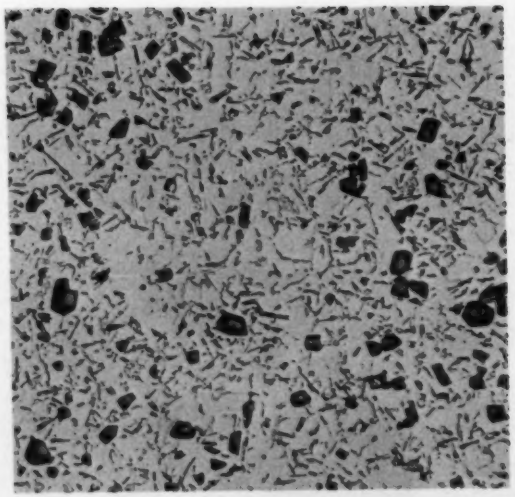


Fig. 8b — 1.25 per cent commercial nucleant addition. Eutectic structure somewhat modified.

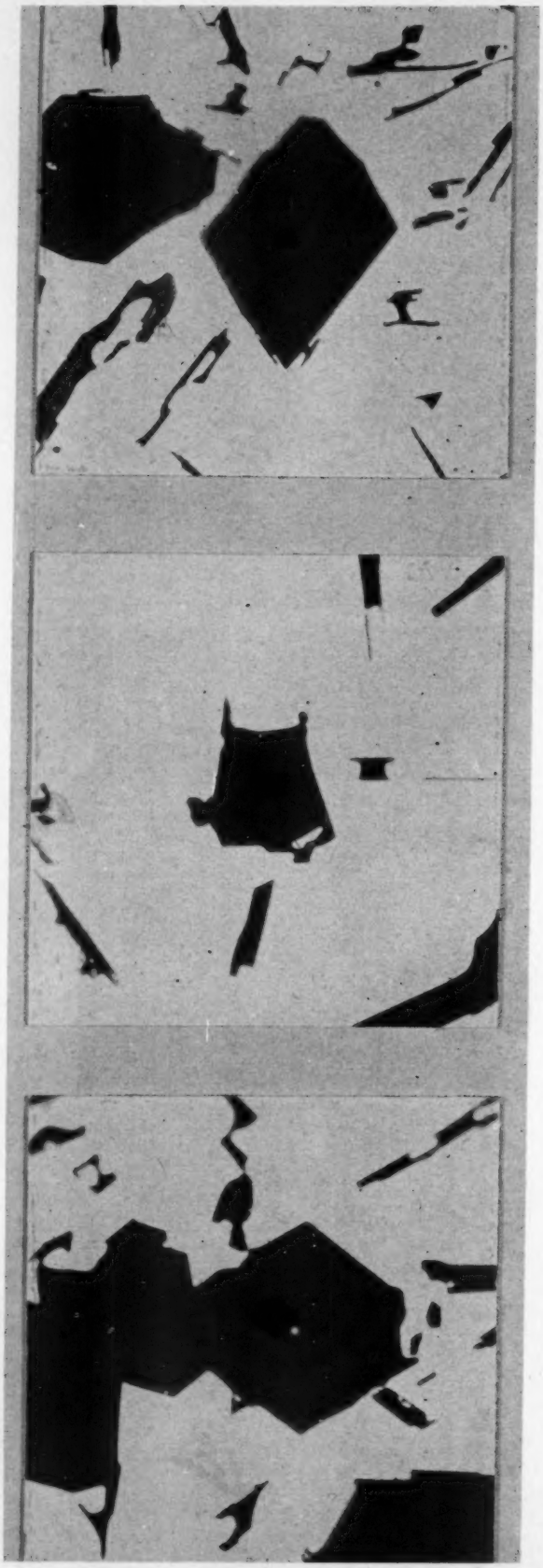


Fig. 9 — Primary crystals of silicon with nuclei of similar configuration. Unetched. 1500 ×.

normal casting temperature of 1400 F (760 C) for pouring.

### *Eutectic Structure*

The fineness (or degree of modification) of the silicon eutectic structure in all of the phosphorus-refined samples studied was heavily dependent on solidification rate, as normally expected, but also appeared roughly proportional to the amount of phosphorus addition by either inoculant. In no case, however, was a structure obtained which approached that of sodium-modified metal. Figure 8 illustrates the effect of the commercial type inoculant on the eutectic structure of sand castings.

## REFINEMENT MECHANISM

The mechanism of refinement by phosphorus, as mentioned previously, has been reported as a nucleation phenomenon. It is theorized that the formation of a crystal of the compound aluminum phosphide (AlP), having the same crystal structure as silicon and a similar lattice parameter, facilitates the nucleation of a primary silicon crystal onto its surfaces. Little data about the nature of occurrence of these nuclei crystals appear in the literature. Confirmation of their presence is necessary as the first step toward understanding the mechanism of refinement. The size of these "seed" crystals is of particular interest here in correlating the amount of phosphorus added to the amount of phosphorus retained in the primary crystals.

In order to determine whether nuclei crystals of detectable sizes were present, the polished samples of the phosphorus refined alloy were examined at about 2000 magnifications. Crystal shapes with configurations almost identical to the host silicon particles were observed in the approximate centers of a small proportion of the primary silicon crystals (Fig. 9).

Since there was a possibility that the shapes observed were actually some other phenomenon rather than crystals of aluminum phosphide (AlP), a sample of the alloy was cast without phosphorus addition. The solidification rate was rapid enough to yield primary crystals of a size approaching that of the phosphorus refined alloy. The rapidly chilled sample did not contain the crystal shapes observed in the phosphorus refined samples. It was, therefore, concluded that nucleating crystals were associated with phosphorus refinement.

Preliminary electron microprobe x-ray analysis confirmed the presence of phosphorus at the seed crystal site. This phosphorus presumably exists as the compound aluminum phosphide (AlP). However, more precise analysis is necessary to provide conclusive evidence as to the exact composition of the nucleating crystals.

### *Added vs. Retained Phosphorus*

To obtain more information about the nuclei crystals observed, the *B* group of permanent mold samples refined with commercial nucleant powder were selected for closer study. Thirty measurements

on each sample gave an average nucleus diameter of from about 1.0 to 2.0 microns. The corresponding ratio of average nucleus diameter to silicon crystal diameter ranged from about 0.1 to 0.2. Samples refined with phosphor-copper, although given only cursory examination, were of about the same order of magnitude. Measurements were taken directly from the polished surfaces, with no correction for the probability that most of the nuclei would not be polished through the maximum dimension.

From these data, a first approximation of the amount of phosphorus contained in the nucleus (as a percentage of the alloy) was made assuming:

1) all of the primary silicon particles to contain a nucleus crystal.
2) the stoichiometric relationship of aluminum phosphide (AlP).
3) 4 per cent primary silicon.
4) the shape of both crystals to be cubes.

Phosphorus contents of about 0.001-0.010 per cent were indicated from the calculations. This amount was considerably less than the 0.04-0.25 per cent phosphorus added to the samples. Since chemical analysis gave 0.001 per cent or less, the obvious conclusion is that a considerable portion of the phosphorus is lost between the addition and casting operations.

Further, according to the calculated percentages, the amount of phosphorus retained did not continually increase with increasing additions. The variation was similar to that observed in this group of samples when comparing the degree of refinement to the percentage of inoculant addition. Both of these variations may be due to the possibilities for unpredictable losses inherent in adding the fine powder to molten metal.

### *Accuracy of Calculated Phosphorus*

Within an order of magnitude, the agreement between calculated and analyzed phosphorus is fairly good, considering the assumptions necessary for calculation and the fact that the chemical analytical techniques were designed to include phosphorus in the matrix as well as in the primary silicon. An accurate analytical method, which will determine the amount of phosphorus in the primary silicon along with other improvements in measuring techniques, would provide a better understanding of the physical mechanism of nucleation. The present data, however, support the nucleation theory of phosphorus refinement. They also indicate that, because of losses during alloying, sufficient excess of phosphorus must be added to insure the retention of the amount necessary for nucleation.

## CONCLUSIONS

The minimum percentage phosphorus addition required for a high degree of refinement appears to lie within a narrow range for both the commercial powder and phosphor-copper inoculants. Below this range of addition, little or no refinement is realized. Above this range, little increase in refinement is affected up to the maximum additions studied.

A high degree of primary silicon refinement can be obtained in a 16 per cent silicon permanent mold cast alloy using a minimum of:

a) 0.40 per cent commercial phosphorus-containing nucleant studied.
b) 0.02 per cent phosphorus added as 15 per cent phosphorus-copper alloy.

Somewhat more phosphorus is required to produce good refinement in the slower cooled sand cast wedges than in permanent mold castings when added as phosphor-copper alloy. Using the phosphor-copper inoculant, this amounted to 0.05 to 0.08 per cent phosporus instead of 0.02 per cent. Silicon particle size, even with optimum additions, is much coarser in sand than in permanent mold castings.

The metallic 15 per cent phosphorus-copper is easier to add to the melt than the commercial nucleant mixture, but it does add copper to the alloy. This amounts to 0.06 per cent copper for each 0.01 per cent phosphorus added.

The remelting of ingots containing optimum phosphorus additions does not have any noticeable effect on the degree of refinement until the third and fourth remelting. Even then, the coarsening is only slight.

For alloys inoculated with the phosphorus-containing commercial nucleant, maximum melt temperatures above 1500 F (815 C) produce some tendency towards coarsening of primary silicon particles.

Eutectic silicon fineness is dependent on the rate of solidification although phosphorus additions do have a slight modifying effect. However, both of these factors are not sufficient to even approach the degree of modification obtainable with sodium.

The presence of crystals in the centers of refined primary silicon crystals indicates that refinement is a nucleation phenomenon, as theorized.

## ACKNOWLEDGMENT

The valuable contributions of the following personnel of Met. Rsch. Laboratories, Reynolds Metals Co., are gratefully acknowledged – G. E. Stein, R. S. Mapes, G. P. Koch and J. L. Jorstad.

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# FEEDING DISTANCE OF BARS IN INVESTMENT MOLDS

by H. Present and H. Rosenthal


## ABSTRACT

A study has been made of feeding range in investment molds. Armco iron, A.I.S.I. 1020 and A.I.S.I. 1095 steel were used to determine the effect of mold temperature and pouring temperature. Only the case of a round bar fed by an infinite riser was considered. The data were found to be linear for the relationship between length of sound bar and bar diameter over the range between ¼-in. and one in. Feeding range increased with mold temperature, pouring temperature and carbon content of the steel.


## INTRODUCTION

Feeding distance of risers for steel bars and plates has been extensively investigated for sand molds. These data are of fundamental interest to those concerned with the gating of castings. Some of the newer ceramic mold casting processes, however, employ conditions which are not compatible with the data developed for sand casting. One such process is investment casting which uses a water-free ceramic mold at temperatures ranging up to 2000 F. These conditions are obviously unlike those of a green sand mold.

It was the purpose of this work to obtain feeding distance data which are applicable to these hot, dry, ceramic molds. It is realized that these data are by no means complete since they relate only to round bars. However, in the absence of other published data, it was felt that even this limited information might be useful to those interested in this field.

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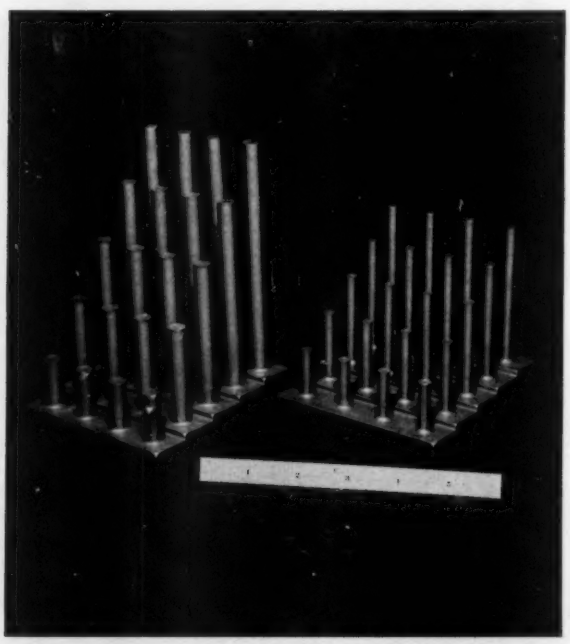


Fig. 1 — Wax patterns for ¼-in. and ⅜-in. bars of graduated length.

## EXPERIMENTAL PROCEDURE

Figure 1 illustrates the design of the test castings for the ¼-in. and ⅜-in. bars. What is shown are wax patterns which eventually form the mold cavity. The mold, when ready for pouring, has the corresponding cavity with the bars vertically aligned and attached directly to the pouring basin. The basin acts as an infinite riser in feeding the bars.

The table lists the various bar parameters investigated. After casting, the bars were sawed off from the riser and radiographed. From the x-ray examination, it could be determined how long a given type of bar could be before it was affected by shrinkage. To make this determination, more than one bar was examined for each individual condition. The number of specimens ranged from two for the one in. bars to five for the ¼-in. bars.

BAR PARAMETERS INVESTIGATED

| Bar Diameter, in. | No. of Bars Per Mold | Bar Length Increments, in. |
|---|---|---|
| ¼ | 25 | ½ |
| ⅜ | 25 | ½ |
| ½ | 16 | ½ |
| ¾ | 9 | ½ |
| 1 | 4 | 1 |

Molds were prepared by standard processing techniques. The pre-dip slurry contained silica flour bonded by ethyl silicate. The backup investment consisted of coarser grades of silica bonded by magnesium pyrophosphate. Flasks were 5 in. square cans of stainless steel. After room temperature drying, the molds were fired 2 hr at 400 F (204 C); one hr at 700 F (371 C); 2 hr at 1100 F (593 C); and 8 hr at the final

mold temperature. For final mold temperature of 1200 F (649 C), it was necessary to heat at 1600 F (871 C) for 4 hr to complete oxidation of the wax residue before dropping the furnace back to 1200 F (649 C).

The steels were melted in a MgO lined, 50 kw, induction furnace. Ferrosilicon and ferromanganese were added to deoxidize the melt. Aluminum was added just before tapping. The metal was poured into a ladle prior to the final transfer to the molds.

To prevent wide variations in pouring temperature, not more than three molds were poured at one time. The maximum variation in pouring temperature was ± 15 F (8 C), as determined by an optical pyrometer. After pouring, exothermic hot top compound was applied to the molds. The molds were shaken out after air cooling to room temperature.

The bars were radiographed using a 220 kv unit with a 2.3 mm focal spot and 0.01 in. lead tube filter. Type M film was used at a distance of 48 in. from the focal spot; 0.005 in. lead screens were also used. Films were examined by a high intensity viewer.

## RESULTS AND DISCUSSION

The effect of mold temperature (Fig. 2) was investigated on A.I.S.I. 1020 steel. Pouring temperature effects were determined on armco iron (Fig. 3). In addition, the effect of carbon content (Fig. 4) was studied with a fixed mold and pouring temperature. The three alloys of differing carbon content were armco iron, A.I.S.I. 1020 and A.I.S.I. 1095 steel. An addition of 0.1 per cent silicon was made to the armco iron to improve castability.

Adams[1] has pointed out that feeding distance will increase with increasing mold temperature if the section is uniform. From Fig. 2, it can be seen that there is little difference in feeding distance for molds at 1600 F (871 C) over molds at 1200 F (649 C). With molds at 2000 F (1093 C), there is a sizable increase of about one-third. It is probable that not all of this improvement with higher mold temperature is real since high mold temperatures tend to promote dispersed shrinkage. Although the sensitivity of the radiographic technique is about 1.5 per cent, detection of porosity is hindered by fine dispersion of the voids.

Figure 3 indicates that an increase in pouring temperature from 3000 F (1649 C) to 3300 F (1816 C) for armco iron more than doubled the feeding range. Here again, the dispersal of shrinkage may account for part of this increase.

The beneficial effect of carbon in feeding distance is shown in Fig. 4. The A.I.S.I. 1020 feeding distance is twice that observed with the armco iron, while the A.I.S.I. 1095 is almost three times that of the armco iron. These results must be interpreted in the light of Fig. 3, which illustrates the effect of pouring temperature, and against the fact that carbon depresses the freezing point of the steel. If the three alloys were compared on the basis of equal superheat pouring temperature, the feeding distances would be much more nearly alike.

In plotting maximum feeding distance against bar diameter, the data appeared to yield a simple linear relationship. These lines all converge toward the origin of the coordinate axes as might be expected.

Bishop, Myskowski and Pellini[2] have shown that, for green sand molds, there is a parabolic relationship between feeding distance and bar thickness. Their work was on a 0.25-0.35 carbon steel with bar sections varying from 2 x 2 in. to 8 by 8 in. The formula is

$$L = 6\sqrt{T}$$

where $L$ is the maximum feeding distance.
$T$ is the bar thickness.

This formula predicts a feeding distance of 6 in.

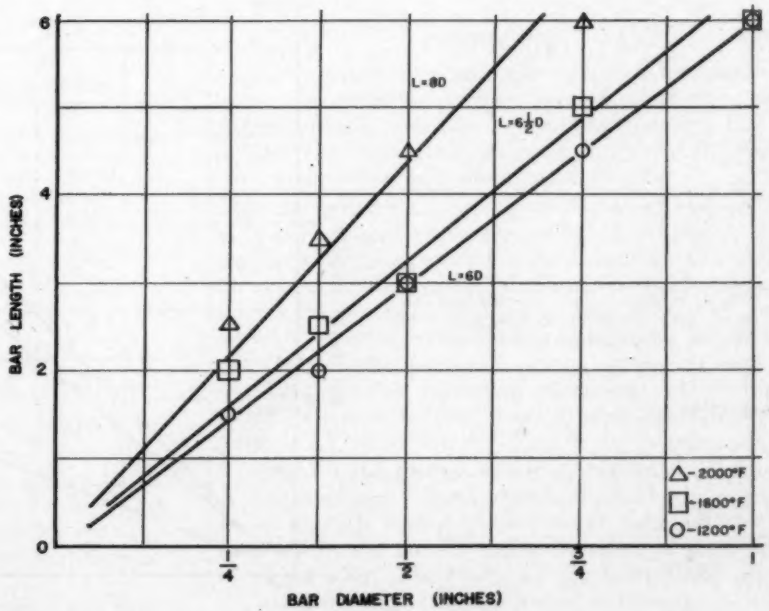



Fig. 2 — Mold temperature effect on A.I.S.I. 1020 steel. Pouring temperature, 3000 F (1649 C).

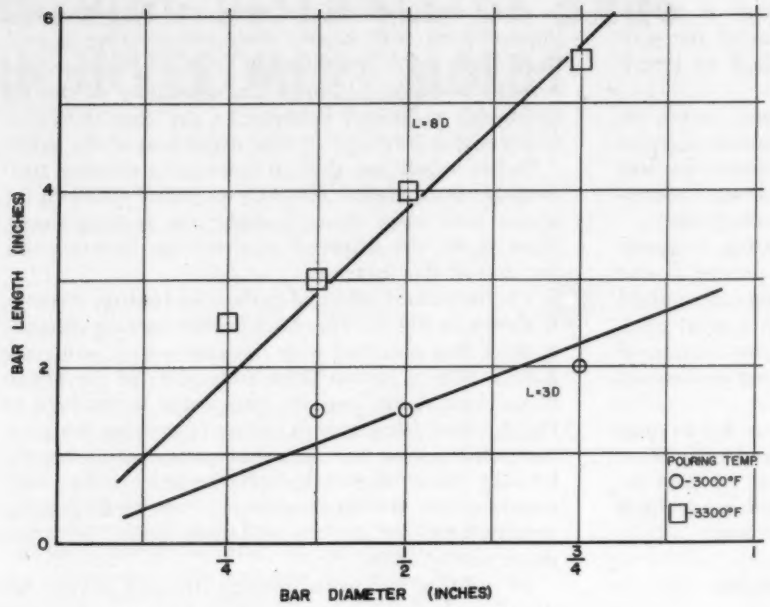



**Fig. 3 — Pouring temperature effect on armco iron. Mold temperature, 1600 F (871 C).**

for a one in. thick bar. Comparison can be made with Fig. 2, since the alloy is nearly the same. It can be seen that, for the 1200 F (649 C) mold temperature, the feeding distance is correctly predicted by the Pellini formula. However, the parabolic relationship does not hold for the smaller diameters.

## CONCLUSION

With carbon steels, maximum feeding distance for round bars is increased by:

1. Higher mold temperatures.
2. Higher pouring temperatures.
3. Higher carbon content (for the same pouring temperature).

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1. C. M. Adams, "Gating and Risering of Investment Castings," *Metal Progress, vol. 69,* no. 5, p. 38 (May 1956).
2. H. F. Bishop, E. T. Myskowski and W. S. Pellini, "The Contribution of Riser and Casting End Effects to Soundness of Cast Steel Bars," AFS TRANSACTIONS, *vol. 59,* p. 171 (1951).

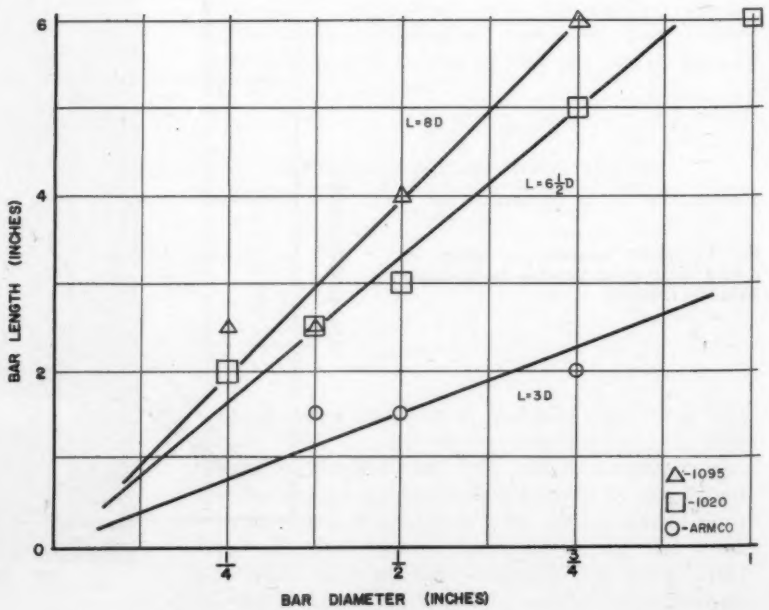



**Fig. 4 — Alloy effect on feeding distance. Mold temperature, 1600 F (871 C). Pouring temperature, 3000 F (1649 C).**

# SYSTEMATIC APPROACH TO SAND DESIGN AND CONTROL

## *Progress report 4 — wood flour*

by A. H. Zrimsek and G. J. Vingas


### ABSTRACT

**In the first three progress reports of this series, basic data were presented which described quite thoroughly the function of clay, water content and mulling on the physical properties of simply clay-sand-water systems.**

**Other than in non-ferrous shops, such simple systems are seldom used in normal foundry practice, but these systems serve as the basis of comparion for the more complex systems containing such ingredients as wood flour, cereal, seacoal, and silica flour. To understand the functions of these additives, it is necessary to first understand the base systems of sand, clay and water.**

**Using the first three progress reports as the basis of comparison, this report will deal with the effect of wood flour as a variable on the physical properties of clay-sand-water systems.**


### EXPERIMENT

The results and observations reported are based on data collected on 42 systems having different combinations of clay per cent, water per cent, wood flour per cent and mulling time. A total of 230 different sand mixes were studied.

**SCREEN ANALYSIS OF BASE SAND USED**

| U.S. Standard Sieve No. | Retained, % |
|---|---|
| 20 | 0.0 |
| 30 | 2.6 |
| 40 | 19.3 |
| 50 | 30.1 |
| 70 | 24.4 |
| 100 | 15.1 |
| 140 | 5.6 |
| 200 | 2.4 |
| Pan | 0.6 |

As in previous reports, a base sand, as described in the table, was used. All data were obtained from sands mulled in an 18 in. laboratory muller with a batch size of 4500 grams. Clay additions of 4.75, 7.45 and 10 per cent for bentonite and 10 or 15 per cent for fireclay were made. At each clay content, additions of wood flour in amounts of 0.75, 1.5 and 3.0 per cent were added. With each clay-wood flour combination at least five different water contents were used.

The mixture was dry mulled 15 sec, water was added, and the mixture wet mulled for 2, 4 and 6 min. The sands produced were placed in airtight polyethylene bags. Green testing was done immediately and dry testing done after specimens had been oven dried at 220-230 F for periods not less than 5 hr. Data were collected on specimens rammed at 1, 3, 5, 7 and 10 rams.

### RESULTS

As in previous reports of this series, results are plotted graphically. The extremely large volume of data collected, if presented in its entirety, would require 84 pages of graphs such as Figs. 1-6b. Presentation of the total data would not only be unwieldy and confusing, but is actually not required to demonstrate the principles involved. The data omitted, however, do support the conclusions to be drawn, but are repetitive.

### DISCUSSION

The function of wood flour additions to foundry sands, according to popular opinion, is to increase green strength, lower dry strength and improve flowability. This conception, although often true in practical application, is a bit over-simplified.

In the initial three reports of this series, the effect of water content on green and dry properties of clay-sand mixtures was established. In Figs. 1, 4 and 11, several of the systems of those reports are reproduced. Figures 1a-1d depict the effect of 0, 0.75, 1.5 and 3.0 per cent wood flour additions on the 4.75 per cent western bentonite system. It is noted that the main effect of increasing wood flour addition is to shift the peak strength point to positions of higher water.

Although the general strength level of the curves is shifted upward slightly, the main cause of increased green strength with increased wood flour additions is caused by the shift of the entire curve to positions of higher water. The curves for green shear and density also shift to positions of higher water content. An examination of Figs. 2a and 2b shows similar effects

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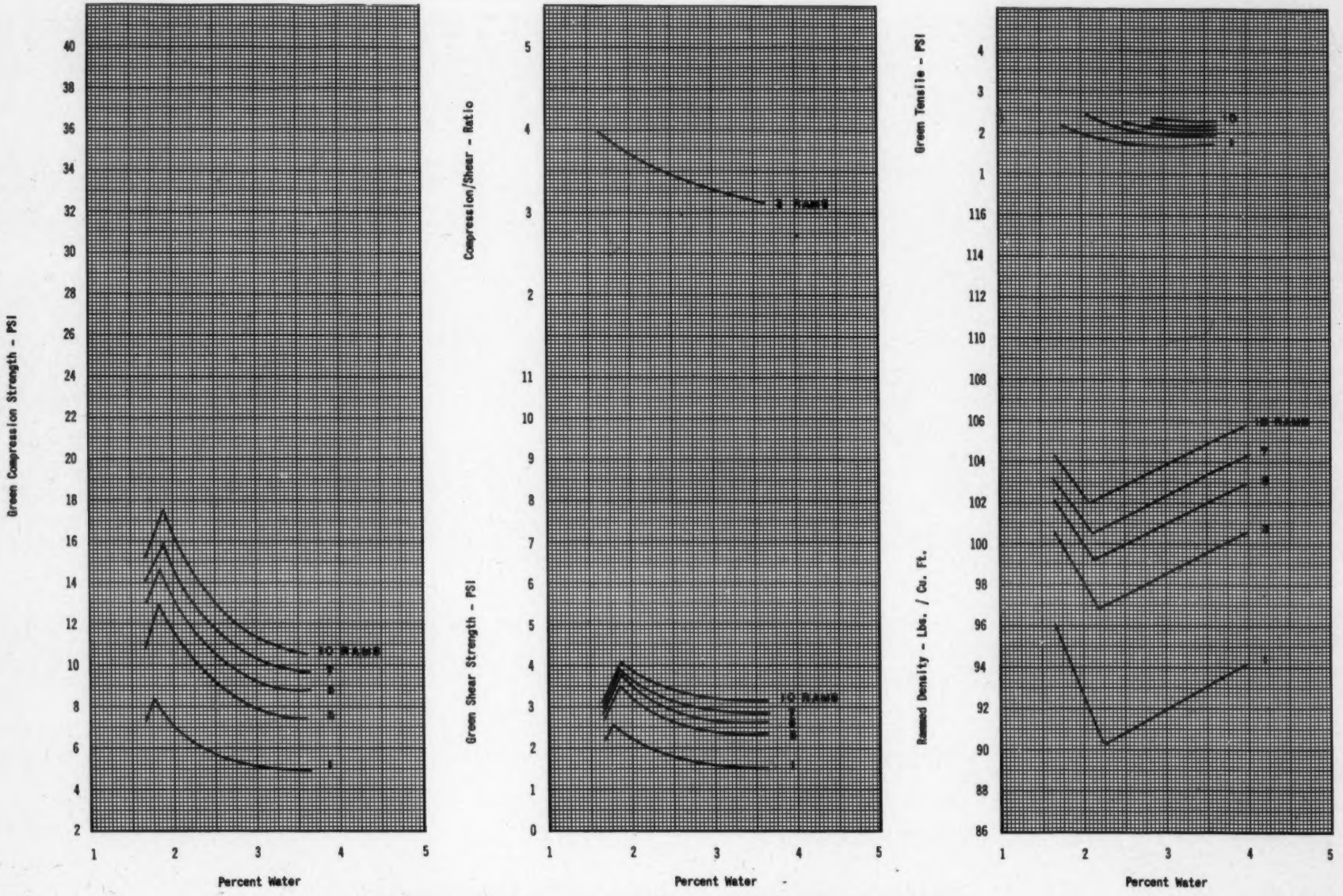



Fig. 1a — Water content and ramming energy effect on physical properties of 4.75 per cent western bentonite bonded Portage silica sand. Mulled 6 min.

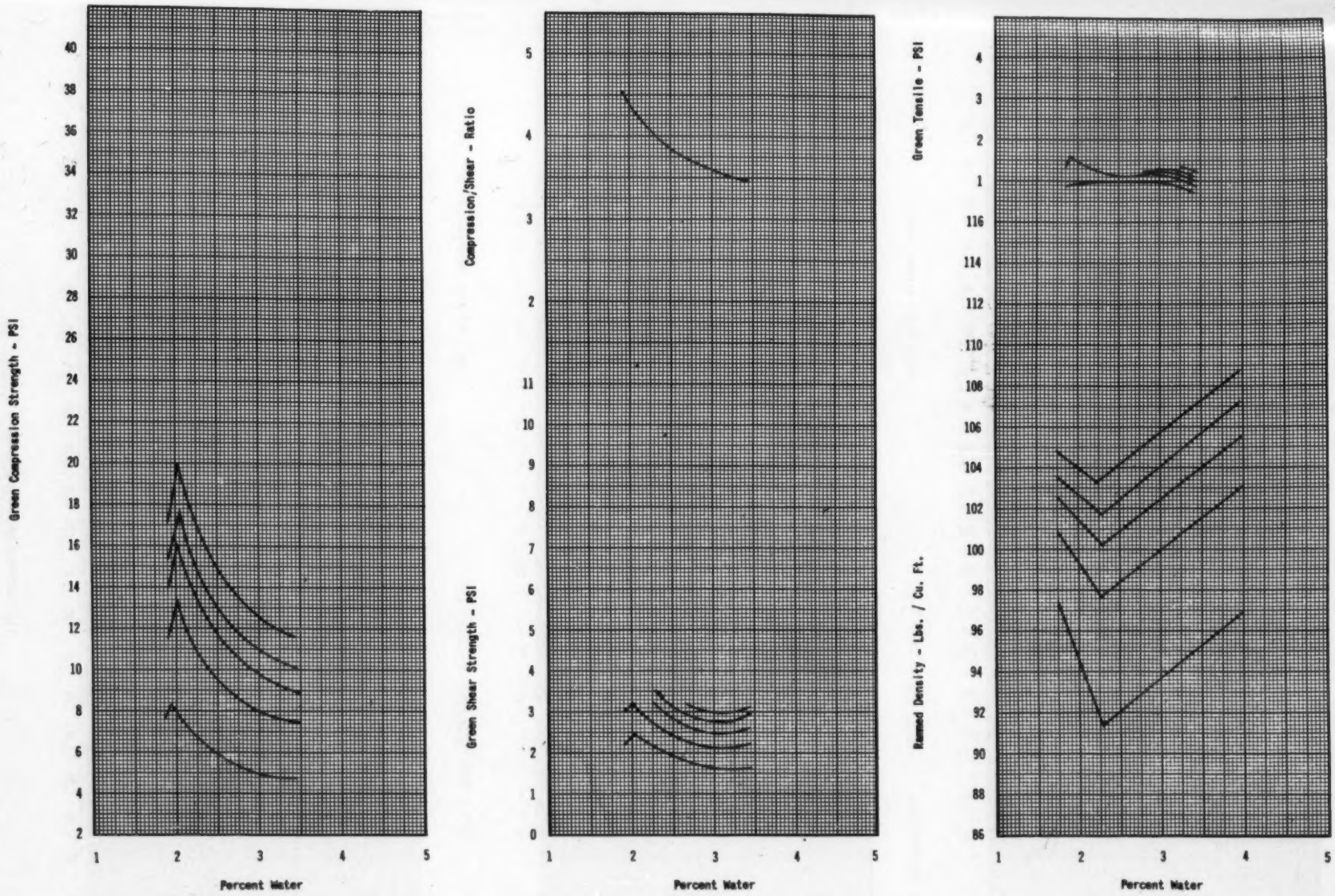



Fig. 1b — Water content and ramming energy effect on physical properties of 4.75 per cent western bentonite, 0.75 per cent wood flour bonded Portage silica sand. Mulled 6 min.

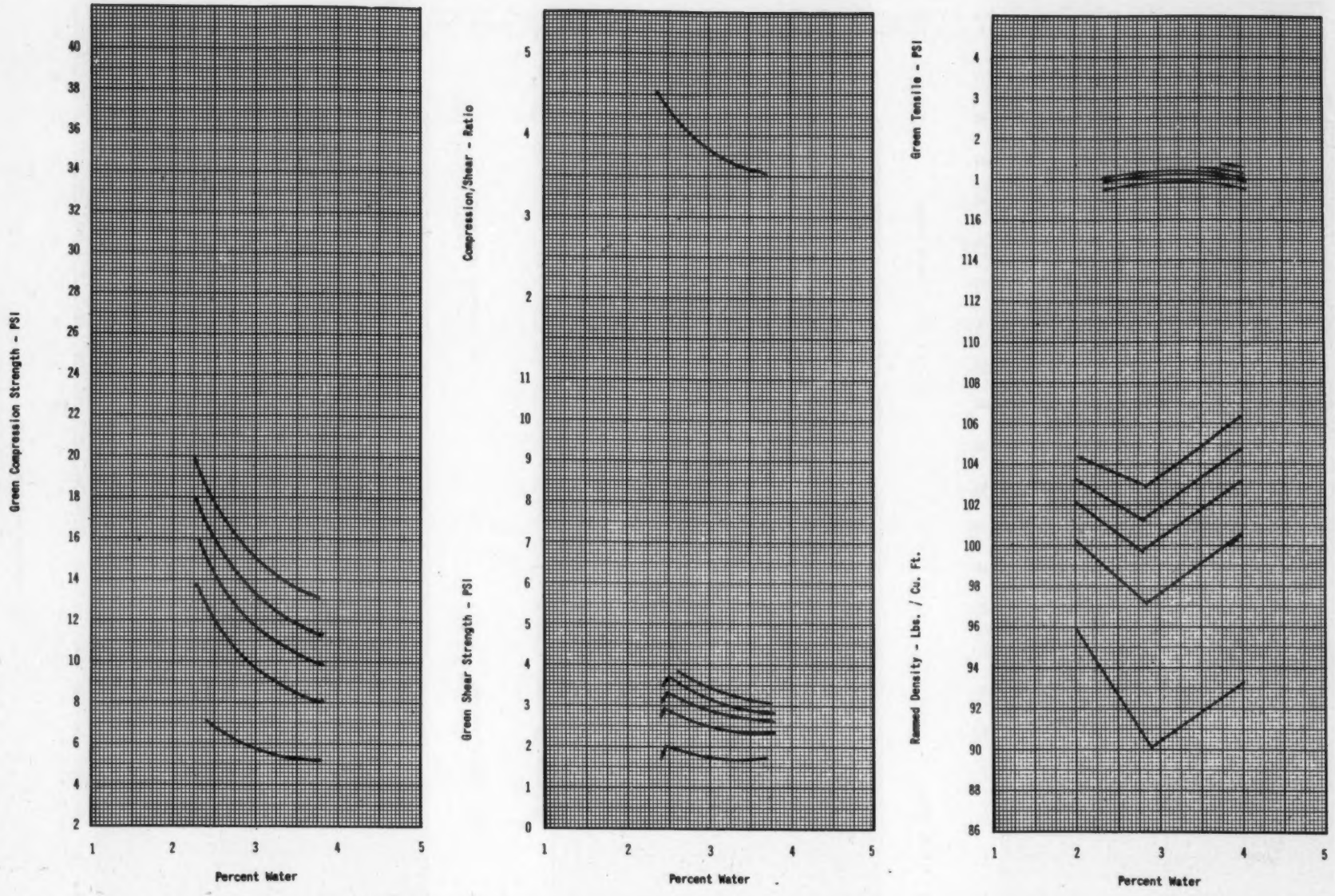



Fig. 1c — Water content and ramming energy effect on physical properties of 4.75 per cent western bentonite, 1.5 per cent wood flour bonded Portage silica sand. Mulled 6 min.

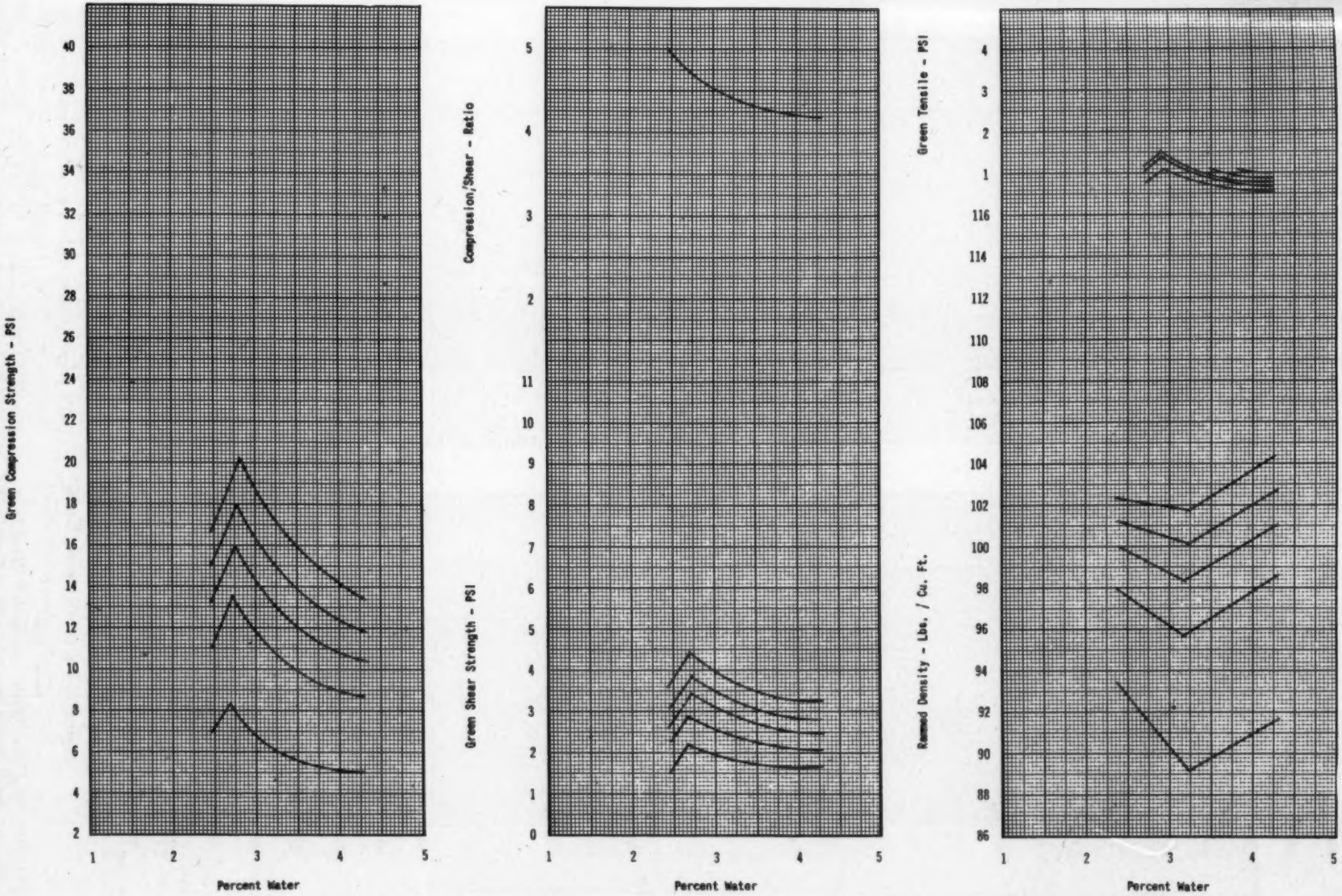



Fig. 1d — Water content and ramming energy effect on physical properties of 4.75 per cent western bentonite, 3 per cent wood flour bonded Portage silica sand. Mulled 6 min.

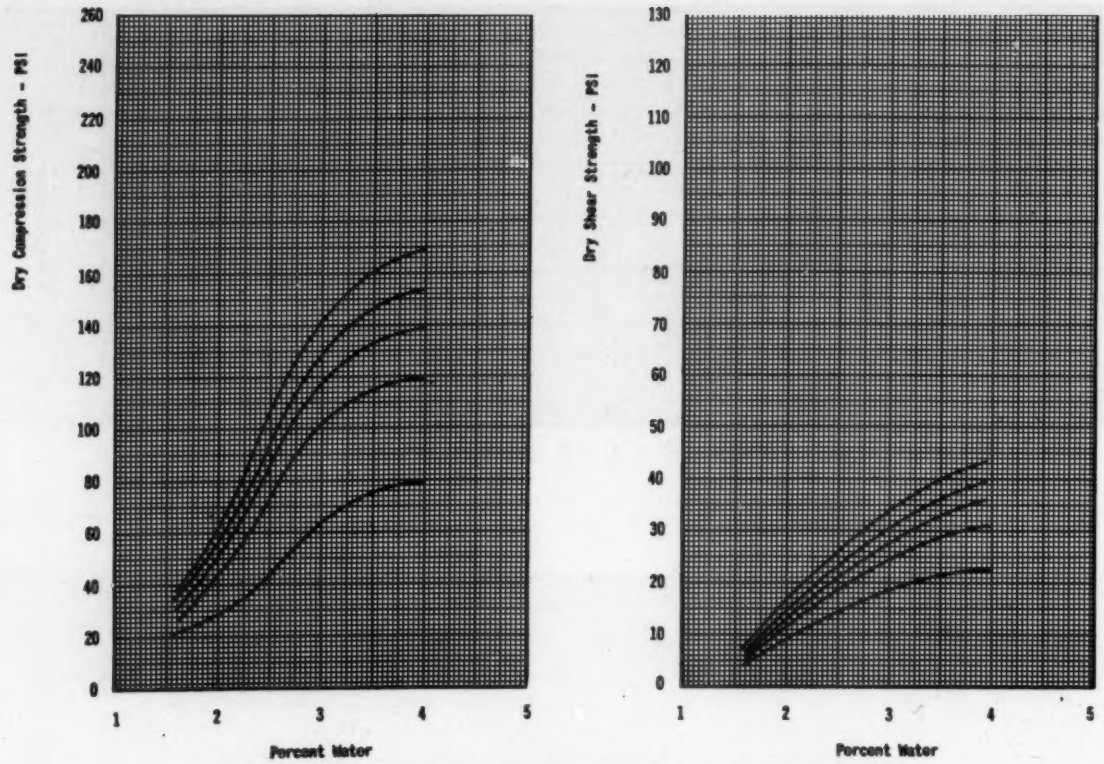



Fig. 2a — Water content and ramming energy effect on physical properties of 4.75 per cent western bentonite bonded Portage silica sand. Mulled 6 min.

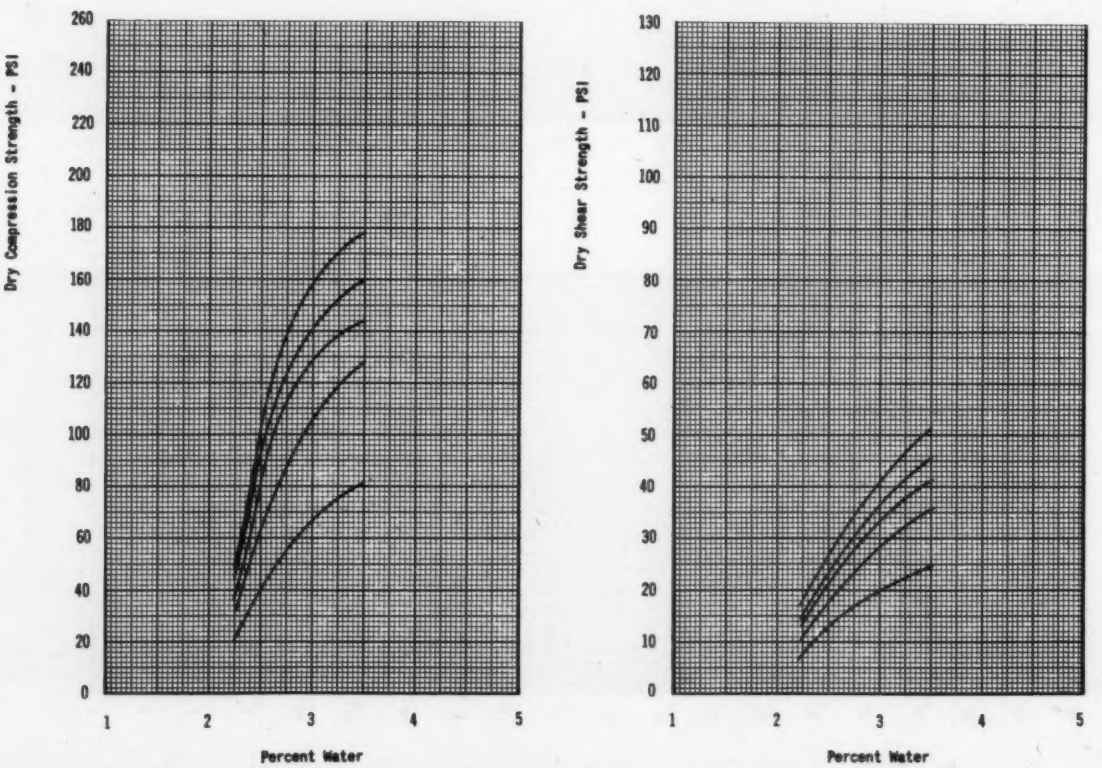



Fig. 2b — Water content and ramming energy effect on physical properties of 4.75 per cent western bentonite, 0.75 per cent wood flour bonded Portage silica sand. Mulled 6 min.

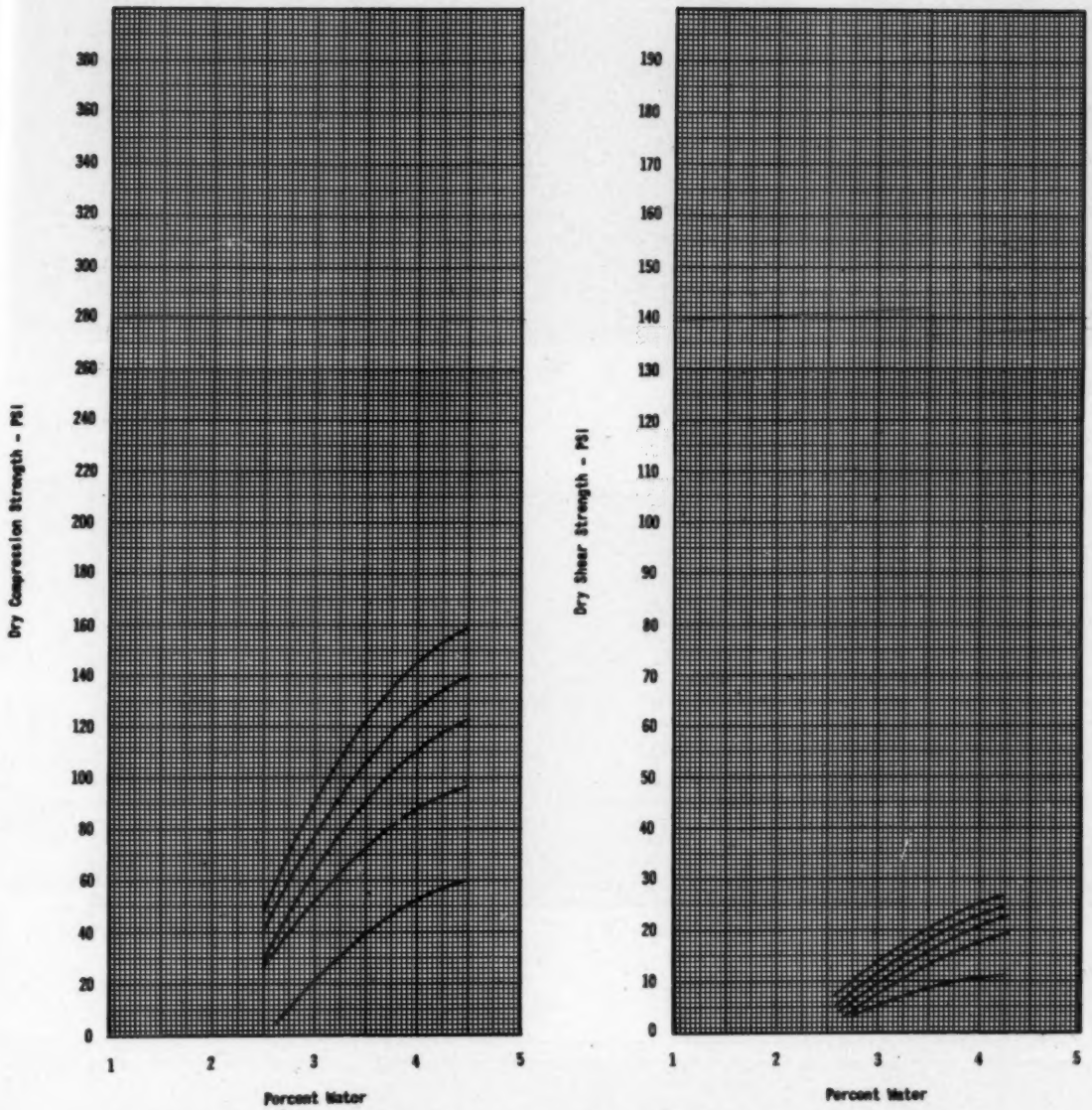



Fig 2c — Water content and ramming energy effect on physical properties of 4.75 per cent western bentonite, 3 per cent wood flour bonded Portage silica sand. Mulled 6 min.

for dry properties. Of course, if water and clay content are held constant while wood flour additions are increased the effect is one of drying the sand, and thereby increasing green and reducing dry compression.

It would seem that a similar reaction can be had by simply reducing water content. This again is oversimplification. If green and dry compression in themselves were capable of describing sand quality, simple reduction in water content would have the same effect as holding water constant and increasing wood flour additions, but such is not the case. While the relationship of green compression to dry compression for western bentonite bonded sands is almost unaffected by wood flour additions, dry shear and the relationship of green compression to green shear are changed substantially.

Comparison of data depicted in Figs. 2a-2b and 3a-3b shows that while dry compression values are simply shifted to points of higher water content, dry shear values are not only shifted to higher water content but actually lowered with increasing wood flour additions.

### *Wood Flour Effect on Southern Bentonite Bonded Sands*

Southern bentonite bonded sands are more severely affected by wood flour additions. While the dry com-

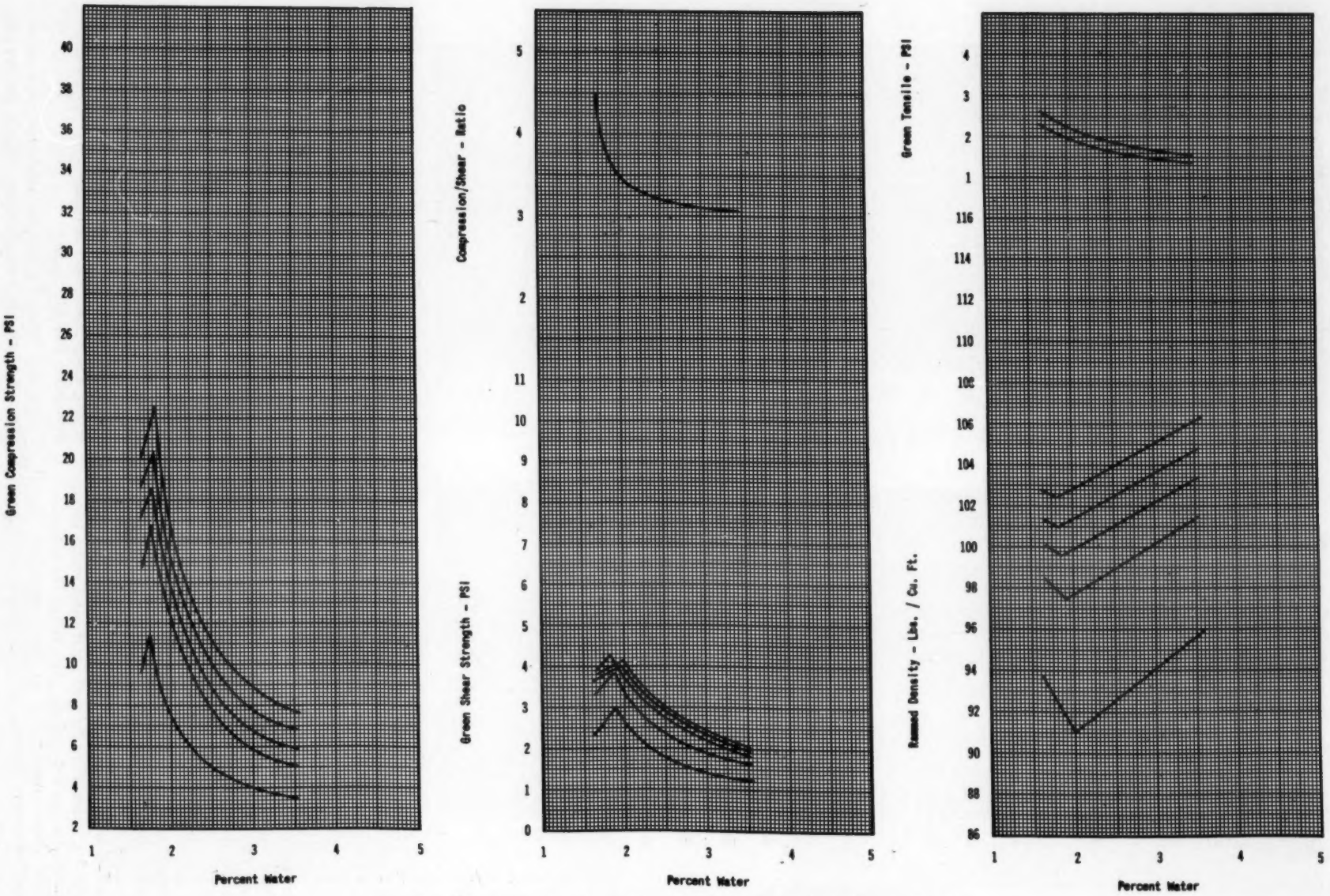



Fig. 3a — Water content and ramming energy effect on physical properties of 4.75 per cent southern bentonite bonded Portage silica sand. Mulled 6 min.

4.75 per cent southern bentonite bonded Portage silica sand. Mulled 6 min.

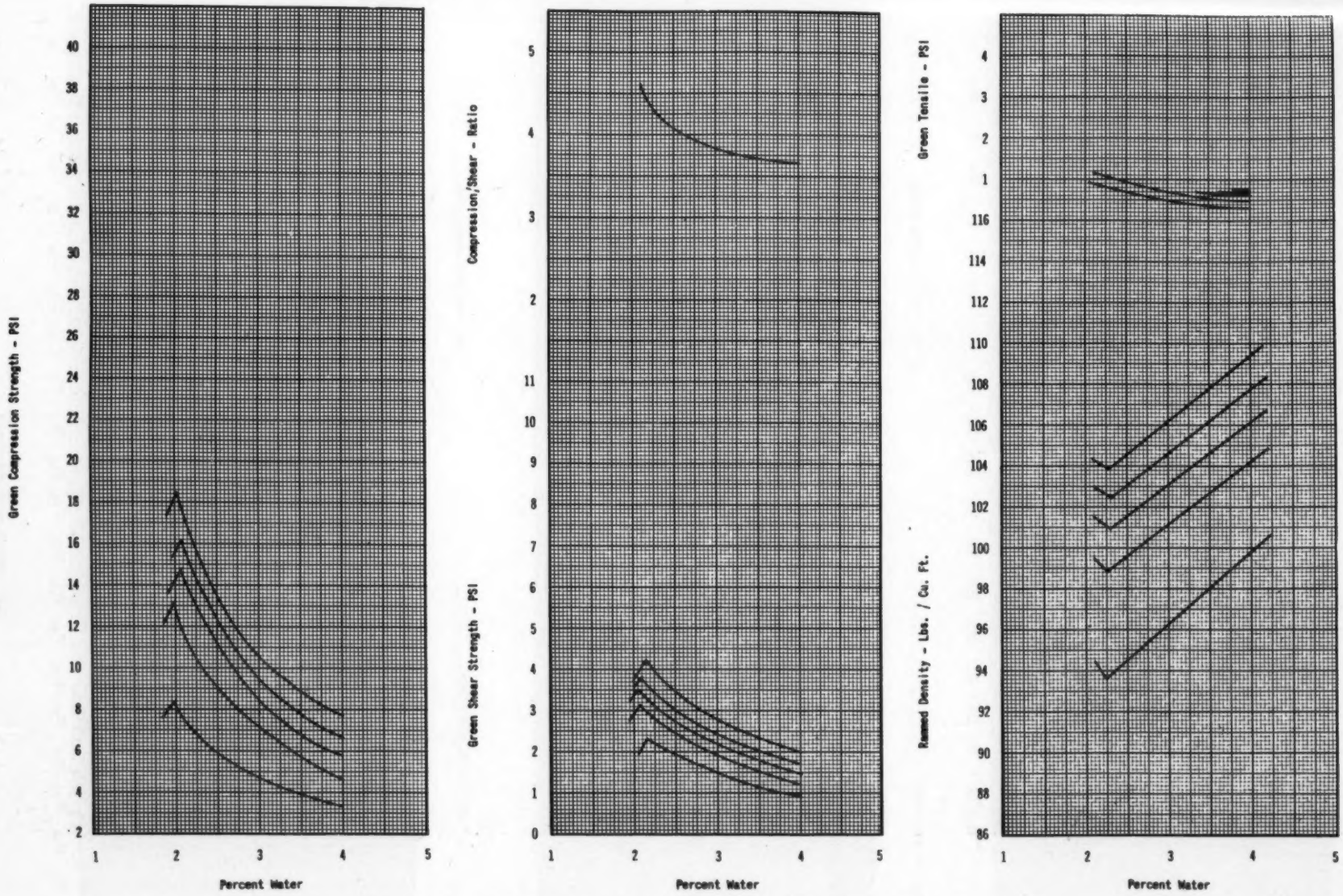



Fig. 3b — Water content and ramming energy effect on physical properties of 4.75 per cent southern bentonite, 0.75 per cent wood flour bonded Portage silica sand. Mulled 6 min.

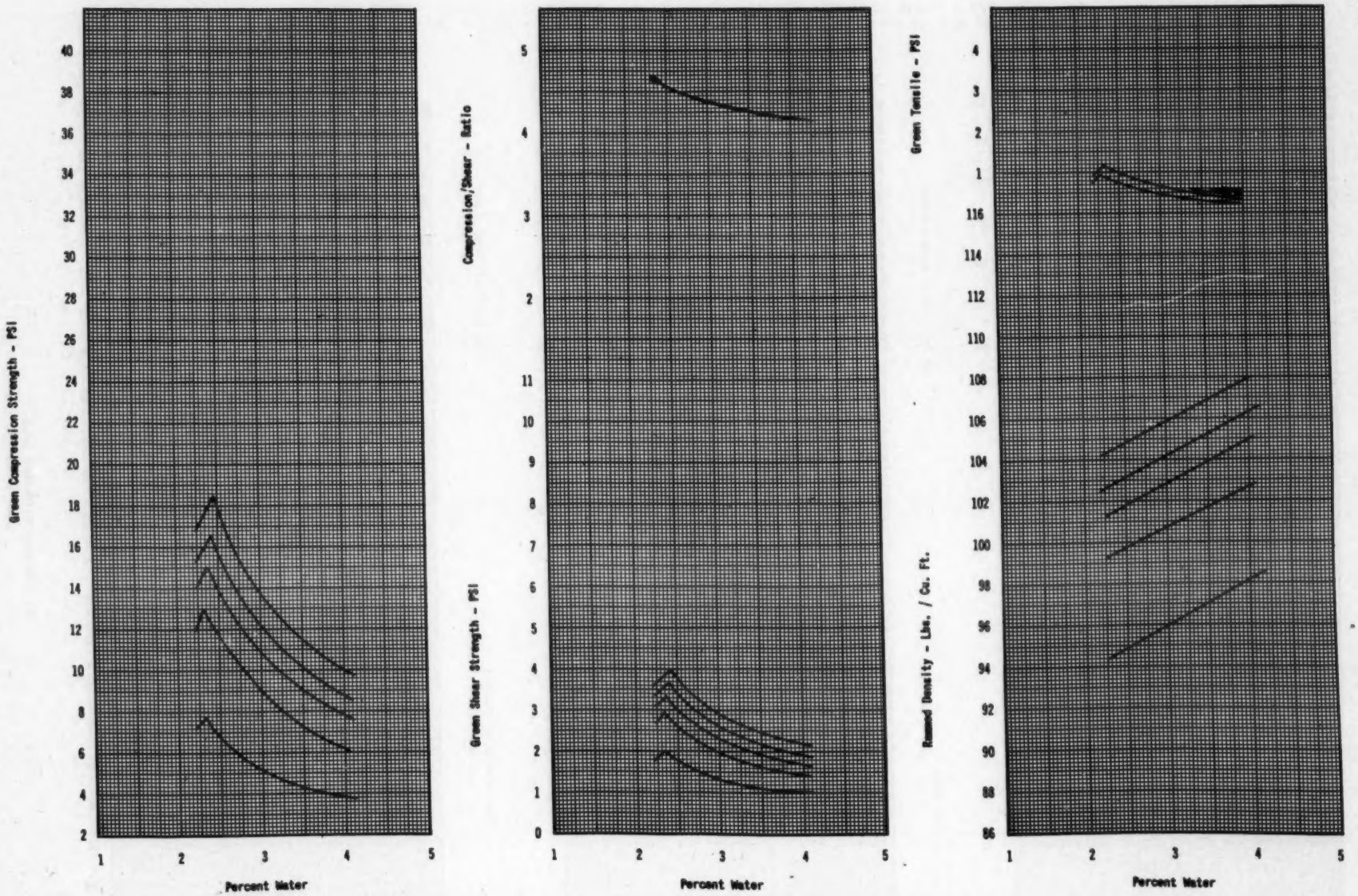



Fig. 3c — Water content and ramming energy effect on physical properties of 4.75 per cent southern bentonite, 1.5 per cent wood flour bonded Portage silica sand. Mulled 6 min.

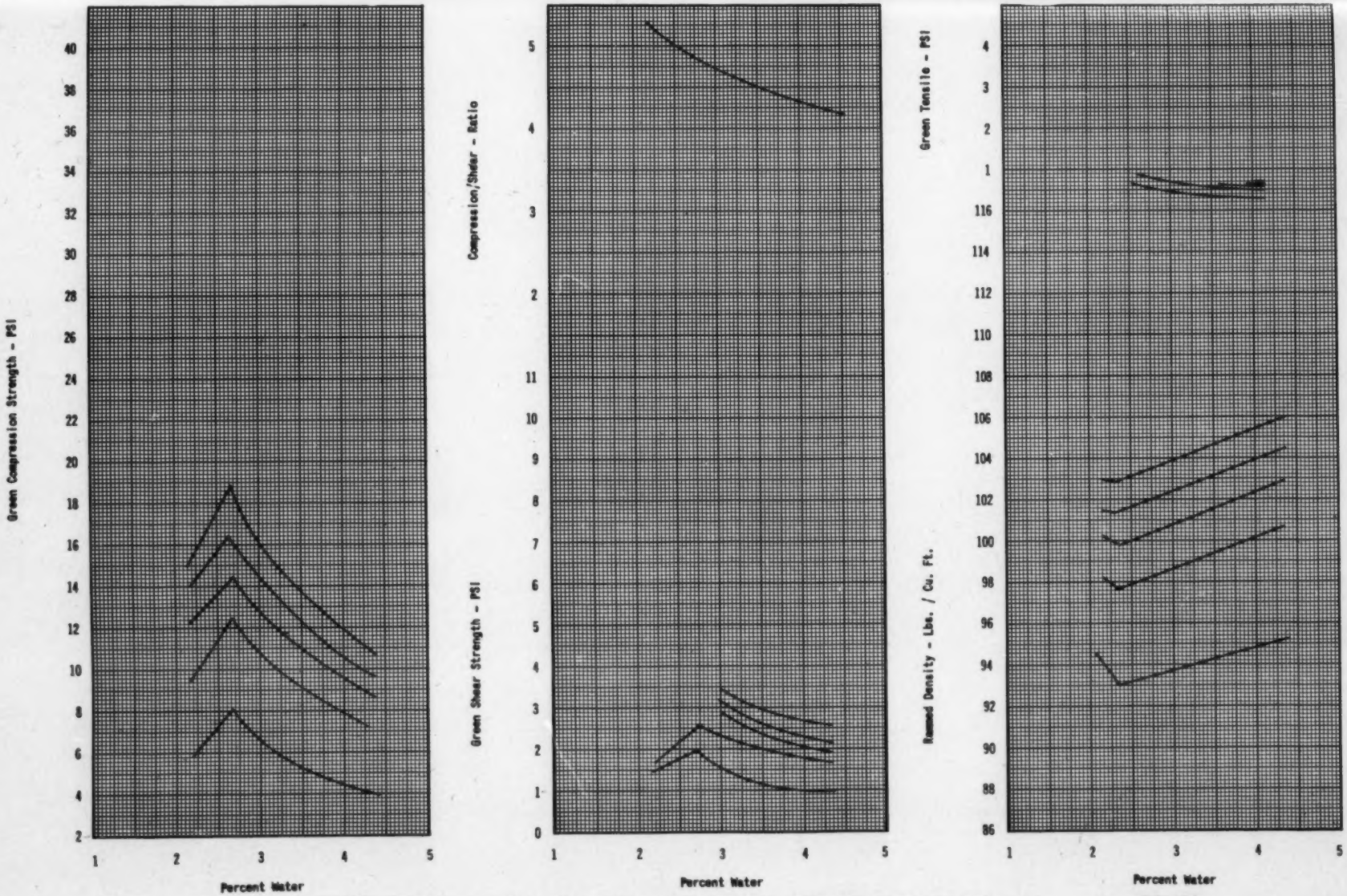



Fig. 3d — Water content and ramming energy effect on physical properties of 4.75 per cent southern bentonite, 3 per cent wood flour bonded Portage silica sand. Mulled 6 min.

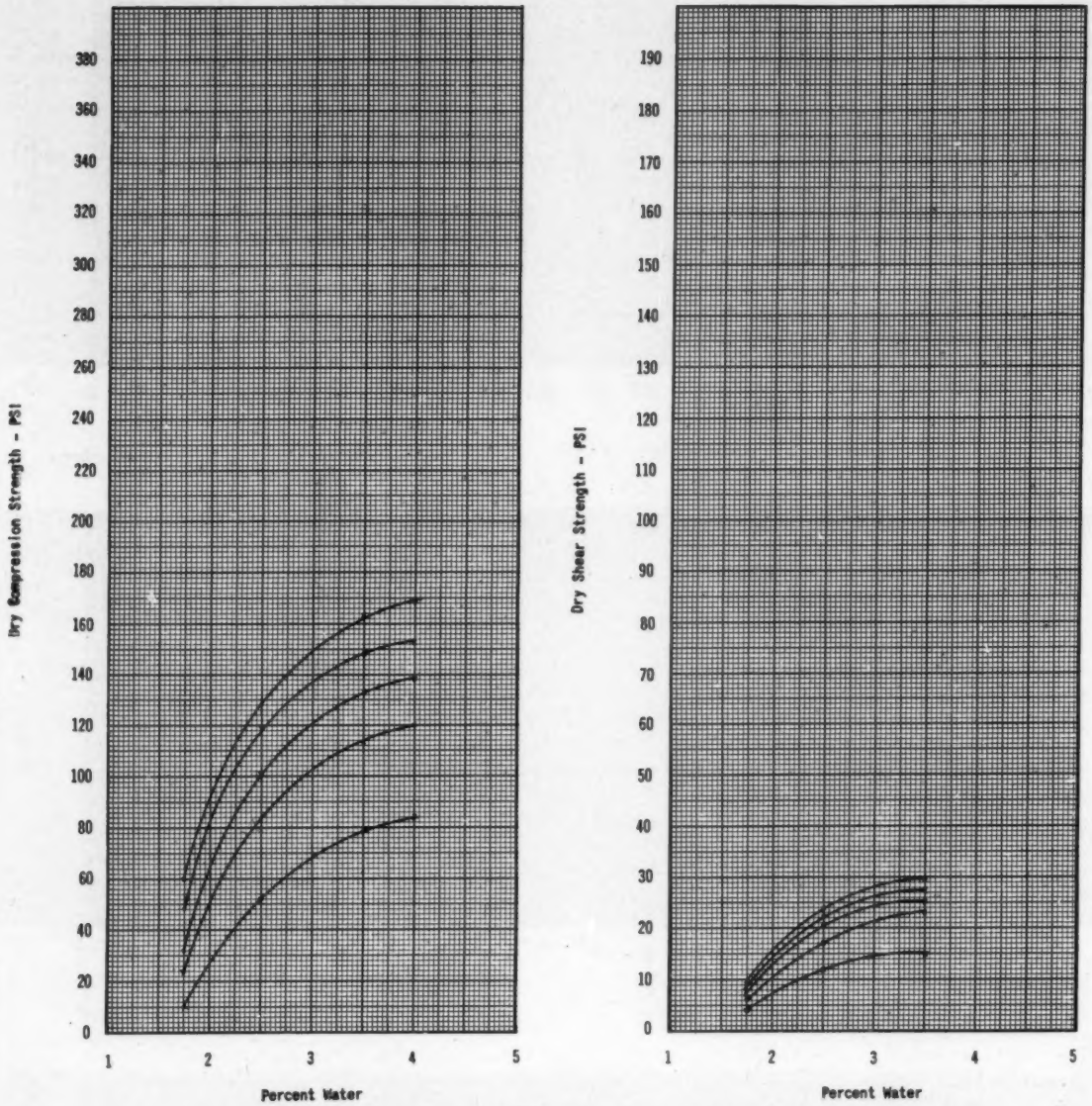



Fig. 4a — Water content and ramming energy effect on physical properties of 4.75 per cent southern bentonite bonded Portage silica sand. Mulled 6 min.

pression strength characteristics of western bentonite bonded sands remain relatively unchanged by wood flour additions up to 3 per cent, both dry compression and shear are drastically lowered by wood flour additions. The effects of wood flour on physical properties of southern bentonite bonded sands are presented in Figs. 3a-3d and 4a-4b.

It will be recalled from earlier progress reports of the series that reference was made to the ratio of green compression to green shear. In the accumulation of data for the series, it has been a significant observation that sands which are wet and tough to the feel, and which would be described by most foundrymen as lacking flowability, have a compression to shear ratio that is near 3:1. It has been observed that this 3:1 ratio can be obtained with any clay-sand mixture through use of high water contents.

Brittle sand can be had from any clay-sand mixture simply by reducing the water content. As water content is reduced, the ratio is raised to between 4:1 and 6:1. One of the main difficulties encountered with simple bentonite-sand mixtures is that the shift from brittle to wet and plastic takes place over a narrow range of water. The water control required

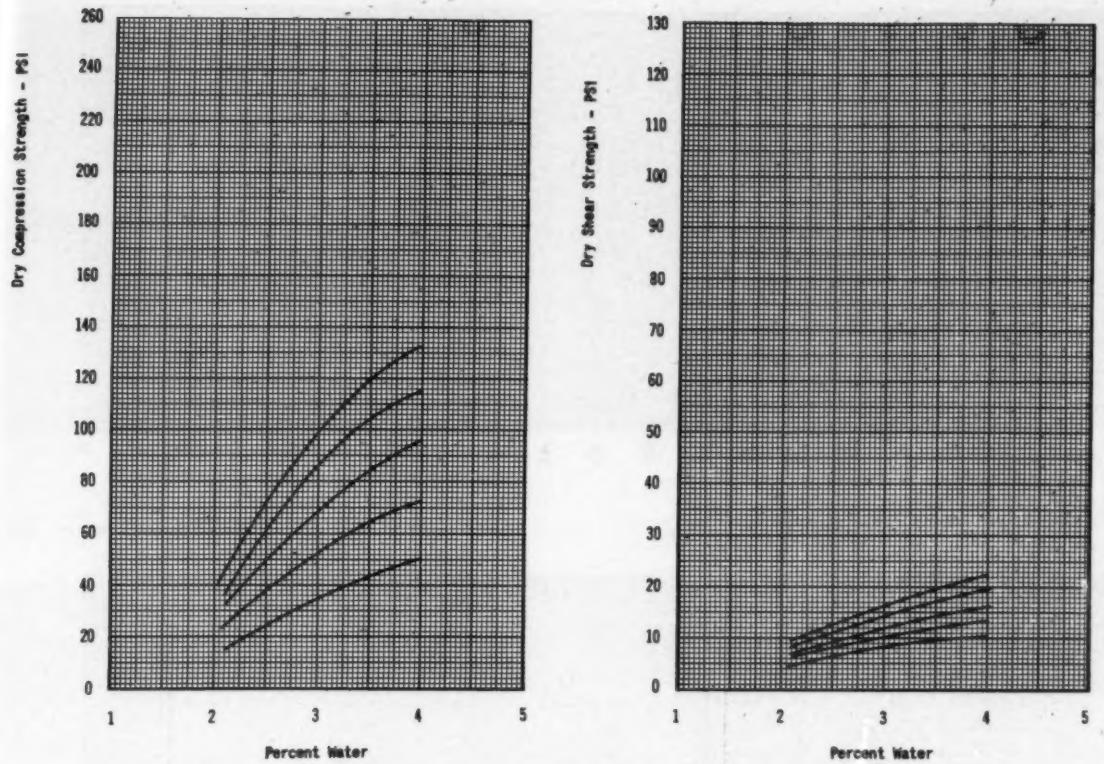



Fig. 4b — Water content and ramming energy effect on physical properties of 4.75 per cent southern bentonite, 0.75 per cent wood flour bonded Portage silica sand. Mulled 6 min.

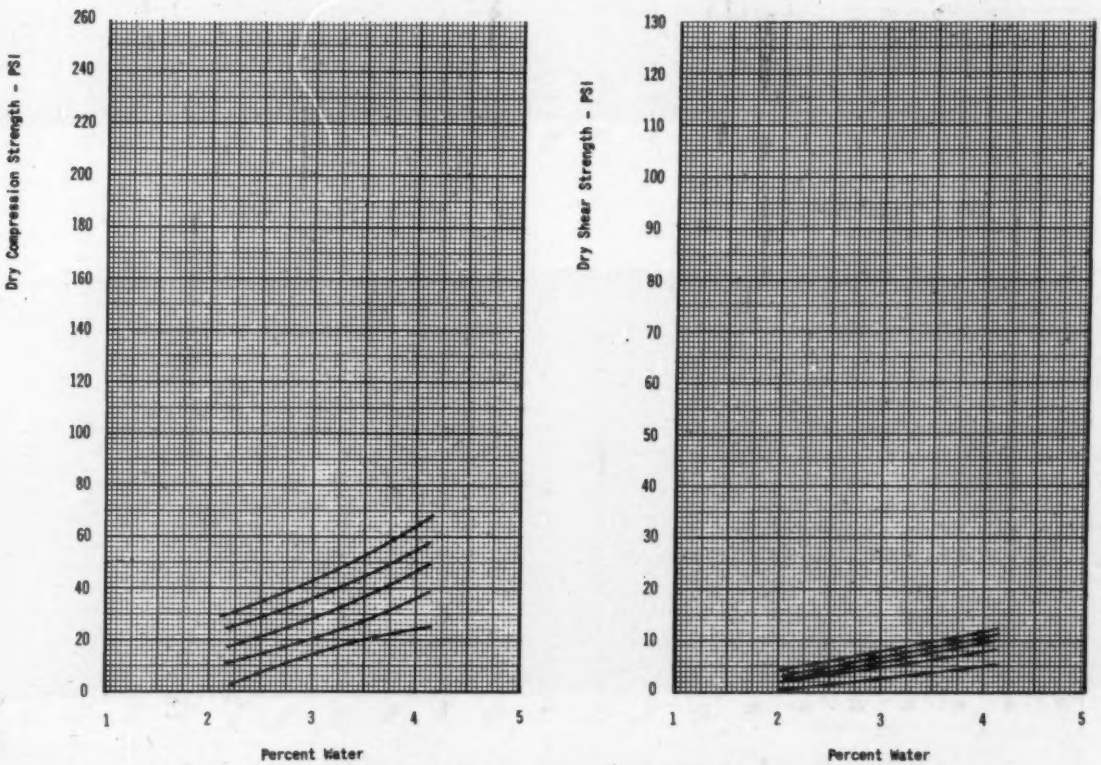



Fig. 4c — Water content and ramming energy effect on physical properties of 4.75 per cent southern bentonite, 3 per cent wood flour bonded Portage silica sand. Mulled 6 min.

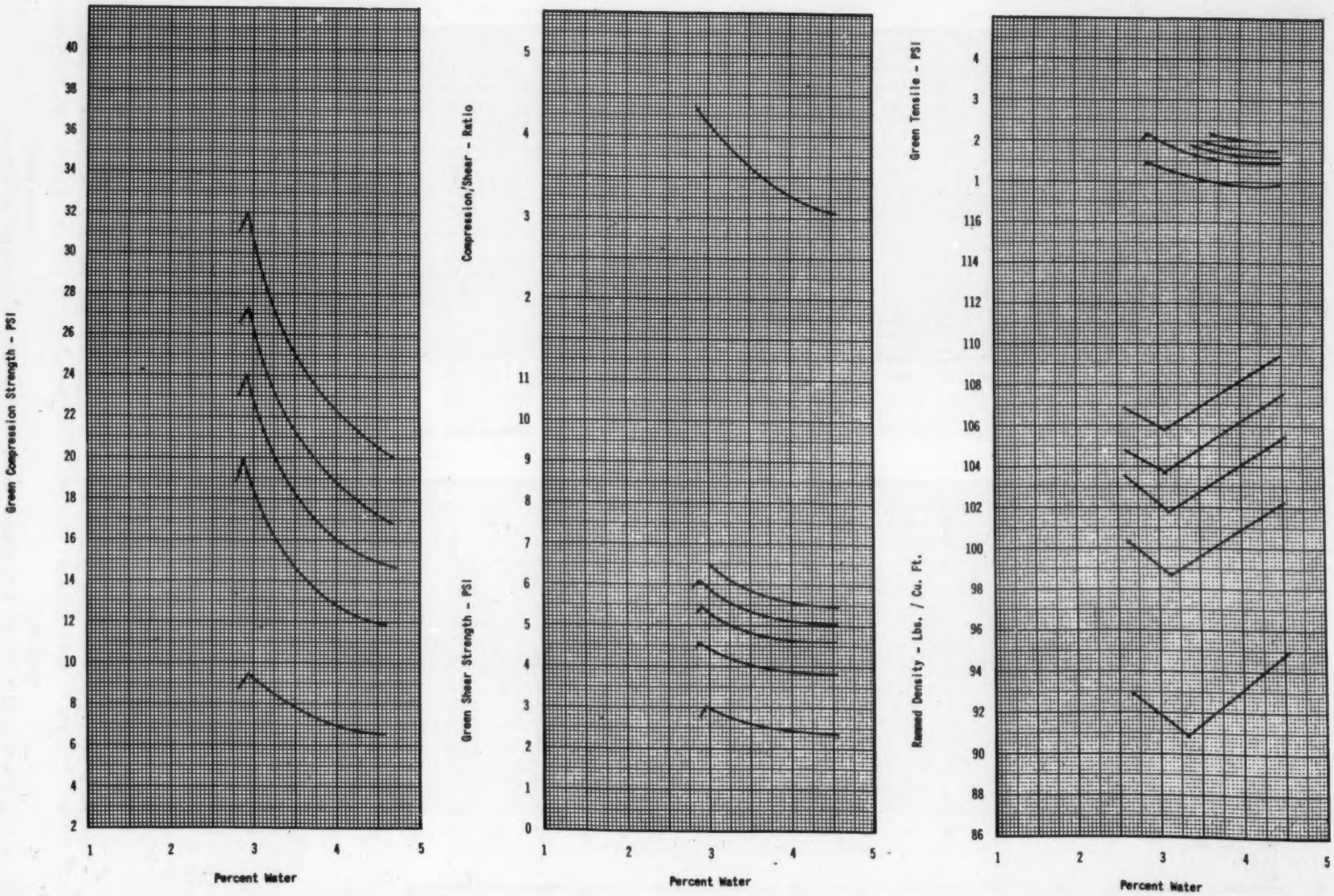



Fig. 5 — Water content and ramming energy effect on physical properties of 7.45 per cent western bentonite, 0.75 per cent wood flour bonded Portage silica sand. Mulled 6 min.

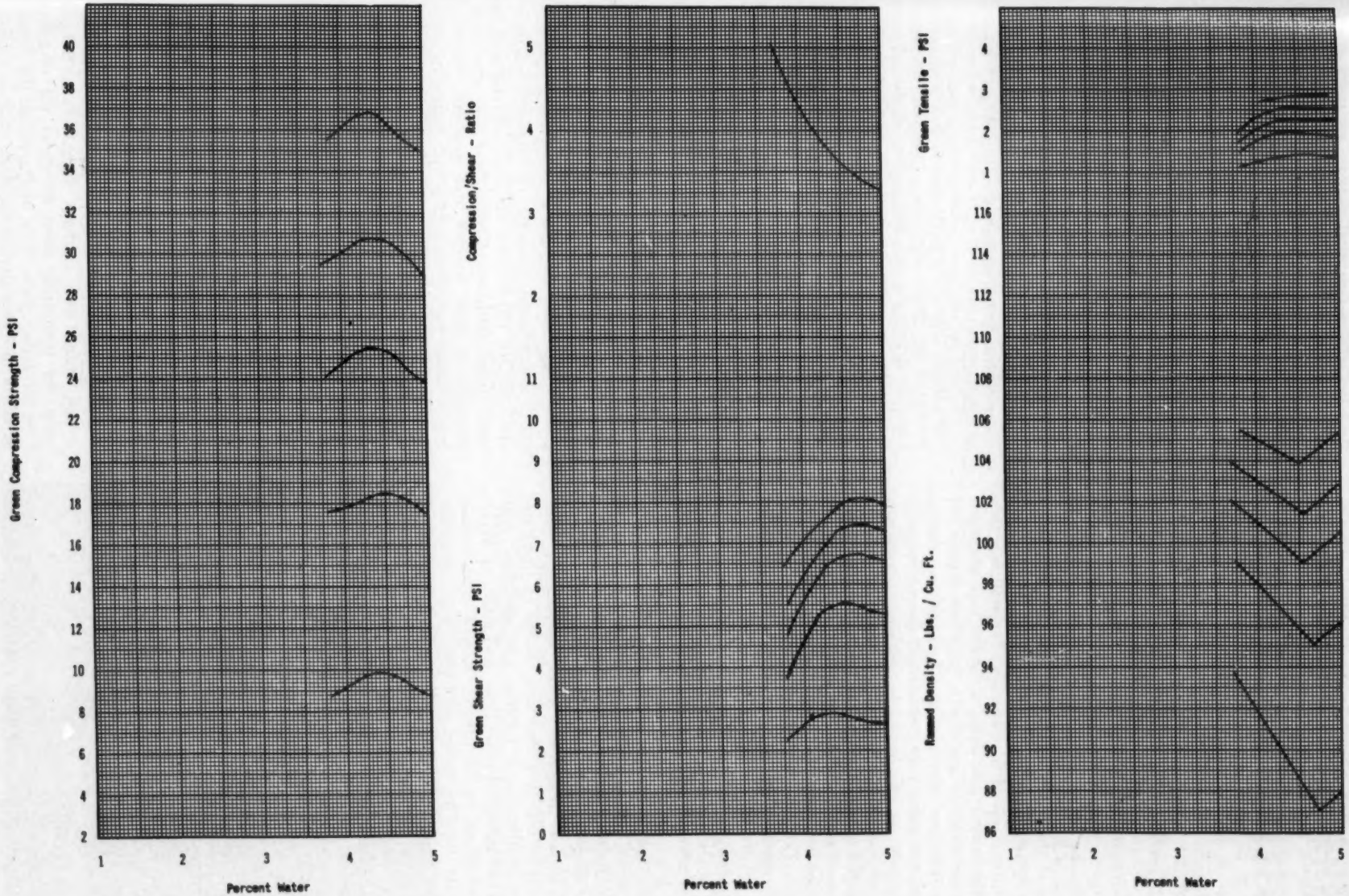



Fig. 6 — Water content and ramming energy effect on physical properties of 10 per cent western bentonite, 3 per cent wood flour bonded Portage silica sand. Mulled 6 min.

to produce a sand that is neither too brittle nor too plastic, while not impossible, is outside the realm of practicality for most foundries.

This is one of the major shortcomings of bentonites. One of the major functions of wood flour in new clay-sand mixtures is to widen this range of water content in which optimum plasticity occurs. These experiments have shown to the satisfaction of the authors that the best molding conditions for sands occur when the ratio of compression to shear is in the range of 3.75:1 to 4.5:1. This is not to say that sands of this nature necessarily produce the best castings, but that they do produce high density molds with a combination of the least amount of effort and least amount of breakage.

### *Wood Flour Function*

As an example of the function of wood flour, consider the system containing 4.75 per cent southern bentonite. Examination of Fig. 3a, which depicts the simple 4.75 per cent southern bentonite-sand-water systems, shows that a control of water content between 1.7 and 1.85 per cent would be required to maintain sands with a compression to shear ratio between 3.75:1 and 4.5:1. Through the addition of 0.75 per cent wood flour, the 3.75:1 to 4.5:1 range can be maintained with water contents between 1.9 and 3.25 per cent. With a 1.5 per cent wood flour addition, the allowable water range is stretched from 2.6 to about 5.0 per cent.

Increase in wood flour to 3 per cent simply produced sands which are brittle (ratio of compression to shear above 4.5:1) over the major portion of the water range. Water contents in excess of 3.5 per cent are required to obtain sands which are not excessively brittle. Wood flour has a similar though not as drastic effect on western bentonite bonded sands, as can be seen in Figs. 1a-1d.

Fireclay-sand mixtures without wood flour additions experience only moderate change in toughness with increasing water content and resemble quite closely the bentonite-wood flour-sand mixtures. Wood flour additions to the fireclay bonded sands simply move the working range to higher water contents.

While wood flour has a potent effect on bentonites at the 4.75 per cent level, its effect at the higher levels of 7.45 and 10 per cent is small. Addition of 0.75 per cent wood flour to sand mixtures containing 7.45 and 10 per cent western or southern bentonite mulled 6 min did not change either feel or the mechanical properties to any perceptible degree. This is shown in Fig. 5. At the level of 1.5 per cent wood flour, sands which were bonded with 7.45 per cent bentonite were affected moderately, but those bonded with 10 per cent were only slightly affected.

Additions of 3 per cent wood flour were required before the plasticity of 10 per cent bentonite bonded sands was highly affected. In general, southern bentonite bonded sands were modified to a greater degree by wood flour additions than were western bentonite bonded sands.

### *Mulling Time Effects*

Although this project included an extensive study of muller effects on wood flour containing sands, the data at 2 and 4 min mulling time are quite erratic and do not lend themselves to intelligent discussion. Even at 6 min mulling time, data collected at the higher bentonite levels of 7.45 and 10 per cent together with 1.5 and 3.0 per cent wood flour were erratic. From the shape of the curves, as shown in Fig. 6, these data indicate a condition of undermulling.

This is similar to the condition expected from a simple western bentonite bonded sand mulled only 2 min. In no instance, regardless of clay type or per cent and wood flour per cent, was 2 min mulling adequate to break up localized clusters of wet wood flour. In many instances wet wood flour clusters persisted to 4 min mulling and beyond. As a result, data collected throughout the project were less predictable than the data collected on simple clay-sand-water systems and presented in earlier progress reports of this series.

In the earlier reports on clay-sand systems, data presented are exact, and the graphical representations are drawn through actual plots of data. In the present report, graphical representations are the best lines which could be drawn through the plotted data. The data of this report would best be represented by bands 5 per cent from the lines drawn, but is presented in the manner shown to avoid confusion.

## CONCLUSIONS

The reader is reminded that the data presented were gathered on new sand mixtures.

The addition of wood flour to clay-sand-water systems does not in reality lower dry compression strength of western bentonite bonded sands. It simply increases the water content required to obtain a given dry compression level. In a similar manner, green compression strength is not increased perceptibly. Only the water content required to obtain a given green compression strength is increased.

Dry compression strengths of southern bentonite sands, on the other hand, are drastically reduced by wood flour additions. Green shear and dry shear strengths of both western and southern bentonite bonded sands experience a drop with increasing wood flour additions. Low clay content sands are affected more by wood flour additions than high clay sands.

While simple bentonite-sand-water systems experience drastic change from brittle to plastic with small change in water, the same combinations with wood flour added experience the change from brittle to plastic over a substantially wider range of water content. Simple fireclay bonded sands exhibit this property, and additions of wood flour did not influence this water range except to shift it to higher water levels.

Difficulty in eliminating wet wood flour clusters through mulling was increased by wood flour additions, and, in general, presence of wood flour retarded mulling action.

## ACKNOWLEDGMENT

The writers wish to thank Magnet Cove Barium Corp. for permitting them to spend the necessary time to gather these data and publish the results based on interpretation of the data.

# MACHINED PATTERNS CAN BE COMPETITIVE

by Ray Olson


## ABSTRACT

**The all machined metal pattern can be competitive to patterns made from some of the newer materials, according to the author. It is up to the pattern shop to adopt time saving methods so these patterns may be produced in a faster, less expensive and more accurate fashion. The resulting end product will be one which will outlast some of the patterns made in newer products.**


## INTRODUCTION

Revolutions in the foundry industry have been occurring at a steadily increasing rate during the past generation. This is particularly true in the field of patternmaking. It is said that some good comes from even the worst of revolutions. Many of those which have exploded on the pattern scene have offered some benefits.

Newly developed synthetic materials to replace wood and metal have elbowed into a firm place in the pattern picture. Precision casting techniques to produce metal patterns to close tolerances in the as-cast condition have appeared on the scene. More recently emphasis has been placed on new and old processes for depositing metal to build up or actually create a metal pattern. All of these are considered revolutionary in nature.

Everyone is interested in this type of development. This includes, of course, those who buy patterns. Too often the foundry is the last place in the manufacturing picture to receive money and attention. In the foundry, patterns are considered a necessary evil. The time necessary, and cost required to produce a good pattern have always been reluctantly tolerated. So when a revolution comes along that will tend to eliminate, in a large sense, the patternmakers, customers hop on the bandwagon. Needless to say, cure-alls do not always work and again patterns are a necessary evil.

## METAL PATTERN EQUIPMENT

Let us give the same energy and attention now directed at new processes to the old way. By old way, machined production metal pattern equipment is meant. For a good solid production pattern that can be readily checked, repaired and maintained, nothing can beat an all machined job. This article does not refer to those patterns which, due to size or production requirements, are more practically made from other materials.

Progress might better be considered using the stand-

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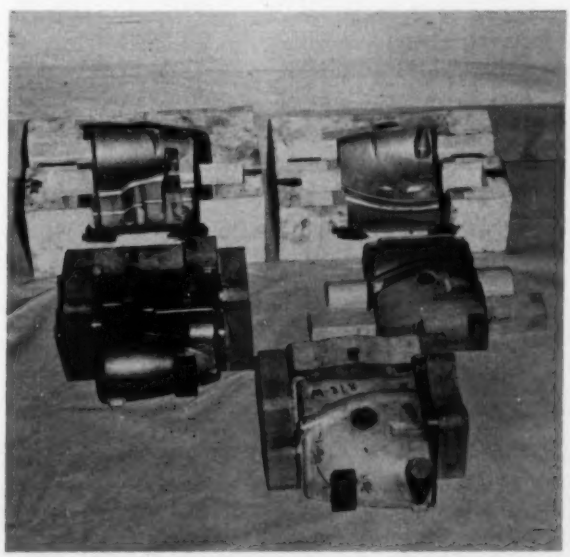


Fig. 1a — A set of pattern and core box duplicating models, with original hardwood core dummy are made from hardwood, plastic and hard plaster.

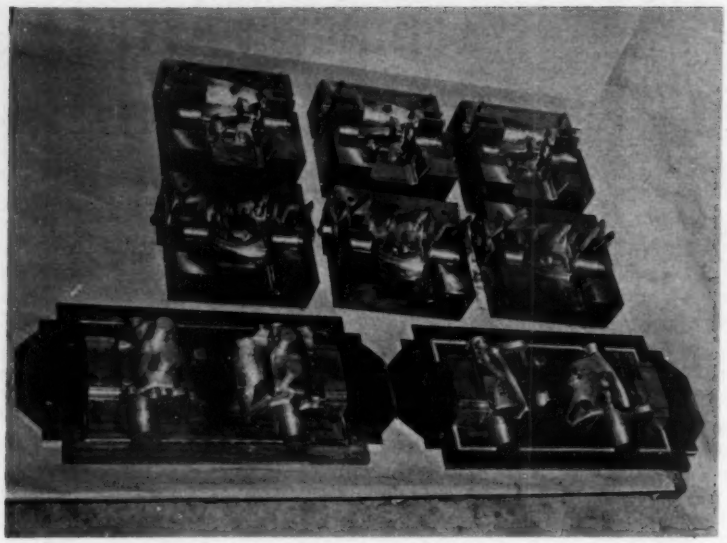


Fig. 1b — A set of three duplicate shell core boxes and two mounted patterns made from the models illustrated.

ard time proved pattern materials processed in a faster, cheaper, more accurate method. Many developments in the art of stock removal or machining have come into widespread usage. In most cases, however, it is not the pattern shops that take advantage of those developments. They are too busy trying to learn all about these revolutionary developments to improve on established techniques.

The major cost and time requirements of producing all machined metal patterns can be drastically reduced. First, let us look at how this job has been done. Depending upon the degree of complexity, many trips have been required between layout, machine and bench for such a pattern. By pattern, any portion or type of pattern equipment is meant. Pattern or core box castings would go through rough layout. Then some flat surfaces and heights would be machined. Back then to layout to prepare for secondary machining operations.

In some cases bench operations, such as template fitting, could occur before secondary machining. After secondary machining, such as vertical surfaces and other operations determined by that second trip to the Layout department, this equipment would make a third trip to layout and subsequently to the machines. The foregoing, in a general sense, typifies what has been considered standard pattern machining practice. In complicated jobs this step-by-step operation sequence can be repeated at least six times before the equipment is ready for final bench hand work.

## PATTERN PRODUCTION COSTS

That has been the problem. In the operations just described lie most of the cost of producing this type of pattern equipment. This is triply painful. The cost is increased to a point where the revolutionary processes are taking over with a substantial loss of business for some of the pattern shops. Second, the time requirements are such that pattern customers, the ever impatient pattern buyers, cannot wait. Last, but certainly not least, is the drain on pattern shop skills. The caliber of patternmakers required–layout men, machinists and bench men, for this method is severe due to the constant mental and manual dexterity involved.

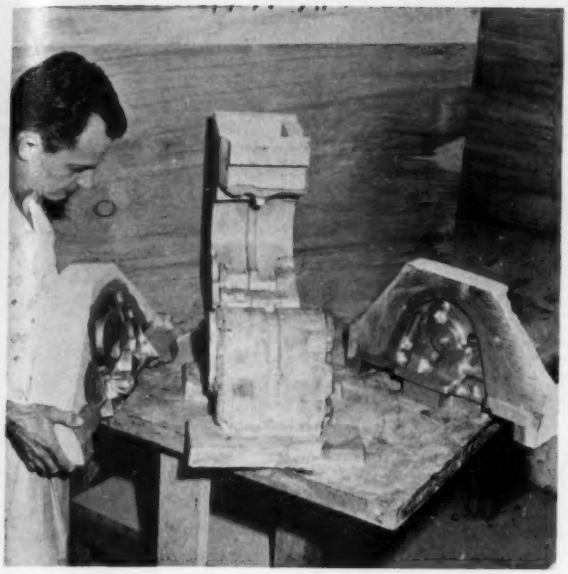
The major factors required to solve this dilemma lie in better pattern engineering and automatic duplicating machines. Machines that will remove metal at a much faster rate than before while reproducing from a model to close tolerances must be used. This rapid metal removal rate is possible because of rigid machine construction and a flood of coolant to the cutting tools. By keeping the cutting tools cool and free from chips they will stay sharp, perhaps ten times as long, as when running at slower speeds on conventional machines without coolant.

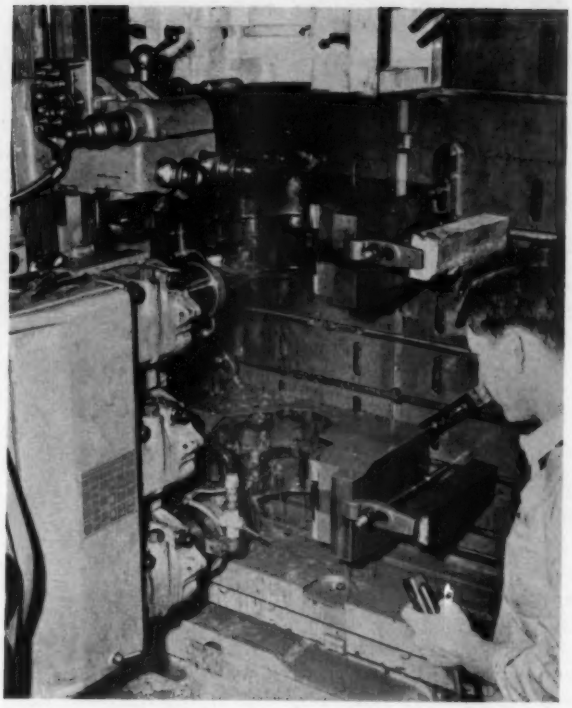
The use of coolant has been more or less restricted in conventional machining because of the necessity to follow layout lines or to watch matching surfaces. To further take advantage of this duplicating equipment, two or more spindles can be used simultaneously. Multiple patterns or core boxes can be cut at one setting of the machine. With some small additional setup time, several pieces can be produced for the price of one.

## MACHINES ALONE ARE NOT THE ANSWER

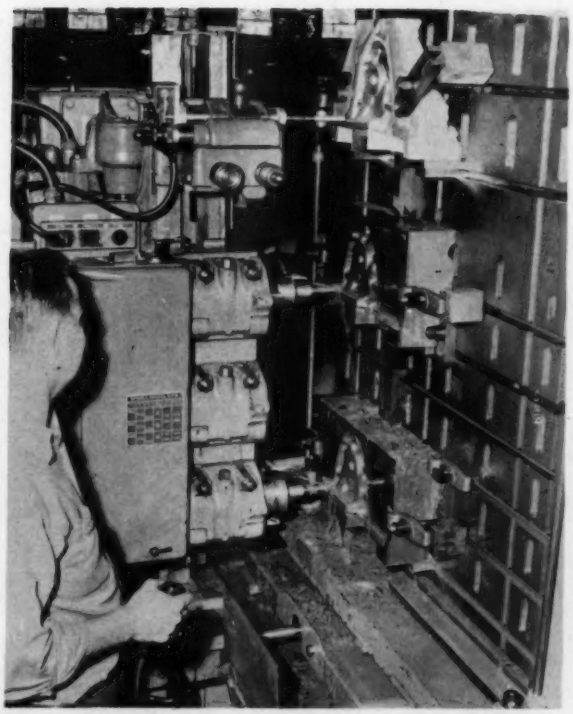
Obviously, this description pertains to milling operations, which constitute the largest volume of machining in most pattern shops. Tracer controlled lathes make repetitive, contour type turning a simple operation. Experience has shown that when four or more pieces are required, a precision template can be made, and the tracing attachment set up with definite savings. In some cases, due to special contours, when only one piece is required, it can be traced.

To make machined patterns competitive through the use of duplicating equipment, the important factors, machine operation and models, must be considered. The first and more obvious concerns the machine operation directly. In comparison with con-









**Figs. 2a, 2b, 2c and 2d — A set of hard plaster duplicating models are made from a plaster master core for a three part core box, and two duplicate core boxes are machined at a time for each section.**

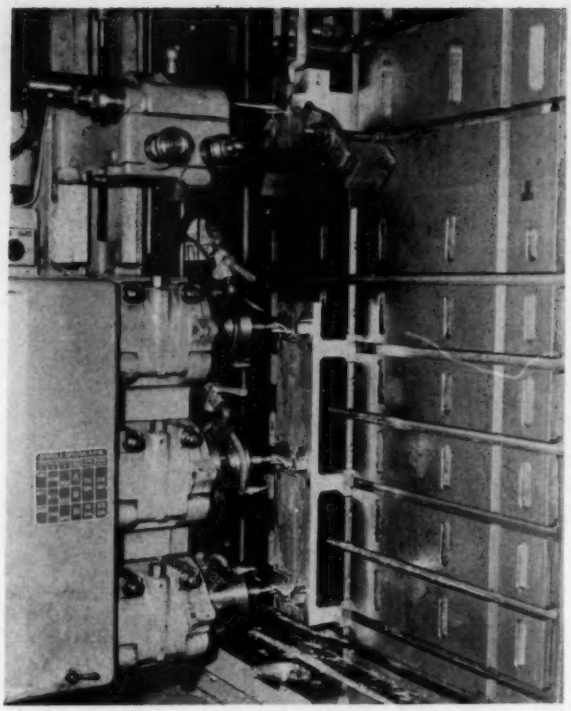


Fig. 3a — One pattern was machined from solid bar stock and test run in the foundry. This pattern (both cope and drag halves) is set up in the top fixture, and three cope and drag patterns are duplicated below.

ventional milling machine operation, cutter size and shape control is an absolute necessity. A minumum of one extra cutter of each specific taper and radius more than the number of duplicating spindles is required. Remember, in multiple spindle duplication the cutters must be identical. Every variation of cutter diameter or radius will be reflected in the machined pattern.

Stylus or tracer points must be made for every different milling cutter used on the duplicator. In most cases four different points are required for each cutter; one for each rough and finish profile, and rough and finish 3-dimensional. These styluses must be cut down or altered, especially those for finishing operations every time milling cutters are sharpened, if known accuracy is to be maintained. Cutters and styluses should be kept in sets and only used for duplicating operations.

It is, of course, fundamental that the duplicator operators be trained in machine potential and operation. They must be supplied with adequate clamps, parallels, keys and other equipment to minimize setup time. It is easy to lose many of the benefits of this type of machine if setup time is exhorbitant.

The second basic problem is more apt to be neglected to a degree in many shops. This is the making of the models or masters for duplication. The finished part cannot be more accurate than its model. Models are usually made from hardwood, metal, plastic and hard nonshrinking casting plaster. The need for working with greater accuracy than ever before

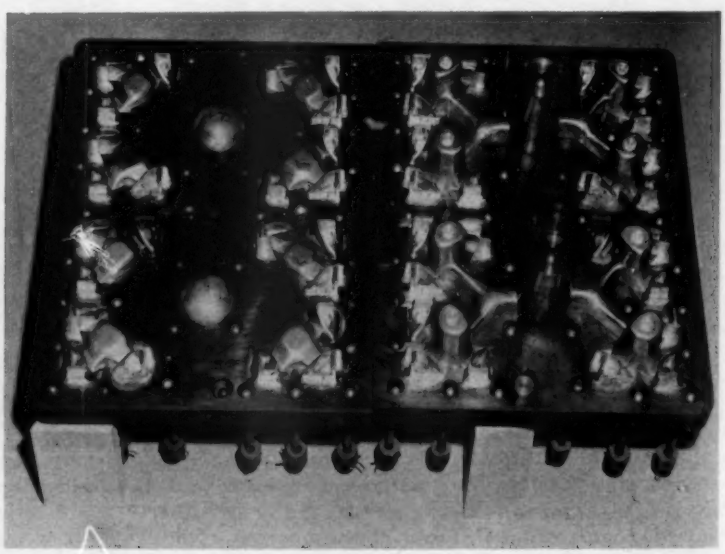


Fig. 3b — Completed pattern plate showing 23 patterns which are duplicated from original pattern, as shown in Fig. 3a.

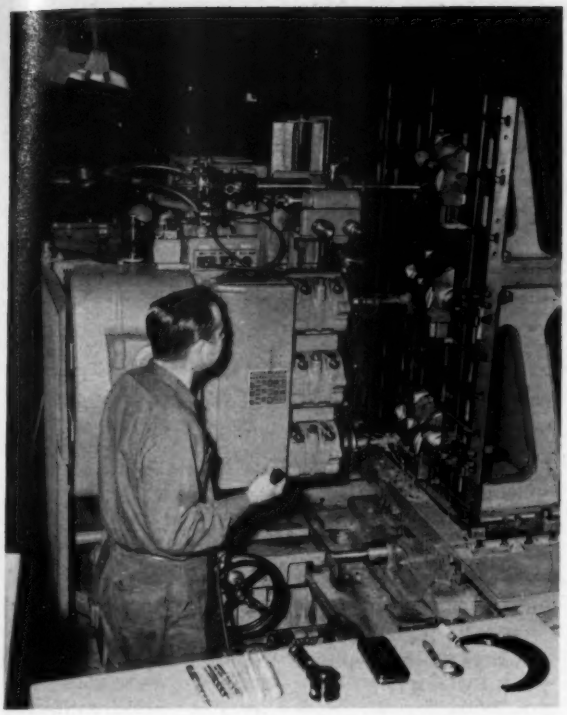


Fig. 4a — Making a finish cut on two special alloy patterns from a combination hardwood and plaster model.

with these materials is immediately apparent. The use of micrometers, height gages and other precision measuring tools becomes more necessary in wood pattern shops. To get the model made right in the shortest possible time is one of the most important factors in a duplicating setup.

## DUPLICATING MASTERS

Before making pattern casting masters, it is sometimes best for the first shop operation to be the making of the duplicating masters. Not all of the dimensional questions come to light in pattern engineering. It is not until the complete object that the engineer is illustrating on the blue print is created that drawing errors or omissions show up. In making the accurate master, at the first step some questions are resolved that otherwise do not come up until the pattern is in the finishing stages of metal work. In case of an engineering error, this sometimes causes costly rework and lost time. Now these questions can be resolved early in the job before much harm can be done.

Duplicating models can be made of one material or a combination of materials as may be practical. For example, assume that both pattern and core box models are to be made for a casting with an irregular contoured wall of uniform metal thickness. All or part of a dummy or core plug would be cut from hardwood to the exact contour of the inside of the casting, allowing only the final casting shrinkage.

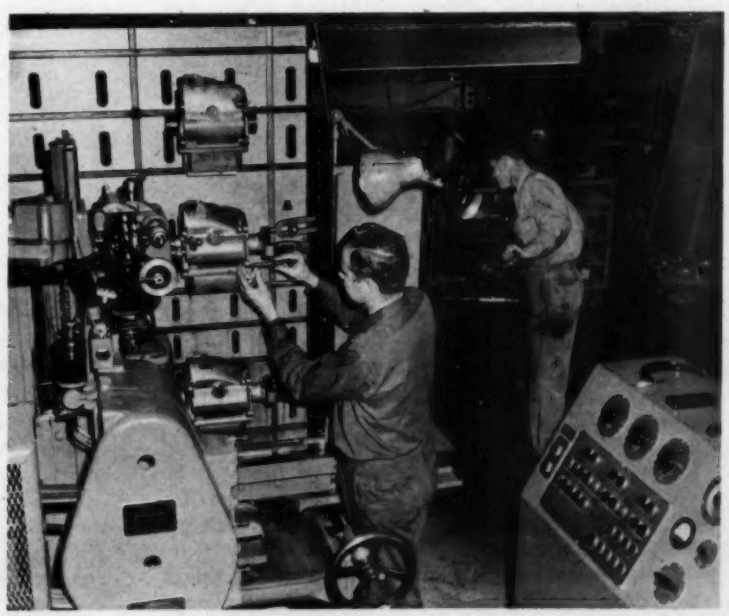


Fig. 4b — Checking duplication accuracy before removing setup from machine saves costly rework.

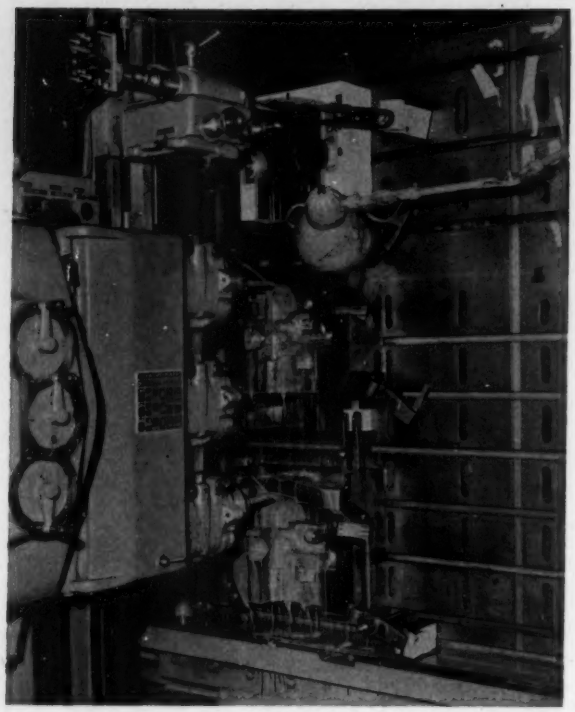


Fig. 4c — By using a hard plaster core box model two impressions of a shell core box are cut to match.

Making a hard, stable casting plaster from this plug will provide a model for duplicating any number of metal core box cavities. With extreme care, this hardwood can be built up with sheet wax or other material to the required metal thickness of the casting. From this a plaster female cast is made. In turn from this female cast, a hard plaster or plastic pattern model for duplicating any number of metal patterns will be made.

During these initial operations, the pattern casting masters can be made through the use of these same materials. Sufficient sheet wax is used to allow for approximately ⅛-in. pattern machining finish, and for the one additional shrinkage required. By using wood reinforcement, mounted patterns, cover core practice and the like, it is possible to get the required pattern castings without master breakage. Duplicating masters should be checked by the Layout department before proceeding with the steps just described for producing casting masters.

Note that the necessity for detailed involvement in the pattern configuration occurs once in the wood shop and once in the metal shop, early in the stages of making the pattern. In the old step-by-step machining practice it is necessary for this involvement, or concentration, to be carried by the metal Layout department all through the making of the equipment. A serious loss of time by critical manpower.

## SMALL PATTERNS

In the case of smaller patterns, it is often more practical to proceed in the old way up through the rough machining operations (heights, etc.) with no duplicating master prepared.

One of these smaller patterns can be machined and benched in conventional operations and used for a duplicating master after checking. Automatic multiple spindle duplicating can eliminate the need for many small master patterns. A stock of material can be kept on hand; cast, rolled or forged, from which metal patterns can be cut from the solid. Pattern deliveries can be shortened by eliminating the master and pattern casting operations.

By taking advantage of the faster stock removal in the duplicating machine, no hardship is created. By using more rolled or forged raw material for smaller patterns and core boxes, the porosity problem of pattern castings is eliminated. This is particularly important with shell molding patterns and core boxes.

Many shops have spent considerable time and effort attempting to make cast to size pattern equipment good enough for their customers. Even though master patterns are right on the button, some casting variation usually occurs, regardless of which precision casting technique is used. It is natural for metal solidifying in a mold to distort to some extent. It is common practice to machine core prints and other critical areas of precision pattern and core box castings. To properly locate these areas on a precision casting requires considerable care in the metal Layout department. In many cases it is necessary to repair or patch where distortion takes place, or surface finish is not satisfactory for bench cleanup. By using the same care on master work, coupled with duplicating practice, it is possible to get a much better job done at a comparable cost.

Many production foundries have found that excessive wear and breakage have dulled the glamour of low cost plastic patterns. While you cannot duplicate an all machined metal pattern for anything like the cost of a plastic pattern, it will outlast the plastic pattern many times. The critical problem here is foundry down-time, which in the case of a broken plastic pattern, can be extremely costly.

If sufficient perseverence can be extended along the lines just covered, the results can be amazing. Rome was not built in a day. To give modern foundries the best tool at the lowest cost in the least time, machined patterns can be competitive.

# NEW PROCESS FOR FOUNDRY IRON

by T. E. Ban, B. W. Worthington and C. D. Thompson


## ABSTRACT

A description of the D-LM Process for making iron is presented. The process makes use of low rank raw materials, conventional metallurgical equipment and techniques of pelletizing, carbonizing and smelting. A review of the history and circumstances which led one author's company to sponsor initial development of the process is given.

Pilot plant operations using a relatively low grade Lake Superior iron ore are described to illustrate the functions and technology of the various process phases. The data and results of this operation show the requirements of electrical energy, iron ore, noncoking coal, limestone and other raw materials for producing foundry grade iron. The simple and rapid method of process control for maintaining the desired iron and slag compositions is described and illustrated with diagrams. Considerations of the requirements for various capacities of commercial size plants are discussed.


## INTRODUCTION

Colonial iron making began in America as early as 1645 with the establishment of successful iron works in the Massachusetts Bay settlements. Bog and rock ores were smelted in the early blast furnaces with charcoal. Excellent grades of iron were produced and used for casting items such as pots, firebacks, scale weights and "sow" bars. The latter were forged and used for making bars, rods and flats. From this beginning, the American blast furnace has evolved into a giant that is nine stories high and produces nearly 3000 tons of iron per day. To attain high productivity and efficiency modern blast furnace technology makes use of humidified blast, fuel injections, oxygen enrichment, high top pressures and beneficiated burdens.

During the last decade beneficiated burdens, comprised of sized ore, sinter, pellets or self fluxing sinter, have been used as a means of increasing blast furnace capacities and decreasing coke requirements. Self fluxing sinter is presently being widely used because it contains the fluxes in a reacted form and possesses beneficial thermal, physical and chemical properties. A research endeavor to beneficiate burdens "beyond" self fluxing sinter led to the development of the D-LM Process, a multi-stage method of making iron.

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## PROCESS DESCRIPTION

Figure 1 presents, in simplified form, flowsheets for the modern blast furnace process and for the D-LM Process. The drawing shows iron ore and limestone formed into self fluxing sinter in a sintering plant, and the coal formed into coke in a by-product coke plant. These materials are charged as layers in the blast furnace where coke reacts with large quantities of air for heating and reducing the charge. Coke also serves as a thermally stable medium for supporting the burden within the furnace. The charge materials are resident in the furnace for periods of 8 to 14 hr, and the liquid products of iron and slag are intermittently tapped.

The blast furnace flowsheet, with the attendant sinter and coke plants, illustrates the practice of altering raw materials to conform to requirements of metallurgical equipment. The D-LM flowsheet in Fig. 1 shows the basic steps of the process in a comparative manner with the blast furnace, and illustrates metallurgical equipment and techniques arranged to conform to the characteristics of raw materials. The techniques shown are raw materials preparation, balling, carbonizing and reduction and finally smelting. Raw materials, consisting of iron ore fines, noncoking coal fines and limestone fines, are proportioned to obtain the desired metallurgical grades of iron and slag.

Preparation of the proportioned ingredients is made to obtain a suitable physical texture for balling and a homogeneous mixture for processing. Preparation may range from a simple mixing circuit, if the raw materials are comprised of concentrates and powdered fines, to a more complex grinding circuit comprised of grinding mills with attendant dewatering equipment. Normally the extent of preparation is controlled to obtain a raw materials structure of minus 20 mesh.

The prepared raw materials are formed into balls or green pellets that are about ½-in. in diameter. These can be prepared in a balling drum or a balling disc. The balls are fed in an even, shallow layer on a sintering machine where they become sequentially dried, ignited and carbonized as they are moved horizontally across the machine in a quiescent state.

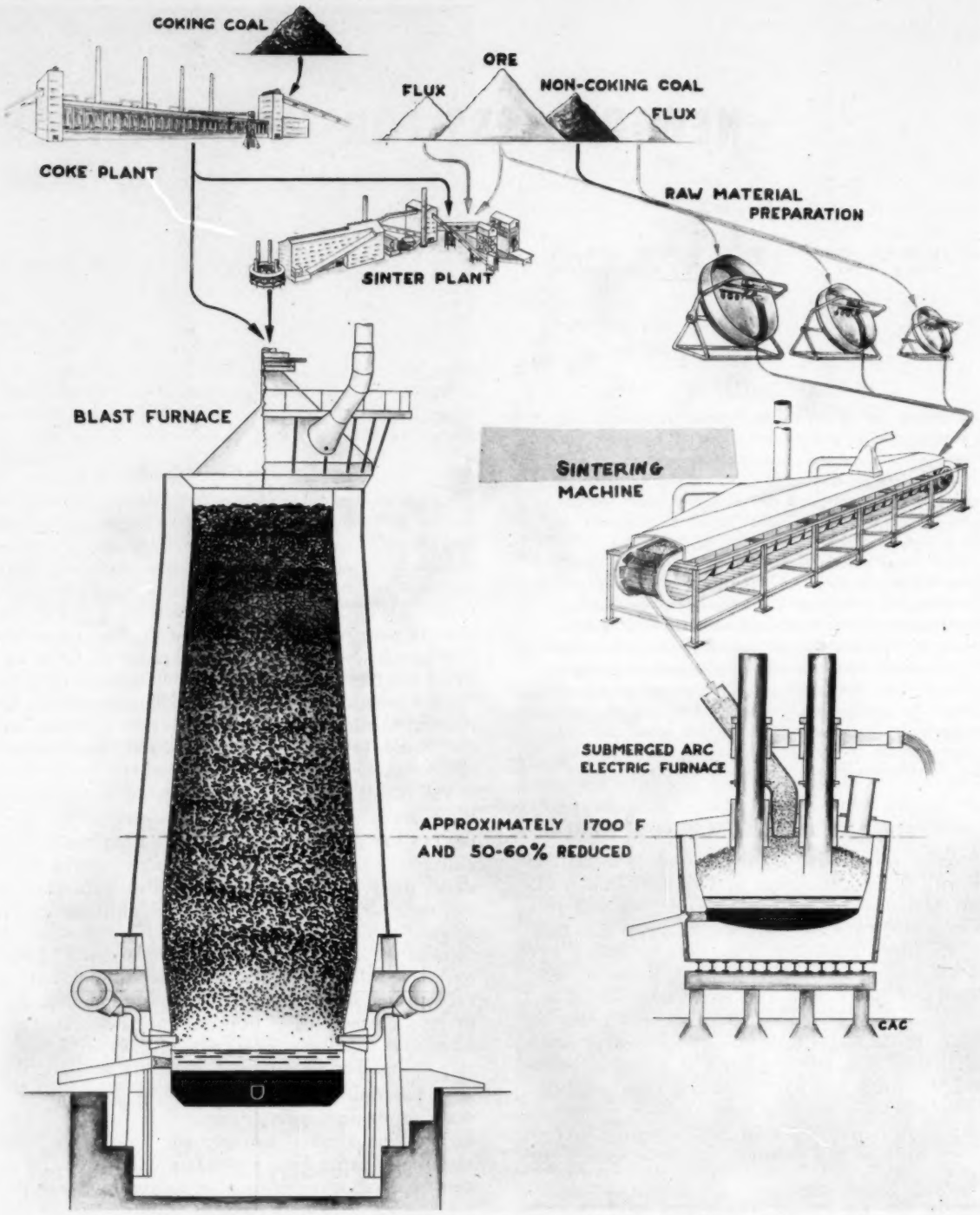



Fig. 1 — Typical modern blast furnace and D-LM processes flowsheets.

Drying is accomplished by recirculating hot gases from the carbonizing zone, and this serves to remove moisture and preheat the balls. Control of the hot gases is used to avoid spalling and preignition during drying.

*Carbonizing Treatment*

The charge is ignited with a stationary torch that serves to initiate the bed reactions in the carbonizing zone. In this zone several important phenomena take place, converting the balls into carbonized pellets. These are:

1. Complex reactions, pyrolytically processing coal into a coke-like structure as a hard char, take place. This serves to bond the ingredients in the pellets and to provide carbon for subsequent reduction and carburization in the electric furnace. The char

bond provides a thermally stable structure similar to the coke matrix of the blast furnace.

2. Combustion reactions within the pellets and pellet bed during carbonizing provide heat and reducing gases. These cause pellets to attain a temperature of about 2000 F (1090 C).
3. The oxides of the ore are reduced to lower oxide levels and some iron is reduced to the metallic state. During carbonizing approximately 50 to 60 per cent of the oxygen is removed from the oxides of the ore.
4. The calcium and magnesium carbonates of the flux are decomposed to an extent that the flux is more than 75 per cent calcined.

The product from the sintering machine is a preheated, self fluxing, self reducing pellet charge that is both chemically and physically beneficiated.

The 1700 F (930 C) line of Fig. 1 illustrates a common extent of processing for both the blast furnace and the D-LM processes. This is acquired by distinctly different metallurgical techniques. In the D-LM Process fine intimate ingredients are composited into pellets and reacted on a sintering machine. In the blast furnace process coarse and individual raw materials from a coke plant and a sinter plant are processed in the shaft portion of the furnace. At the 1700 F (930 C) line, the expired processing time of the two techniques of making iron is significantly different.

Approximately 15 to 20 min have been required in the D-LM Process to produce the carbonized pellets, while 4 to 5 hr have been required in the blast furnace to process the raw materials to a similar thermal and chemical state.

In the D-LM Process a conventional submerged arc electric furnace is used for completing the high temperature hearth reactions which produce the liquid iron and slag. This type of furnace is currently used in several European and Latin American countries for producing pig iron, and it is similar to electric furnaces used in North America for making ferro-alloys, calcium carbide and phosphorous. The furnace is particularly adaptable for final smelting in the D-LM Process because the short stock column does not require high strength burden materials, the high temperature crucible formed from the buried arc system gives rise to rapid smelting rates and minimum quantities of gases pass through the burden column. Because of charge reactivity, considerable indirect reduction takes place to give a CO to $CO_2$ ratio of about 1.6.

Raw materials of this new process are proportioned and controlled to produce target grades of iron and slag. These are usually foundry or basic grades of pig iron, although silvery pig iron and white pig iron can be made with facility. Slags are normally designed for unit basicity, but this can be altered in accordance with the desired metallurgy.

## PROCESS DEVELOPMENT

One author's company is a manufacturer of centrifugally cast water and gas pipe and, as such, is a large consumer of pig iron. During national emergencies, or when there was a shortage of pig iron, it became necessary to curtail or even discontinue operations because of a lack of this essential raw material. After a study of market conditions and plant operations in this country and in Europe, it became evident that in a highly competitive industry under severe economic conditions, those cast iron pipe companies which produced their own pig iron would possess an economic advantage. It became advisable to investigate ways and means of producing iron.

### *Iron Production Methods*

During 1948 a study was made of various known methods of producing iron. The blast furnace, acknowledged for centuries as the most economic means of producing pig iron, did not appear attractive at this time, because it was only economical when furnace capacity was in the range of 1200 to 1500 tons per day. With two widely separated pipe making plants two furnaces would be required, and the capital investment would be in the range of $50,000,000 to $80,000,000. Furthermore, such a solution would necessitate getting into the merchant pig iron business, since there would be produced 1600 tons of iron per day in excess of the requirements. Also, it was believed unlikely that the blast furnace metal could be used directly for casting.[1]

From many standpoints—unit size, compactness and low investment cost—electric furnace smelting, as represented by many successful pig iron furnaces in Europe, was of interest. However, normal electric furnace operation can compete with the blast furnace only under a specific set of conditions, namely, when cost of electricity is extremely low and when insufficient metallurgical coking coals are available. Generally, these conditions do not prevail in many industrial areas of the United States.

In view of these considerations, it was decided to discuss the problem with Battelle Memorial Institute. Before offering any specific approach to the research problem, it was proposed to first make a fundamental study of the various phases of the blast furnace process, namely, 1) preheating, 2) reduction and 3) smelting. This was initiated to determine the reactions rates that governed residence time in the blast furnace, and to indicate the type of equipment which would accomplish these functions in the shortest possible time.

These general conclusions were made:

1. A tendency for a marked increase in reaction rates between intimate mixtures of finely divided materials was confirmed. Mixtures of finely ground ore and coal were almost completely reduced in 15 min at 2000 F (1090 C).
2. The sintering machine was found to be the most effective means of preheating and prereducing large tonnages of bulk materials. With this mechanism a furnace charge could be preheated to 1600 F (870 C) in a matter of 15 min or less, whereas several hours might be required by other conventional apparatus.
3. Under proper conditions at 3000 F (1650 C) all of the normal hearth reactions of the blast furnace could be completed within one hr.

4. When pelletized mixtures of 60 parts of fine ore and 40 parts of fine coal were reacted, the product obtained was 30 per cent completely reduced and had considerable physical strength. Although some reduction was anticipated, the amount of reduction, strength of product and residual carbon retained in the sintered product were unexpected.

In view of these conclusions, it was believed that if proper conditions were maintained and satisfactory equipment was used for separating the various functions of the blast furnace, the time requirements of smelting could be decreased markedly and hence smaller tonnage operations might be economical.

### *Pilot Plant Flowsheet Circuit*

Research was continued with an extensive series of bench scale tests and the results showed optimum conditions for attaining the most desirable pellet properties on the sintering machine. This work justified establishing a small pilot plant flowsheet circuit using a patented sintering machine and a short shaft blast furnace. The results of these pilot plant experiments were successful with regard to the application of the sintering machine for prereduction and preheating, but it was learned that the short shaft blast furnace had shortcomings.

Economic projection of this type furnace indicated that it would be necessary to equip it with many auxiliaries required for a regular blast furnace. This evaluation also showed that if prereduced pellets were available, an electric furnace installation would be more economical from the standpoints of both capital and production costs.

Because of early assistance and association with the project, one author's company sponsored permanent pilot plant facilities to administrate commercialization of the process. To acomplish this, a newer and larger pilot plant was erected in Cleveland to treat a wide variety of raw materials for establishing bases for economic evalutions and engineering designs.

## PILOT PLANT OPERATIONS

This pilot plant is equipped for processing a wide variety of materials from the natural bulk state to pig iron at a capacity of about 10 tons/day. Operating techniques for treating a group of raw materials can be described by the procedures used with a certain ore, coal and limestone in an evaluation program. As an example, the procedures and techniques used for making a foundry grade of iron from a Mesabi "earthy" type of ore are described. However, it should be emphasized that many different types of iron ores and concentrates and coals have been applied in the pilot plant using various techniques.

A high volatile bituminous, 1¼ by 0 in. Ohio coal was used as a reductant, and initially it was dried and ground to minus 20 mesh. A powdered Ohio limestone was used as the flux. The dried coal and limestone powders were stored in bins for feeding the circuit. The minus ½-in. Mesabi Ore and 80 per cent of the required amounts of coal and limestone were fed from individual feeders in accordance with burden designs for making a foundry grade iron containing 3.5 per cent C, 2 per cent Si and a slag with a basicity ratio of 1.3. The raw materials were conveyed to a hammer mill equipped with ¼-in. grate bar apertures.

The hammer milling action thoroughly mixed the ingredients and reduced the coarse portions of the ore to minus ¼-in. Use of powdered limestone and coal reduced the overall moisture content of the mixture, and eliminated problems of sticking and clogging in the mill normally encountered with this type of ore. In this case the dry grinding and hammer milling represented the raw materials preparation required for processing.

The hammer milled mixture was stored in a large bin and was fed by a table feeder into a pug mill. Final additions of powdered limestone and coal were added to the mixture in the pug mill. The control of the additions was determined by analyses of the iron and slag.

### *Balling Disc and Sintering Machine*

The pug mill discharge was conveyed to the 7½-ft diameter balling disc where it was formed into ½- by ⅜-in. balls. These were fed to the 2 by 16 ft sintering machine where they were dried, ignited and carbonized. The balling and carbonizing apparatus are shown in Fig. 2. The machine discharge, consisting of hot reduced pellets, was collected in a bin and fed to a screen where minus ⅛-in. fragments and fines were removed. Sized material was discharged into a skip and raised to a furnace feed hopper, and fines were recirculated and fed in controlled amounts with the powdered limestone and coal into the pug mill.

The furnace was choke fed from the hopper through a single feed pipe located within the center of the furnace electrode delta. Figure 3 shows the electric furnace with the charging equipment. With the method of charging, the furnace was continually filled and buried electrode arcs were maintained during smelting. Electricity for smelting was supplied through three 10 in. electrodes that were connected to a 500 kva transformer. Electrical conditions were controlled by adjusting furnace voltage and electrode levels to obtain the desired furnace temperatures. Normally the product was tapped at about 2800 F (1540 C).

Pig iron and slag were tapped from a single tap hole and were collected in a series of 24 in. ladles. Slag was skimmed from the ladles, and the iron was cast in pig molds arranged in a cascading fashion. Approximately twenty 90 lb pigs were obtained from each cast. Samples of the iron and slag were taken for chemical analyses, and a standard A.S.T.M. type C-3 chill block sample was poured to obtain a visual indication of the carbon and silicon contents of the iron.

## PILOT PLANT TEST RESULTS

The operation previously described was used for a pilot plant test to evaluate a relatively poor quality Mesabi ore, a high sulfur noncoking coal and a low grade limestone. These were selected for evaluating inferior grades of raw materials in this new process.

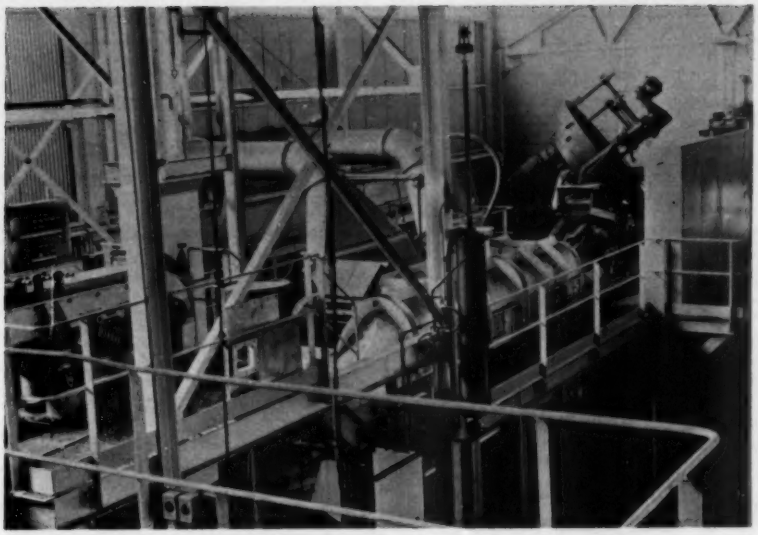


Fig. 2 — Interior view of pilot plant showing balling and carbonizing apparatus.

Analyses of these raw materials are given in Table 1. The target grade of metal for this evaluation was a foundry grade of iron containing 3.25 to 3.5 per cent C, 2.0 to 2.25 per cent Si and 0.05 per cent S. This grade was selected because it approximates the composition of Class 30 cast iron.

After completion of the pilot plant test, the operating data, samples, analyses and product weights are compiled. The results of this compilation are used to make a materials flow diagram that designates the quantities of raw materials required to produce one net ton of product. These data and the operating data serve as a basis for the evaluation of a plant.

The results of this pilot plant test are shown in the materials flowsheet of Fig. 4, and analyses of the products are presented in Table 1. The flowsheet shows the principal raw materials required for producing one ton of foundry iron in the pilot plant flowsheet to include 3830 lb of iron ore, 1803 lb of coal, 1326 lb of limestone and 1218 kwh of electrical energy.

This pilot plant evaluation illustrates the use of low rank, low cost raw materials and operating conditions with high slag volumes and relatively high electrical energy requirements. The high slag volume was beneficial for attaining a low sulfur iron with the use of the high sulfur coal as a reductant. The electricity requirement of 1218 kwh/net ton of pig iron, shown in Fig. 4, was obtained from a pilot plant furnace operation with an efficiency of about 70 per cent. Translation of these data to a commercial size furnace with an efficiency of about 82 per cent would show a requirement of 1040 kwh/net ton of iron. The basis for this scale up, and the translation of pilot plant results, has been previously shown in detail.[2]

## PROCESS CONTROL

Control of the raw material ratios is used as the principal means of process control for maintaining the target grades of iron and slag. Through use of additive feeders, additions of powdered limestone and coal can be controlled to meet a desired metal-

TABLE 1 — ANALYSES OF RAW MATERIALS AND PRODUCTS

| Analyses (Natural Basis) | Raw Materials | | | Products | |
|---|---|---|---|---|---|
| | Iron Ore | Coal | Lime-stone | Iron | Slag |
| Moisture, % | 10.0 | 2.5 | 0.5 | — | — |
| Fe, % | 47.6 | 2.8 | 0.4 | — | 0.7 |
| Acids ($SiO_2$ & $Al_2O_3$), % | 11.5 | 8.2 | 5.9 | — | 42.4 |
| Bases (CaO & MgO), % | 1.4 | 0.2 | 48.3 | — | 56.1 |
| LOI, % | 7.2 | — | 42.5 | — | — |
| S, % | 0.1 | 2.5 | 0.2 | 0.06 | — |
| VM, % | — | 36.6 | — | — | — |
| C, % | — | 48.3 | — | 3.53 | — |
| Si, % | — | — | — | 2.02 | — |
| Mn, % | — | — | — | 0.72 | — |
| P, % | — | — | — | 0.12 | — |

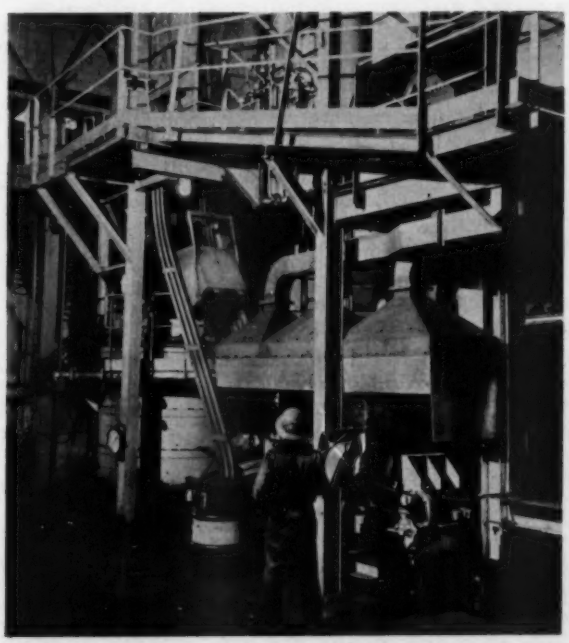


Fig. 3 — Interior of pilot plant showing screen, skip and electric furnace.

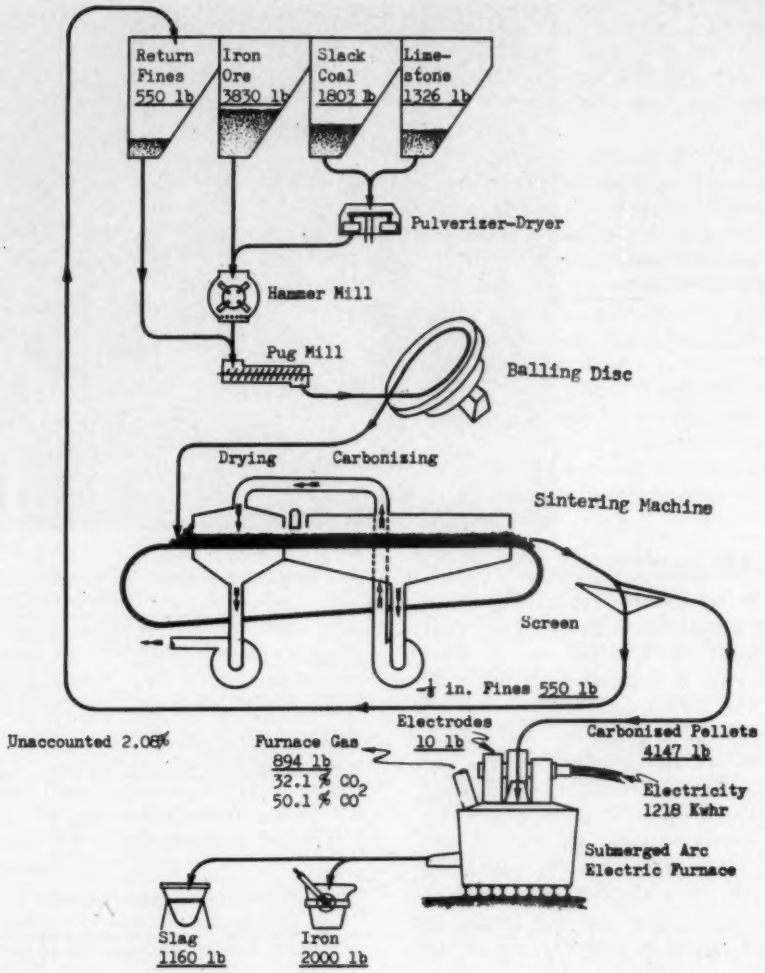



**Fig. 4 — Materials flowsheet of pilot plant operation using Mesabi iron ore mixture, Ohio coal and powdered limestone. Basis: Natural raw materials for one net ton of iron.**

lurgical change. Within 15 min following such a change, the composition of the carbonized pellets or electric furnace feed alters in accordance with the charge. This method of control is illustrated in Fig. 5, which shows changes in the coal to ore ratio to be followed within 15 min by changes in the per cent of carbon in the carbonized pellets.

The FeO in the slag reflects the changes of carbon in the system and lags the change of the coal to ore ratio by about 1½-hr. Similar control is shown by the changes of the limestone to ore ratio and the resulting basicity of the slag. Figure 5 illustrates data obtained during a period when experimental control was being examined in the pilot plant.

Control of the pellet composition can also be attained by altering the retention time and draft rate of carbonizing. However, this form of control is used less frequently because it disturbs optimum operating conditions. Control of the temperature within the furnace, and to a limited degree the composition of the products, is obtained by altering the electrical operating conditions of the furnace.

## COMMERCIAL PLANT CONSIDERATIONS

Many varieties of ores, coals and fluxes have been tested in this pilot plant. The purpose of the tests is to establish raw material requirements, to acquire basic engineering design and to obtain necessary data for economic evaluations of raw materials in this process. There are many technical reasons for this evaluation. The raw material preparation for each type of ore differs and may include crushing, dry grinding, wet grinding or simple blending. The burning and carbonization characteristics of each coal differ and will require research for determining the proper operating conditions to obtain optimum pellet strength and reduction. It is also important to obtain data of the operating conditions for the electric furnace with regard to each group of raw materials and the desired analysis of iron.

Translation of the pilot plant data into economic considerations requires a thorough evaluation of the developed flowsheet along with the engineering layout. The flowsheet determines raw material, energy and equipment requirements. The engineering lay-

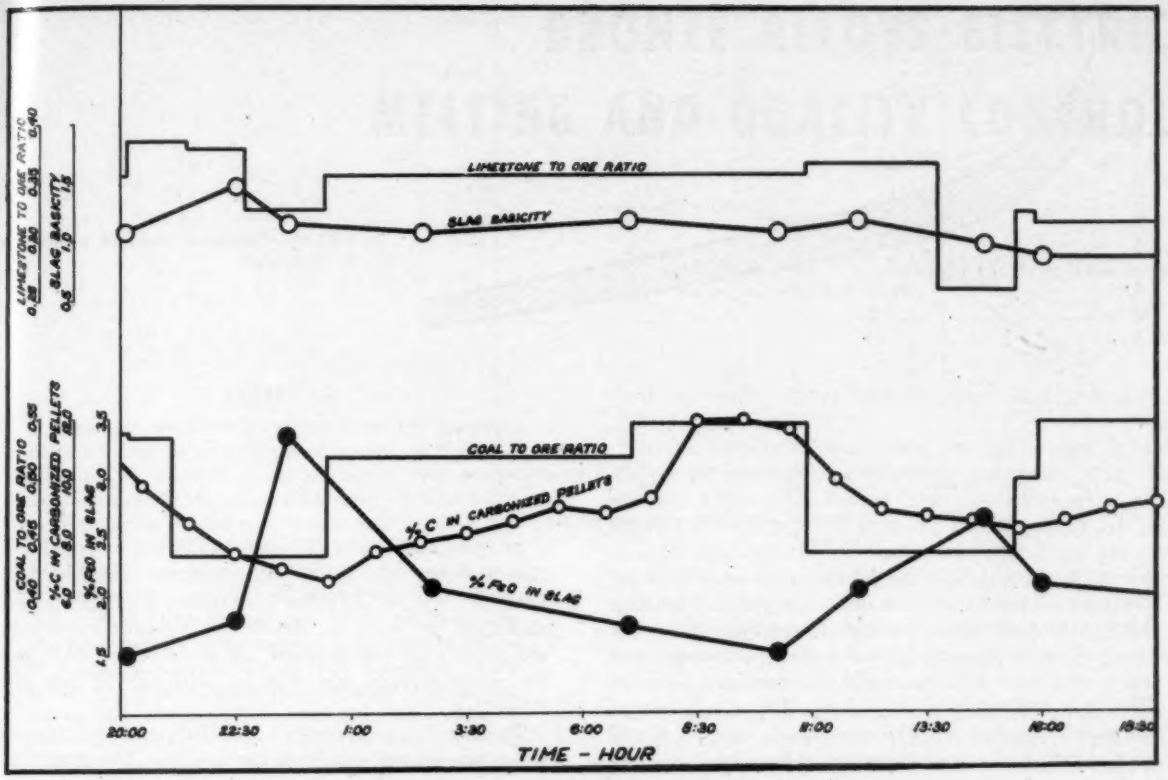



Fig. 5 — History of metallurgical effects from burden alterations.

out designates labor or manpower requirements and capital charges.

Raw material requirements and costs will depend on the grade of the material. Transportation costs will also influence raw materials costs at the proposed plant. Examples of economic evaluations using relatively high grade raw materials in this process were presented in a recent publication.[3] This showed the costs of iron ore and electricity to vary considerably with location of the plant site. The examples illustrated raw materials costs to change pig iron production costs about $10 when there is a variation of ore costs of $3, coal costs of $2 and electricity costs of 4 mils.

Relatively poor grade raw materials were used in the previously described pilot plant tests. The Mesabi iron ore contained 47.6 per cent Fe (natural basis), the coal was a noncoking coal containing 2.5 per cent S and the limestone contained 5 per cent acid insolubles. Even though these low grade raw materials may be obtained at a low unit costs, the advantages may be offset by the greater energy requirements for producing iron with high slag volumes. An estimated cost of these raw materials in a Great Lakes Area would be about $36, as presented in Table 2. These can be contrasted to higher grade raw materials in the St. Lawrence River Area, also given in Table 2. These costs include iron ore, coal, flux, electrodes, water and total electricity.

Production cost considerations for different size plants for this process include the raw material costs and operating costs which will vary with the size or capacity of the plant. Figure 6 illustrates operating costs as a function of the daily capacity of different size plants. As the tonnage decreases from 550 to 100 net tons/day, the operating costs, based on 336 days/year operation, increase from $9.85 to $23.52. This increase is attributed to increased manhour requirements and increased fixed costs including interest, amortization, insurance, etc.

From these data and the data of Table 2, it can be seen that in an area with a favorable raw material situation such as the St. Lawrence River area, total production costs for a net ton of foundry iron can range from $38 for a 550 net ton/day plant to

TABLE 2 — ESTIMATED RAW MATERIAL COSTS

| Raw Material | Lake Erie Area Mesabi Ore Quantities | Unit Cost | $/ NTPI | St. Lawrence River Area[2] Canadian Concentrate Quantities | Unit Cost | $/ NTPI* |
|---|---|---|---|---|---|---|
| Iron Ore | 1.71 GT @ | $10.61 | $18.14 | 1.52 GT @ | $9.50 | $14.44 |
| Noncoking Coal | 0.90 NT @ | 6.34 | 5.71 | 0.84 NT @ | 9.00 | 7.56 |
| Limestone Fines | 0.66 NT @ | 3.08 | 2.03 | 0.20 NT @ | 2.50 | 0.50 |
| Electricity | | | | | | |
| Smelting | 1040 kwh | | | 890 kwh | | |
| Auxiliary | 60 kwh @ | 0.008 | 8.80 | 60 kwh @ | 0.004 | 3.80 |
| Electrodes | 10 lb @ | 0.14 | 1.40 | 8.2 lb @ | 0.14 | 1.15 |
| Water | 9000 gal @ | 0.03/1000 | 0.27 | 9000 gal @ | 0.03/1000 | 0.27 |
| | | | $36.35 | | | $27.72 |

*Dollars per net ton of pig iron.

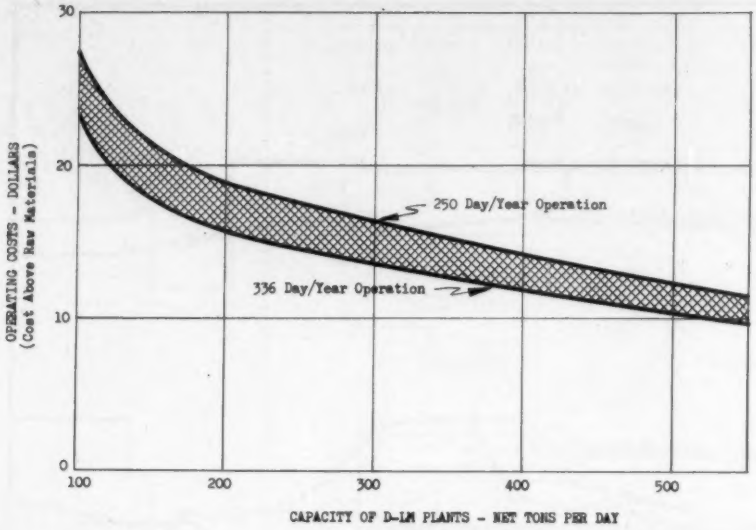



Fig. 6 — Operating costs as a function of capacities.

$51 for a 100 net ton/day plant based on 336 days/year operations. In an area with a slightly less favorable raw material situation such as the Lake Erie area, the total production costs for a net ton of foundry iron can range from $46 for a 550 net ton/day plant to $59 for a 100 net ton/day plant. These cost ranges compare favorably with the estimated cost of $58.40 for hot metal in a possible average foundry.[4]

If fewer operating days per year are to be considered, the operating costs will be increased from $2 to $4/ton depending on the size of plant. These costs are illustrated as the upper line in Fig. 6, and are based on operations at 250 days/year (five day week).

## SUMMARY

The D-LM Process is a novel method of making iron using low rank raw materials and conventional metallurgical equipment. Iron ore, noncoking coal and limestone are prepared and blended in the metallurgical proportions required for making pig iron and slag. Blended fine materials are formed into balls about ½-in. in diameter and are placed on a sintering machine where they are sequentially dried, ignited and carbonized. During carbonization, the coal forms a coke-like matrix which bonds the pellets, the iron ore is reduced to an extent where more than 50 per cent of the oxygen is removed from the ore and the limestone is partially calcined. The hot, self-fluxing, prereduced pellets are charged to a submerged arc electric furnace where final smelting takes place, and liquid iron and slag are formed. Liquid products are tapped intermittently and the iron can be pigged or used as hot metal.

The operating procedures for evaluating a relatively poor grade Mesabi Ore and a high sulfur coal in a pilot plant included preparing the raw materials by dry grinding the coal and hammer milling a mixture of iron ore, coal and powdered limestone. This mixture was then balled, carbonized and smelted in the laboratory flowsheet circuit. The data from the continuous test were carefully evaluated and showed that the 500 kva pilot plant furnace required 1218 kwh/net ton of pig iron containing 3.53 per cent C and 2.02 per cent Si. Based on previous experience this electricity requirement can be scaled to 1040 kwh for a commercial size furnace with an 82 per cent efficiency.

Principal process control is maintained by changing the proportioned ratios of ingredients. The new ratio is shown to effect the composition of the furnace feed within approximately 15 min following the change. The composition of the carbonized pellets can also be controlled by draft conditions on the sintering machine.

Considerations for various size commercial iron plants for this process are shown in the operating costs of plants decreasing in capacity from 550 to 100 net tons/day. The operating costs range from about $10 to $23/net ton of foundry grade iron. Total production costs will depend on the sum of the operating costs and raw material costs, and can range from about $38/net ton of iron to about $59/ton of iron depending on the plant capacity and on the grade and source of raw materials.

## ACKNOWLEDGMENT

The writers express appreciation to three organizations for their contributions to this development. In this respect the contributions by McWane Cast Iron Pipe Co., McDowell Co., Inc. and Battelle Memorial Institute are acknowledged. Special efforts for this presentation by Messrs. D. C. Violetta, C. A. Czako and C. J. Nelson are appreciated.

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# BRONZE ALLOYS ELECTRIC MELTING AND QUALITY CONTROL

by R. V. Barone, T. F. Godfrey and R. A. Rosenberg


## ABSTRACT

A general discussion is presented of low frequency induction melting including melt covers and theory of operation. A method of quality control was devised and employed in the authors' company's bronze foundry to maintain stricter control over melt analyses. The zinc losses of A.S.T.M. B62 (85-5-5-5) were found to be 0.8 per cent during melting and procedures were then developed to replace these losses. The change in zinc content (from 4.8 per cent to 5.5 per cent) produced an increase in the average yield from 16,250 psi to 16,750 psi, while exhibiting little correlation of the tensile strengths and elongations obtained from approximately 70 production tests per month. This resulted in economic savings of over $700 per month on raw materials, and in improved consistency in alloy foundry properties and mechanical properties.

X-ray fluorescence was employed to monitor the zinc content of the 85-5-5-5 alloy over a period of 4 months, in addition to approximately 70 wet chemical analyses made per month over the entire 10 month testing period. The use of x-ray fluorescence is discussed and its advantages as an accurate, rapid means of analysis is cited.


## INTRODUCTION

The bronze foundryman has always desired maximum yield from both his equipment and his time. In order to fulfill this wish and to obtain and maintain an operation at peak efficiency, a system of quality control which includes systematic reviews of all foundry operations is essential.

Quality control, beginning with the initial charge segregation, must be maintained until the completed casting leaves the foundry. It is of extreme importance to check on the condition of a casting in various stages of its production, for if a casting is made poorly in the foundry it becomes both useless and wasteful to perform any subsequent operations on it.

In melting bronze, it is imperative that deleterious impurities be minimized. Major impurities to be guarded against in melting tin bronzes and red brasses fall into two categories. The first includes metals such as iron, antimony, silicon and aluminum which alter the foundry properties. Next, are the nonmetallics such as sulfur, sand, slag and gases including carbon monoxide, carbon dioxide, water vapor and sulfur dioxide.

These impurities may enter the melt either in the charge or from contact of other materials with the molten alloy. The former may be adequately controlled by proper segregation and cleaning of the charge, thus minimizing both mixed metal and gangue forming materials. The later source includes all substances which contact the molten metal. Furnace linings, ladle linings, skimmers, molds, slags, covers and the atmosphere fall into this category and all should be as clean and dry as possible.

The purpose of this work was to devise a procedure for the close control of metal chemistry. This was felt to be important since if metal chemistry varies extensively, 1) the resulting alloy may fall out of chemical specifications limits, 2) metal costs may be excessive since low cost alloy constituents may not be maximized, 3) the mechanical properties may not meet specifications and 4) the foundry properties (e.g, fluidity, shrinkage, etc.) may change thus producing erratic casting results.

## LITERATURE REVIEW

### *Melt Control Impurity*

Foreign elements such as silicon and aluminum are known to have detrimental effects on casting properties. Colton[1] presents data on red brasses which illustrate that as little as 0.005 per cent of either aluminum or silicon will cause a reduction in both tensile strengths and per cent elongation, while 0.20 per cent aluminum, 0.02 per cent manganese or 0.10 per cent magnesium will cause the shrinkage formation to change from a randomly dispersed interdendritic type to a gross localized type.

St. John[2] points out that for alloys such as red brass and tin bronze, even a change in metal chemistry within specification limitations may alter the optimum pouring range considerably. For example, a variation of the iron content from 0.05 to 0.25 per cent may cause a 50 F difference in pouring temperature. Variations in nickel content may also produce considerable change. Ames[3,4] has shown that the pouring temperature affects the solidification characteristics of tin bronzes, and that under production conditions it can have a profound effect on pressure tightness, tensile strength, ductility and density.

Its effect on the physical properties of A.S.T.M. B62 (85-5-5-5) has been substantiated by the A.S.T.M.

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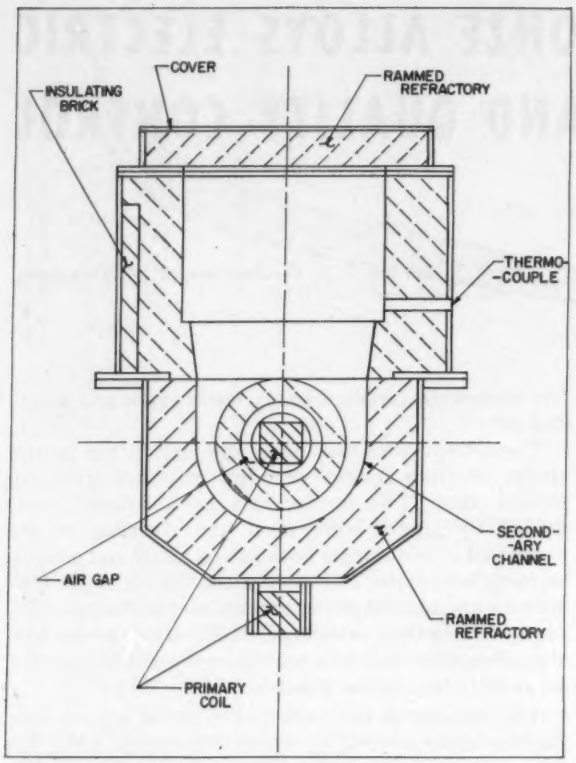



**Fig. 1 — Schematic cross sectional view of one at the low frequency induction furnaces.**

production tests[5] performed at the authors' company. Due to this, a pouring range for not only the cast tensile bars but also each individual casting made in the company is specified.

Unfortunately, the literature on the removal of impurities is limited. However, there are several articles which discuss the production of copper and copper-base alloys which may be helpful if impurity reduction is necessary.[6,7,8,9,10,11]

### *Melt Covers*

Basically, there are five types of covers which can be used in the refining process—acid, basic, neutral, oxidizing and inert. Silica is the most common acid flux which will react with basic or amphoteric substances and produce a silicate slag. The basic fluxes include lime (CaO), either hematite or magnetite, soda ash or sodium carbonate. These react with the acidic substances. Borax ($Na_2B_4O_7$) and fluorspar ($CaF_2$) are two common neutral fluxes employed where there is either little slag forming material, or when it is desirable to reduce the liquidus temperature of the slag to produce a fusible material which will absorb the impurities.

The oxidizing type would include materials such as niter or sodium sulfate which supply oxygen and remove impurities by a process involving oxidation reactions. Another cover used primarily to prevent or decrease metal oxidation is the inert cover which works by reducing the amount of contact between the melt and the atmosphere. These include gases which do not react with bronze such as argon or nitrogen.

Baer, et al,[12] reported their results on the mechanical properties of the valve bronze melted, 1) in air, 2) under dry charcoal and 3) under damp charcoal, which showed that the first produces favorable mechanical properties, the second is much less favorable and the third method results in extremely poor properties. With the low frequency induction furnaces at the authors' company, a graphite flake cover, if used prior to 8 hr shut down in the evening, helps to prevent a hard crust buildup of material in the furnaces without noticeably affecting the mechanical properties of the material. The latter was determined by comparing the tensile data of heats poured early in the day with those for heats poured later.

## EXPERIMENTAL PROCEDURE

### *Melting Apparatus*

Melting was carried out in low frequency induction furnaces using commercial B-62 ingot. Chemistry and physical properties of the alloy are listed in Table 1. A schematic diagram of one of the two furnaces used is shown in Fig. 1. The furnace consists essentially of, 1) steel laminations, 2) primary coil, 3) refractory lining and 4) secondary channel with the pool of molten metal. Furnaces employed were 6000 lb capacity, and are well suited for melting bronze alloys.

**TABLE 1 — STANDARD SPECIFICATIONS FOR 85-5-5-5 ACCORDING TO A.S.T.M. B62-52**

| | Chemical Requirements, % | |
|---|---|---|
| | Minimum | Maximum |
| Copper | 84.0 | 86.0 |
| Tin | 4.0 | 6.0 |
| Lead | 4.0 | 6.0 |
| Zinc | 4.0 | 6.0 |
| Nickel | — | 1.0 |
| Iron | — | 0.30 |
| Phosphorus | — | 0.05 |
| **Tensile Requirements** | | |
| Tensile Strength, psi | minimum | 30,000 |
| Yield Strength, psi* | minimum | 14,000 |
| Elong. in 2 in., % | minimum | 20 |

*Yield strength is determined as the stress producing an elongation under load of 0.5 per cent which is 0.01 in. in a gage length of 2 in.

Low frequency units such as were employed are especially good for brass and bronze melting because:

1. Melt losses incurred are low, thereby effecting significant savings in material costs. Also, good reproducibility is thereby provided; melt chemistry variation due to metal losses is minimized.
2. Furnaces inherently set up a stirring action in the metal which tends to reduce inhomogeneity in the bath.
3. The furnace is efficient in transferring electrical power to heat since the heat is generated in the metal itself. In a 6 month period (6,984,000 lb), the cost of power averaged $6.44/ton, including

the cost of holding the furnaces on low power at night, during holidays and weekends.

4. Since metal is in the furnace at all times, thermal fluctuations are minimized so that the refractory life is excellent. The monolithic hand rammed linings, which are currently in service, have thus far been used for melting about 10,000,000 lb of metal. It appears that they will last for at least another two million lb.
5. Working conditions around the furnaces are comparatively good, since the temperature in the working area is low and the atmosphere relatively clean.

In general, the furnaces are best suited for melting one alloy. However, they may be used interchangeably between two similar alloys such as *M* Metal and 85-5-5-5.

The power input for the induction furnaces is dependent on the magnetic flux induced in the continuous loop of metal, illustrated in Fig. 1, which forms the secondary coil. The power input can be varied by applying a higher voltage. This is done by decreasing the number of effective turns on the furnace coil by means of tap switches. The power adjustment is used to compensate for alloys with different electrical conductivities. Several taps are provided so that the melting rate can be varied. A low voltage tap is also provided for holding the heat to prohibit metal solidification in the channel while the furnace is not in use for production.

### *Chemistry Control*

In order to illustrate that the control of metal chemistry to close tolerances was possible, it was decided that the zinc content of the B-62 metal* would be monitored. Zinc was chosen because it is the most difficult element to control due to vaporization losses which occur. In addition to the regular chemical samples which are analyzed by wet chemistry in accordance with the A.S.T.M. material standards, one sample was cast from each of the two furnaces used to melt B-62 alloys for every 4 hr of operation, and analyzed for zinc using x-ray fluorescence. Figures 2 and 3 show a picture of the x-ray fluorescence unit and a schematic diagram of its operation, respectively.

X-ray fluorescence[13] is extremely useful, since the method analyzes for zinc directly with both good precision and accuracy. In addition, the method is not a function of the state of chemical combination of an element so that no chemical separations need be made as in wet chemical analysis. Analytical results may be obtained within 10 min after a sample is cast so that the unit can therefore be used for control purposes. Another advantage is that the specimens are undamaged by the analysis; thus "standard samples" have an indefinite life.

X-ray fluorescence is most useful for the elements which have characteristic wave lengths between 0.5 and 2.5 Å (1 Å = $10^{-8}$ meters).[13] This is due to the fact that longer wave lengths which are produced by the elements of lower atomic number are readily absorbed by both air and the counter window. Thus, they are not easily measurable. The lower wave length limit which corresponds to the elements of higher atomic number is imposed primarily by the voltage placed on the x-ray tube, since high voltages must be used to excite short wave length high energy x-rays. Fortunately, due to the atomic constitution of the elements, the ones with high atomic numbers emit secondary characteristic x-rays in the desired range which although not as intense, can be excited with lower voltages and may be used for analysis.

*Deoxidized with 2 oz of 15 per cent copper phosphorus/100 lb of metal.

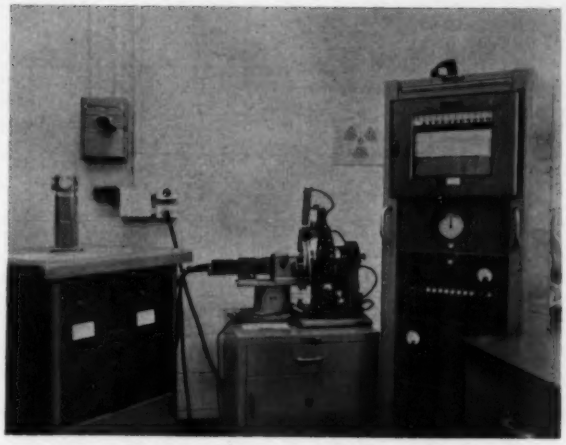


Fig. 2 — X-ray fluorescent equipment used for determining metal chemistry.

### *Plant Operation*

In order to see where improvements could be made in controlling metal chemistry, the flow of materials through the plant was examined. Figure 4 is a schematic diagram of the operation and shows that the charge consists of gates and sprues, scrap, chip, ingot and zinc. The diagram illustrates that a majority of

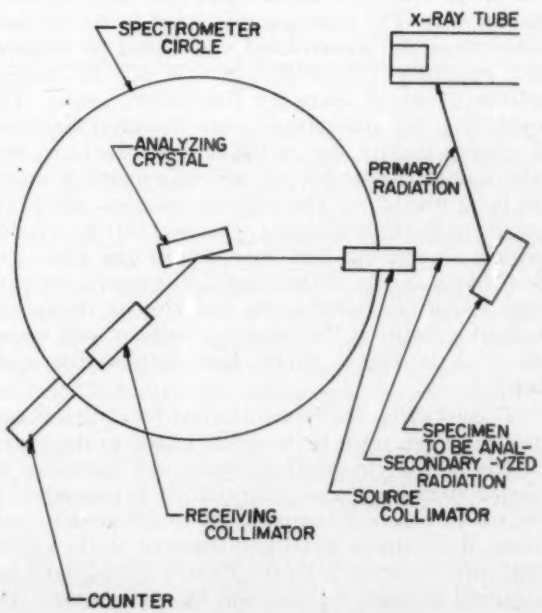



Fig. 3 — Schematic diagram of x-ray equipment.

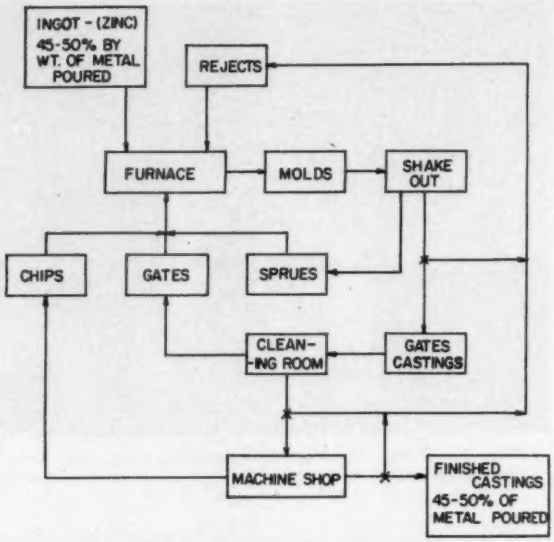



**Fig. 4 — Schematic diagram of the flow of materials in the operation. Average production data for B62 — green sand — 475,000 lb/mo, yield 50-55 per cent; shell sand — 175,000 lb/mo, yield 40-45 per cent.**

the charge components are recycled material, and that the only new components are ingot and zinc. Similarly, all the metal which leaves the system is in the form of good castings, melt loss and metal handling losses.

Obviously, to obtain a balanced system, the weight of metal added should equal the weight of metal leaving the system, namely, the castings, plus the losses which are incurred in the process by vaporization in the slag or in handling. Next, the flow of materials being recycled was studied to see if a definite pattern could be established.

Figure 4 illustrates that after the metal is poured, the sprue heads are accumulated and then returned to the melt. The cycle for this metal is the shortest, since these are accumulated and added in amounts not exceeding 10 per cent of the 1000 lb charges, within 24 hr of operating time after casting. The cycle time for the returns from the cleaning room fluctuates slightly, due to the amount of cutting that the castings require but are normally remelted within 48 hr of operation. The majority of chips which are produced in the machining operation will be recycled within 15 working days from the casting date. The last source of the charge material, namely, the castings which are found to be defective by inspections occurring between the processes outlined and shown as an *X* in Fig. 4, would have intermediate cycle times.

Theoretically, the system is ideal for chemical control purposes, since if the metal added to the system is within the composition limits and there are no preferential metal losses incurred, it is impossible to be out of chemical composition. Under actual conditions, if no mixed metal is introduced to the system, the only deviation from the ideal is due to melt loss incurred in both the slag and by vaporization. The latter is the most important for chemical control, since analysis of the slag has shown that the loss here for copper, zinc, lead and tin are in approximately the same ratio as in the base alloy. There is preferential removal of iron, however, and thus this constituent is the only one being reduced.

### *Chemistry Control*

Thus, the chemistry of any major element may be controlled by adjusting the level of the new material and by compensating for any losses. On this basis, to stabilize the zinc content of the cast material a sufficient amount of zinc must be added to compensate for the zinc lost through vaporization, and the zinc level of the new material added to the system must also be adjusted to the desired final aim. It was found that the latter could be accomplished since the composition of the ingot is known. Thus, it was decided that the correct amounts would have to be determined empirically and that in time, if a consistent practice was used, the overall composition would approach a given quantity asymptotically.

The control system was put into effect first by marking the zinc content on each pallet of ingot (approx. 2000 lb) as it was received, using the certified chemical analysis supplied by the ingot manufacturer. The table, shown in Fig. 5, was then used to relate the various ingot compositions and amount of ingot in the charge to the weight of zinc which should be added. This was given in terms of a number of pure zinc blocks of known weight. The resulting addition which appeared to give the most consistent data for an aim of 5.5 per cent zinc allowed 8 lb/1000 (0.8 per cent zinc) for the zinc lost in

**85-5-5-5 ZINC ADDITIONS FOR A 1000 LB CHARGE**

| Zn (Ingot), % | Weight of Ingot in 1000 lb Charge | | |
|---|---|---|---|
| | 900 | 650 | 350 |
| | No. Zinc Blocks | | |
| 5.7 | 3 | 3 | 3 |
| 5.6 | 3 | 3 | 3 |
| 5.5 | 3 | 3 | 3 |
| 5.4 | 3 | 3 | 3 |
| 5.3 | 3 | 3 | 3 |
| 5.2 | 3 | 3 | 3 |
| 5.1 | 4 | 3 | 3 |
| 5.0 | 4 | 4 | 3 |
| 4.9 | 4 | 4 | 3 |
| 4.8 | 4 | 4 | 3 |
| 4.7 | 5 | 4 | 3 |
| 4.6 | 5 | 4 | 4 |
| 4.4 | 5 | 5 | 4 |
| 4.3 | 6 | 5 | 4 |
| 4.2 | 6 | 5 | 4 |
| 4.1 | 6 | 5 | 4 |
| 4.0 | 6 | 5 | 4 |

| No. Zinc Blocks | Weight, lb |
|---|---|
| 3 | 10 |
| 4 | 13 |
| 5 | 16 |
| 6 | 19 |

**Fig. 5 — Chart used to determine the proper zinc additions for ingots of varying zinc content.**

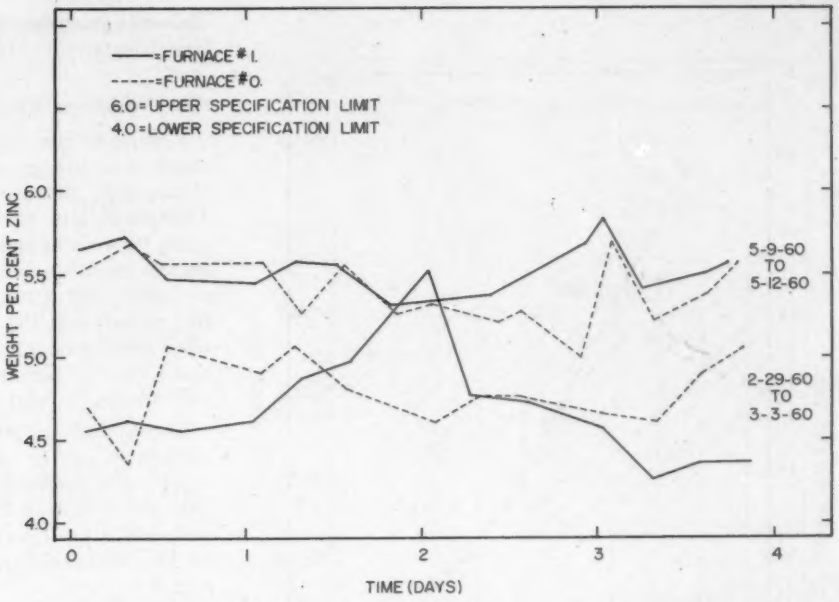



Fig. 6 — Plot of the zinc content in the B62 metal vs. time.

vaporization plus the amount of zinc necessary to bring the ingot composition up to 5.5 per cent.

Note that the figures were given for three different charges which had 350, 650 and 900 lb of ingot, respectively. This was done to cover any charge component shortage due to fluctuations in the flow of recycled materials. The results of the additions were followed daily by a plot of analyses versus time. Figure 6 is a plot made when the additions were just started and during a week nearer the end of the project, respectively.

In observing the plot of melt zinc content over a period of time, it was noted that a larger loss was incurred over the weekends. Experimentation showed the amount necessary for replenishing this loss amounted to 0.5 per cent of the metal in the furnace. Eventually, this amount was added early Monday mornings to compensate for the weekend vaporization losses.

### *Tensile and X-ray Tests*

Since it was mandatory that the foundry properties of the alloy should not be negatively affected during standardization of the zinc content, data were obtained from the daily tensile test results and wet chemical analyses made in accordance with the A.S.T.M. chemical and physical requirements.

During the last four months the zinc chemistry was monitored by x-ray fluorescence. The advantage of this was that the bronze was analyzed for zinc directly rather than by difference, and also that the results could be obtained within 10 min of casting so that any changes due to furnace additions could be checked while the metal was still in the furnace and corrections could be made if necessary.

A working diagram for zinc is shown in Fig. 7. The reproducibility of the working diagram and x-ray analysis method was checked periodically. In addition to analyzing samples by wet chemistry in order to check the results, several other tests were made. One involved pouring a series of five test bars from the same crucible and analyzing the five samples. The results are given in Table 2. Note, that the standard deviation of the value obtained for the per cent Zn was only 0.04 per cent.*

**TABLE 2 — ANALYSIS OF 5 SAMPLES**

| Sample | Avg. Time for 32,000 counts (sec) | Zn, % |
|---|---|---|
| 1 | 42.5 | 4.96 |
| 2 | 42.4 | 4.97 |
| 3 | 42.5 | 4.96 |
| 4 | 42.2 | 5.02 |
| 5 | 42.0 | 5.06 |
| Avg. | 42.3 | 4.99 |
| Standard Deviation | 0.21 | 0.04 |

The condition of the surface of the specimen and its effect on the fluorescence results was also investigated. It was found that this could alter the analyses, since the amount of characteristic x-ray produced is a function of the area exposed to the primary radiation and the surface area exposed varies with the degree of surface polish. The cast samples, an example of which is shown in Fig. 8, were all finished on a 60 grit belt the same as the standard sample used to standardize the machine prior to each group of analyses, so that the surface area exposed remained constant.

It was also found that approximately $\frac{1}{64}$-in. or more of the surface of the cast specimen had to be removed in order to obtain reproducible analyses.

*The fact that x-ray fluorescence is influenced by the other alloying constituents is exhibited in Fig. 7, which shows a working diagram for B-61. Note that the slope of the curve changed. Thus, working diagrams have to be made for each series of alloys.

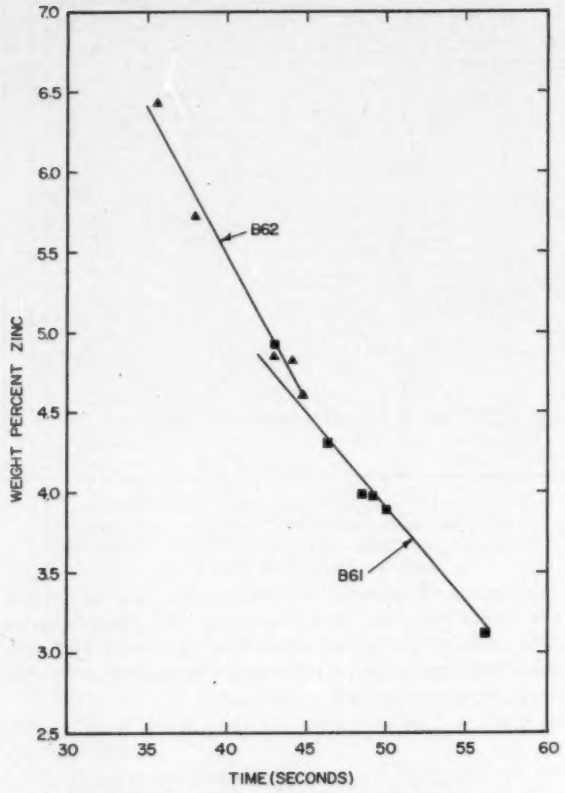



Fig. 7 — X-ray fluorescent analysis working diagram for the analysis of zinc in B61 or B62.

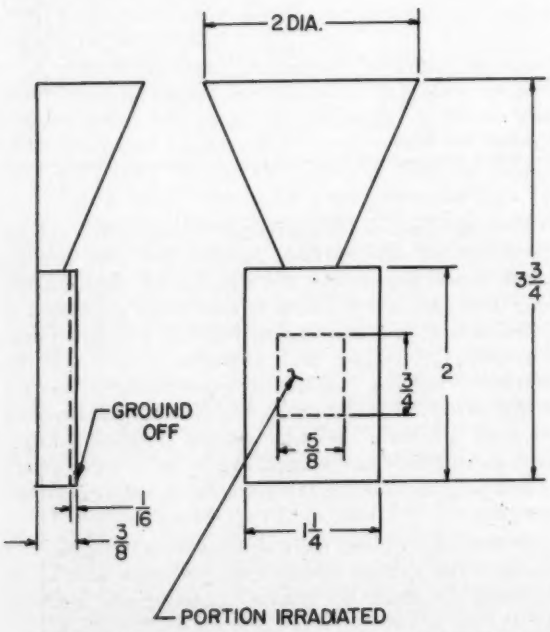



Fig. 8 — Cast sample specimen used for x-ray fluorescent analysis.

This was probably the result of segregation and surface defects.

### *Zinc Additions Results*

Results of the zinc additions are shown in Fig. 9, which is a bar graph of the average zinc chemistry obtained by wet analysis for 10 months of operation. The curve also contains the average data obtained using fluorescent analysis. The largest deviation found for the monthly averages between the x-ray and wet chemistry was 0.15 per cent, which occurred in the first month that the x-ray tests were made. The variations were probably caused by the sample pieces not being taken simultaneously, and also because the wet zinc analysis is obtained by difference.

Corresponding standard deviations for the four months in which the fluorescent data were obtained exhibited a decrease, and are listed in Table 3. Note, that the deviation for the first two weeks of March was ±0.28 per cent, while during the last two weeks of May the standard deviation was only ±0.20 per cent.

TABLE 3 — STANDARD DEVIATIONS FOR FOUR MONTHS OF FLUORESCENT DATA

| | Zn, % | | | | | |
|---|---|---|---|---|---|---|
| Month | Wet Analysis | X-Ray Analysis | Deviation | Y.S., psi | Elong., % | T.S., psi |
| Sept. | 4.81 | | | 16,256 | 26.1 | 35,597 |
| Oct. | 4.75 | | | 16,265 | 26.0 | 35,308 |
| Nov. | 4.92 | | | 16,472 | 24.9 | 35,080 |
| Dec. | 4.94 | | | 16,464 | 26.0 | 35,476 |
| Jan. | 4.96 | | | 16,528 | 25.7 | 35,291 |
| | | | 284* | | | |
| Feb. | 5.05 | 4.90 | 292 | 16,671 | 25.9 | 35,401 |
| March | 4.91 | 4.81 | 299 | 16,585 | 25.6 | 35,299 |
| April | 5.20 | 5.23 | 227 | 16,735 | 25.5 | 35,400 |
| May | 5.50 | 5.43 | 216 | 16,665 | 25.6 | 35,229 |
| | | | 200** | | | |
| June | 5.51 | | | 16,857 | 24.8 | 35,301 |
| $\bar{x}$ | 5.06 | | | 16,550 | 25.6 | 35,338 |
| $\sigma$ | 267 | | | 194 | 477 | 140.6 |
| r | | | | +0.8 | −0.5 | −0.2 |

*First two weeks of Feb. using fluorescent analysis.
**Last two weeks of May using fluorescent analysis.

The plots, of the average monthly tensile, yield and per cent elongation vs. time, are illustrated in Figs. 10, 11 and 12, respectively. Each of these points is the average value of approximately 70 tests. These data show no significant change in either the average per cent elongation or tensile strength. However, the yield strength does increase noticeably.

To see if the change in any of the three mechanical properties measured were significant, linear correlation factors (Appendix) were calculated for all these variables versus the average zinc content. The wet chemical data were used in this case, since the fluorescent data were available for only 4 months, and the wet chemical samples were actually taken from the full Web test bar casting.

A value of 0.8 was obtained for the yield strength vs. the average zinc variation while values of −0.5 and

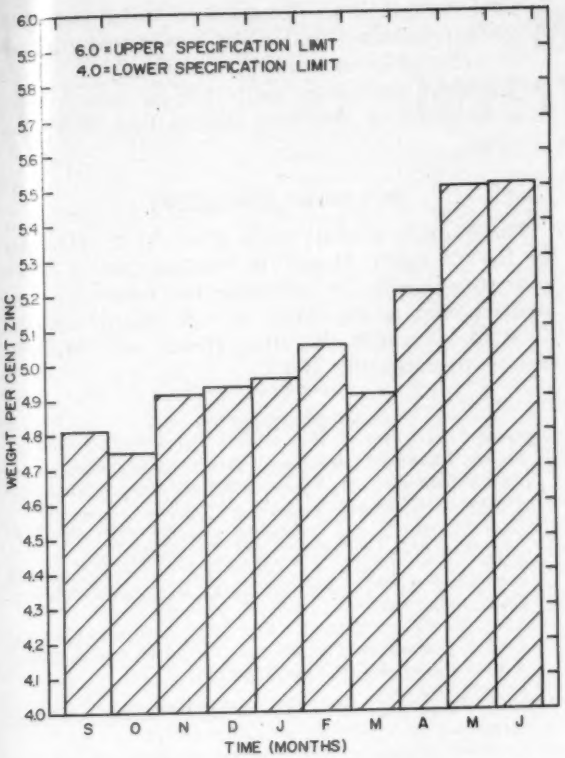



Fig. 9 — Bar graph of the average monthly zinc chemistry vs. time.

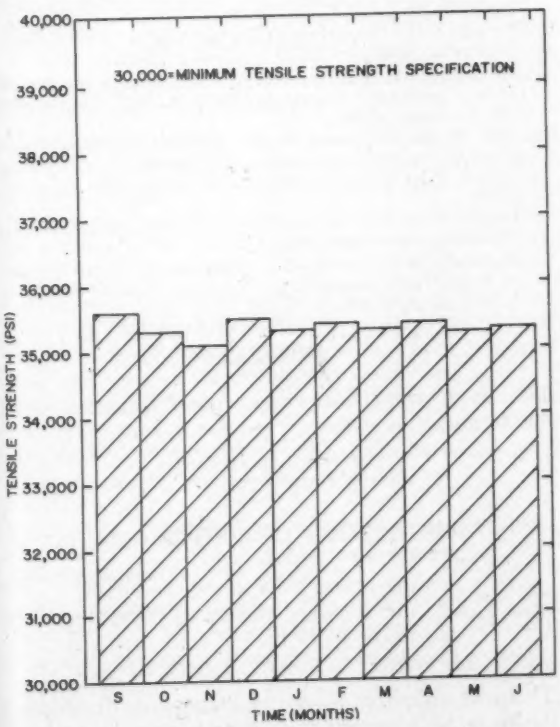



Fig. 10 — Bar graph of the average monthly tensile strength vs. time.

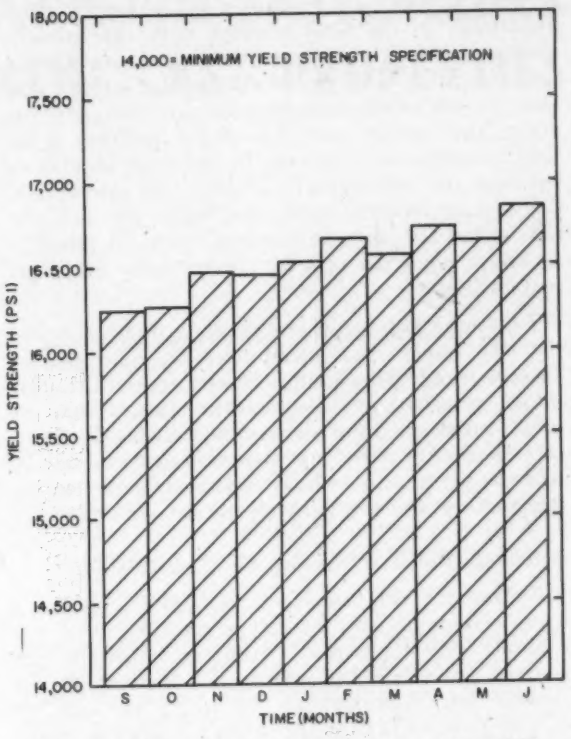



Fig. 11 — Bar graph of the average monthly yield strength vs. time.

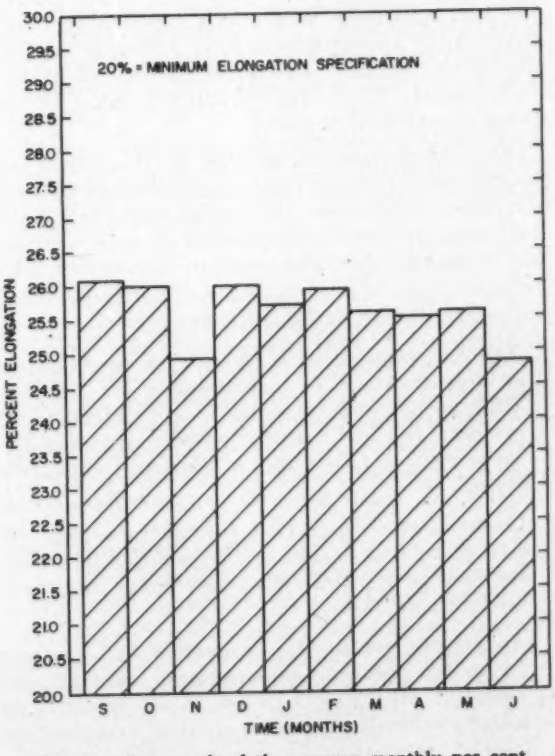



Fig. 12 — Bar graph of the average monthly per cent elongation vs. time.

−0.2 were obtained for both the tensile strength and per cent elongation, respectively. This signifies that the change in the yield strength with zinc content was significant. Throughout this last analysis, it is assumed that the other chemical variables, namely, the lead, tin and other impurities, do not change to any appreciable extent, and that their variation is insignificant and cancelled out by the large number of test bars and data taken. To check this assumption, the average monthly values for both tin and lead were calculated, and the standard deviation found to be under ±one per cent for tin and only ±0.06 per cent for lead.

Tests were also made to determine whether or not the zinc content had an appreciable effect on the fluidity of the 85-5-5-5 alloy. Several series of fluidity spirals were cast, and the results indicated that the zinc content had little effect in the allowable chemical range. The data corroborated the findings of Rosenberg, et al,[14] which indicated that pouring temperature was the most significant variable.

### *Economic Value of Raising the Zinc Content*

Raising the zinc content to a higher level has a considerable effect on the cost of the final alloy. For example, over the period of time in which the zinc content was raised, the average zinc content was increased from approximately 4.8 to 5.5 per cent, or 0.7 per cent. Since in the plant's operation about 550,000 lb of castings are made monthly, this means that an additional 3850 lb of zinc were being used monthly. Thus, assuming that the increased zinc loss is offset by the decrease in the density of the alloy and neglecting both effects, 3850 lb of a material costing approximately $0.12 a lb is being substituted for the same amount of 85-5-5-5 which cost approximately $0.30, thus effecting a material cost saving of over $700/month.

Other advantages attributable to the system are that the control is more positive and the yield strength is increased while both the tensile strength and per cent elongation exhibit no marked decrease attributable to the increased zinc content.

## CONCLUSIONS

1. The zinc composition of 85-5-5-5 can be controlled to close tolerance by compensating for zinc loss which in our case amounted to approximately 0.8 per cent, and by adjusting the composition of the input material to the level of the final composition desired.
2. Low frequency induction furnaces are ideal for controlling bronze chemistry in addition to providing low power costs, low melt losses and good working conditions.
3. An increase in the zinc content of 85-5-5-5 from 4.8 to 5.5 per cent amounts to a material cost saving of over $700 a month in an operation in which 550,000 lb of castings are produced monthly.
4. Increasing the zinc content of 85-5-5-5 from 4.8 to 5.5 per cent resulted in an increase of the average yield strength from 16,250 psi to 16,750 psi, without appreciably affecting the tensile strength, per cent elongation or fluidity.
5. The x-ray fluorescent method is an accurate and rapid means of obtaining zinc analyses in 85-5-5-5 alloy.

## ACKNOWLEDGMENT

The authors wish to thank Prof. M. C. Flemings for his comments. Messrs. D. Winslow and J. Falco were instrumental in obtaining the many chemical analyses used in this work. Mr. R. Burns and Mr. L. Babb prepared the illustrations, and Miss R. Nelson prepared the drafts.

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## APPENDIX

The linear correlation factors were obtained using the formula:

$$r = \frac{E\,(yi - \overline{yi})\,(xi - \overline{xi})}{n\,\sigma_x\,\sigma_y}$$

where $r$ is the correlation coefficient.

Xi, yi are the two variables.

$\overline{Xi}$, $\overline{yi}$ are the means of the two variables.

$\sigma_x$, $\sigma_y$, are the standard deviation of $x$ and $y$, respectively.

n is the number of pieces of data.

This will vary from −1 to +1, depending on whether the slope of a meaningful curve is positive or negative and will equal ±1 only if the points fall in a straight line.

# MAGNESIUM-BASE ALLOYS INVESTMENT CAST PROPERTIES

by K. Herrick


## ABSTRACT

**Using the investment casting process, four magnesium-base alloys were evaluated — AZ92A, AZ91C, AZ81A and AM100A. Average grain size and massive compound ratings were determined as a function of section size, investment mold temperature and metal pouring temperature. Tensile properties were found to be approximately those obtained by sand casting methods. The alloys were classified according to their tendency toward microshrinkage as observed on ten typical production parts.**

**The author concludes that magnesium alloys AZ92A, AZ91C, AZ81A and AM100A can be cast satisfactorily in a variety of shapes common to the investment casting process for nonferrous alloys. Tensile properties of the four alloys exceed the minimum requirements of Government and aeronautical specifications. AZ91C alloy had the best combination of mechanical properties and castability for use with the investment casting process.**


## INTRODUCTION

The history of magnesium investment castings dates back to the middle of late 1940's. The feasibility of casting magnesium-base alloys in investment molds has been demonstrated in previous publications.[1,2,3,4] During the early years of production, magnesium parts were used primarily for weight saving. A large percentage of the castings were furnished in the as-cast condition and did not require radiographic or fluorescent penetrant inspection. Almost all parts were produced in alloy AZ92A. As engineers became more familiar with the investment casting process, structural parts were designed.

Mechanical properties and soundness were now of prime importance. This was evidenced by the publication during 1960 of Aeronautical Material Specification 4453 which covers investment castings in alloy AZ92A, T6 temper and a tentative specification for investment castings in alloy AM100A, T6 temper.

It is well known that the properties of cast magnesium alloys are quite sensitive to the rate of freezing as well as many other factors. The purpose of this work was to determine the tensile properties of four investment cast magnesium alloys, and to investigate some of the factors which influence the investment cast properties. The alloys selected were AZ92A, AZ91C, AZ81A and AM100A.

## METHODS AND PROCEDURES

### *Mold Preparation*

All investment molds were made using the lost wax process. The wax patterns were invested with a typical gypsum investment. This material consists of approximately 30 per cent plaster of paris and 70 per cent silica refractories. Water added to make the pouring slurry was 42 per cent of the weight of the dry investment. The slurry was vacuumed prior to and after pouring the investment in the flasks in order to remove trapped air bubbles. Borofluoride compounds were added to the slurry during mixing to inhibit reaction of the molten metal with the cured mold.

Wax removal, burnout (calcining at high temperature) and oven cooling of the molds to the proper temperature for casting were accomplished in an air circulating batch-type oven. Flask temperature was controlled to ± 20 F. The entire cycle took 18-24 hr, depending on the size of the mold.

Pressure for casting was obtained by applying suction on the surface of the mold opposite the sprue opening. Molds were flushed with sulfur dioxide immediately prior to casting.

### *Metal Preparation*

One hundred lb charges of the alloys were melted in steel crucibles in a gas-fired furnace using the standard open pot melting practice. Commercially alloyed ingot was used to prepare all four alloys investigated. Chlorination for degassing and an addition of 20 grams of hexachlorabenzene at 1450 F for grain refinement were used on all alloys. The pouring temperature was adjusted in the casting ladle by chilling the metal with carbon rods.

### *Specimens and Testing*

The standard 0.252 in. diameter, one in. gage length tensile specimen was used for all tensile tests.

These cast-to-size specimens were tested in accordance with American Society for Testing Materials standards for separately cast bars. The test bars were cast vertically, four in 5 in. diameter by 7 in. mold, using the rigging shown in Fig. 1.

Average grain sizes and compound ratings were determined for all alloys in the heat treated condition

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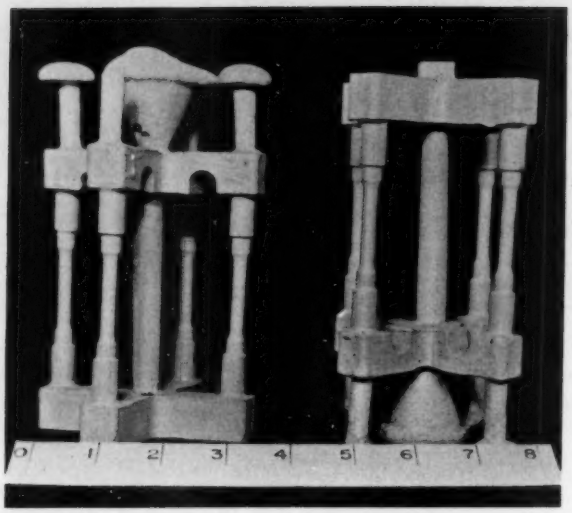


Fig. 1 — Rigging used on 0.252 in. diameter investment cast test bars.

according to the procedures outlined by George.[5] Specimens for microscopic examination were obtained from cast slugs of ⅛, ¼, ½ and ¾-in. diameter by 4-in. long. The observed area was a transverse section located approximately one in. from the end opposite the sprue opening. The rigging used for the cast slugs is shown in Fig. 2.

Radiographs were taken using a 160 kv unit which has a 2.4 mm focal spot. A fine grained industrial quality film was used. Radiographs were compared to the standards of American Society for Testing Materials Specification E98-53T.

### *Heat Treatment*

Specimens were solution heat treated in an electrically heated circulating air furnace whose atmosphere contained a minimum of 0.5 per cent sulfur dioxide. Alloy AZ92A castings were solution heat treated using the following anti-germination schedule:

1. Castings were held 6 hr at 765 ± 5 F.
2. Temperature was reduced to 665 ± 5 F and held for 2 hr.
3. Temperature was raised to 765 ± 5 F and held for 10 hr.
4. Air quench.

Alloy AZ91C, AZ81A and AM100A castings were solution heat treated using the following anti-germination schedule:

1. Castings were held 6 hr at 775 ± 5 F.
2. Temperature was reduced to 665 ± 5 F and held for 2 hr.
3. Temperature was raised to 775 ± 5 F and held for 10 hr.
4. Air quench.

Artificial aging was applied in the same furnace used for solution heat treatment.

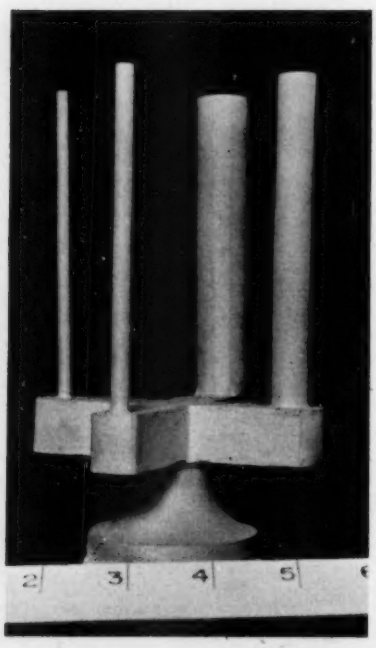


Fig. 2 — Rigging used on ⅛, ¼, ½ and ¾-in. diameter investment cast slugs.

## RESULTS

Compound ratings and average grain sizes were determined for each alloy at each of three mold temperatures, two pouring temperatures and the four section sizes previously described. The results are listed in Table 1.

Tables 2 through 5 summarize the effect of investment mold temperature and pouring temperature on the tensile properties of the alloys.

The tensile yield strength, ultimate strength and elongation of AZ92A, AZ91C and AM100A are presented as a function of aging time for a series of four aging temperatures in Figs. 3 to 14, inclusive. Investment mold temperature was 70 F and pouring temperature was 1240 F for all test bars in this aging test.

An evaluation of the castability of the four alloys was based on the visual and radiographic results of the ten different castings shown in Figs. 15 through 17. Visual appearance of all the alloys was the same. Table 6 summarizes the radiographic results of the ten castings produced in the various alloys. All castings were graded only on microshrinkage. Rigging, flask temperature and pouring temperature was held constant even though the alloy was changed or several alloys were used for the same part to compare radiographic quality.

## DISCUSSION OF RESULTS

Investment mold temperature, metal pouring temperature and section size had minor affects on the average grain size of all alloys investigated. An arbitrary dividing line set by some sand casting foundries between fine and coarse grain diameter in a 0.505

TABLE 1 — INVESTMENT MOLD TEMPERATURE, POURING TEMPERATURE AND SECTION SIZE EFFECT ON THE AVERAGE GRAIN SIZE AND COMPOUND RATING OF FOUR HEAT TREATED MAGNESIUM ALLOYS

| Alloy | Mold Temp., F | Pouring Temp., F | Avg. grain dia. at various section sizes (x 10-3 in.) | | | | Compound rating at various section sizes | | | |
|---|---|---|---|---|---|---|---|---|---|---|
| | | | 1/8 | 1/4 | 1/2 | 3/4 | 1/8 | 1/4 | 1/2 | 3/4 |
| AZ92A | 650 | 1360 | 3 | 4 | 4 | 4 | 3 | 4 | 6 | 8 |
| | 650 | 1240 | 3 | 3 | 4 | 4 | 3 | 4 | 5 | 7 |
| | 400 | 1360 | 3 | 3 | 4 | 4 | 1 | 3 | 4 | 6 |
| | 400 | 1240 | 3 | 4 | 4 | 5 | 1 | 3 | 4 | 6 |
| | 70 | 1360 | 3 | 3 | 3 | 3 | 1 | 2 | 4 | 5 |
| | 70 | 1240 | 3 | 3 | 3 | 4 | 1 | 2 | 4 | 6 |
| AZ91C | 650 | 1360 | 3 | 3 | 3 | 3 | 1 | 1 | 3 | 4 |
| | 650 | 1240 | 3 | 3 | 3 | 4 | 1 | 1 | 4 | 5 |
| | 400 | 1360 | 3 | 3 | 4 | 4 | 1 | 1 | 3 | 4 |
| | 400 | 1240 | 2 | 3 | 3 | 2 | 1 | 1 | 3 | 4 |
| | 70 | 1360 | 3 | 4 | 4 | 4 | 1 | 1 | 3 | 3 |
| | 70 | 1240 | 3 | 3 | 3 | 3 | 1 | 1 | 2 | 3 |
| AZ81A | 650 | 1360 | 3 | 3 | 3 | 4 | 1 | 2 | 3 | 5 |
| | 650 | 1240 | 3 | 3 | 3 | 3 | 1 | 2 | 2 | 4 |
| | 400 | 1360 | 3 | 3 | 3 | 3 | 1 | 1 | 2 | 4 |
| | 400 | 1240 | 3 | 4 | 3 | 3 | 1 | 1 | 2 | 3 |
| | 70 | 1360 | 3 | 3 | 3 | 3 | 1 | 1 | 2 | 3 |
| | 70 | 1240 | 3 | 3 | 3 | 3 | 1 | 1 | 2 | 3 |
| AM100A | 650 | 1360 | 3 | 3 | 3 | 4 | 1 | 2 | 6 | 7 |
| | 650 | 1240 | 3 | 4 | 4 | 3 | 1 | 1 | 5 | 7 |
| | 400 | 1360 | 3 | 3 | 4 | 4 | 1 | 2 | 5 | 7 |
| | 400 | 1240 | 3 | 4 | 4 | 4 | 1 | 2 | 5 | 6 |
| | 70 | 1360 | 4 | 3 | 4 | 4 | 1 | 2 | 5 | 6 |
| | 70 | 1240 | 2 | 2 | 3 | 4 | 1 | 1 | 5 | 6 |

Analysis of Metal

| | AZ92A | AZ91C | AZ81A | AM100A |
|---|---|---|---|---|
| Al, % | 8.95 | 8.64 | 7.87 | 10.10 |
| Zn, % | 2.10 | 0.56 | 0.94 | 0.11 |
| Mn, % | 0.20 | 0.22 | 0.24 | 0.22 |

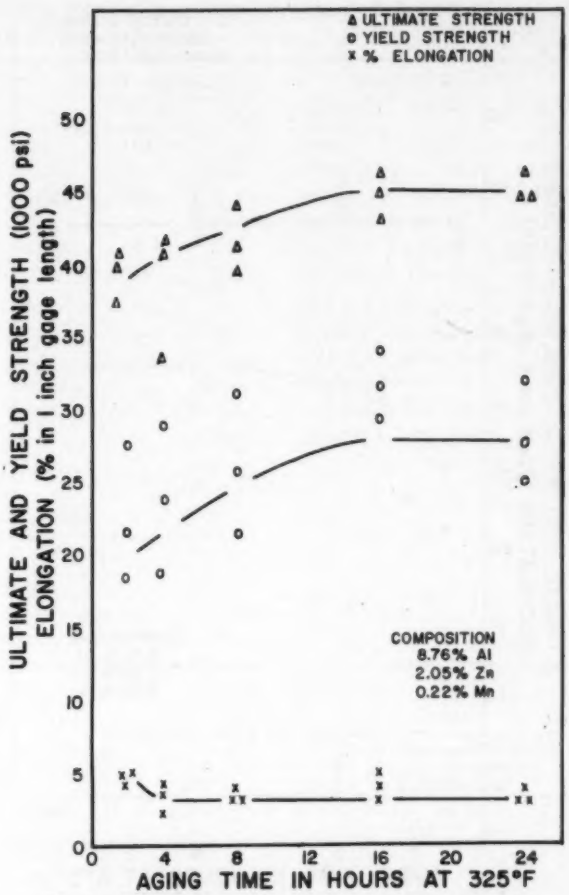



Fig. 3 — Tensile properties of investment cast AZ92A, solution treated and aged at 325 F.

TABLE 2 — INVESTMENT MOLD AND POURING TEMPERATURE EFFECT ON THE TENSILE PROPERTIES OF AZ92A

| Chemical Composition | Mold Temp., F | Pouring Temp., F | Temper | No. of Bars | Avg. Properties | | | Range of Properties | | |
|---|---|---|---|---|---|---|---|---|---|---|
| | | | | | TS, ksi | YS, ksi | E, % | TS, ksi | YS, ksi | E, % |
| B | 650 | 1360 | T6 | 4 | 28.8 | 22.0 | 1.0 | 26.1-31.9 | 20.7-23.4 | 1.0 |
| B | 650 | 1240 | T6 | 4 | 37.2 | 20.1 | 2.8 | 32.7-38.3 | 17.5-21.8 | 2.0-3.0 |
| A | 400 | 1360 | T6 | 4 | 37.6 | 24.0 | 2.4 | 36.9-38.2 | 22.5-25.5 | 2.0-3.0 |
| A | 400 | 1240 | T6 | 3 | 40.8 | 25.8 | 2.5 | 38.8-42.1 | 25.1-26.2 | 2.0-3.0 |
| A | 70 | 1360 | T6 | 4 | 35.2 | 23.2 | 2.0 | 34.1-37.5 | 21.5-26.8 | 2.0 |
| A | 70 | 1240 | T6 | 4 | 41.9 | 25.7 | 2.8 | 41.6-42.4 | 23.4-27.4 | 2.5-3.0 |
| B | 650 | 1360 | T4 | 4 | 28.7 | 13.5 | 3.3 | 28.3-29.0 | 13.4-13.7 | 3.0-4.0 |
| B | 650 | 1240 | T4 | 3 | 35.4 | 13.9 | 7.7 | 34.3-36.6 | 13.8-14.0 | 6.5-9.0 |
| A | 400 | 1360 | T4 | 3 | 35.4 | 14.8 | 6.2 | 34.4-36.4 | 14.3-15.4 | 6.0-6.5 |
| A | 400 | 1240 | T4 | 4 | 36.5 | 15.6 | 6.2 | 35.7-36.7 | 14.1-16.3 | 5.0-7.0 |
| A | 70 | 1360 | T4 | 4 | 34.4 | 16.6 | 7.5 | 34.2-34.6 | 14.3-18.2 | 7.0-8.0 |
| A | 70 | 1240 | T4 | 4 | 40.3 | 16.1 | 7.5 | 39.1-41.5 | 13.9-18.5 | 7.0-8.0 |
| A | 70 | 1240 | F | 4 | 27.4 | — | 3.0 | 26.8-27.8 | — | 2.0-4.0 |

Heat treatment: T4, 16 hr at 765 F.
T6, 16 hr at 765 F and 5 hr at 425 F.

Chemical Composition

| | A | B |
|---|---|---|
| Al, % | 8.95 | 9.60 |
| Zn, % | 2.10 | 2.16 |
| Mn, % | 0.20 | 0.22 |

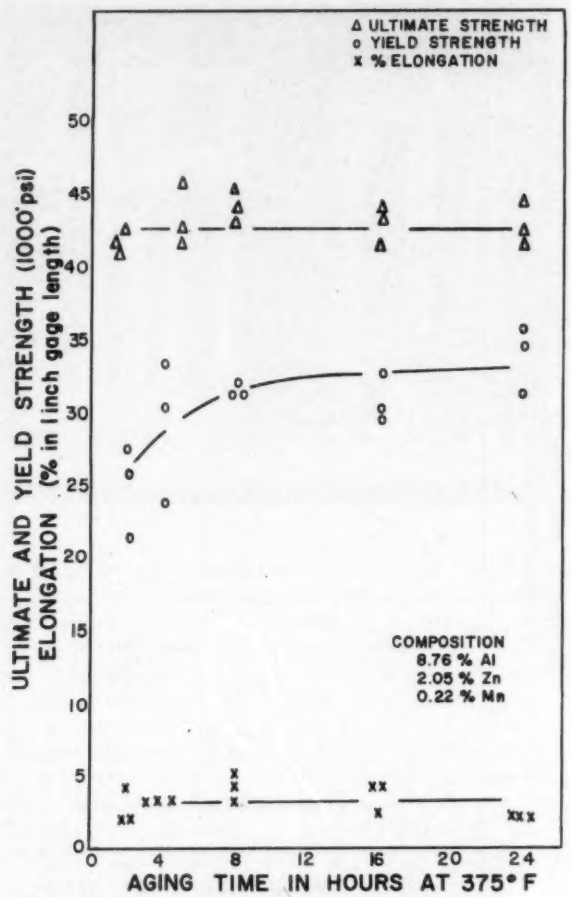



Fig. 4 — Tensile properties of investment cast AZ92A, solution treated and aged at 375 F.

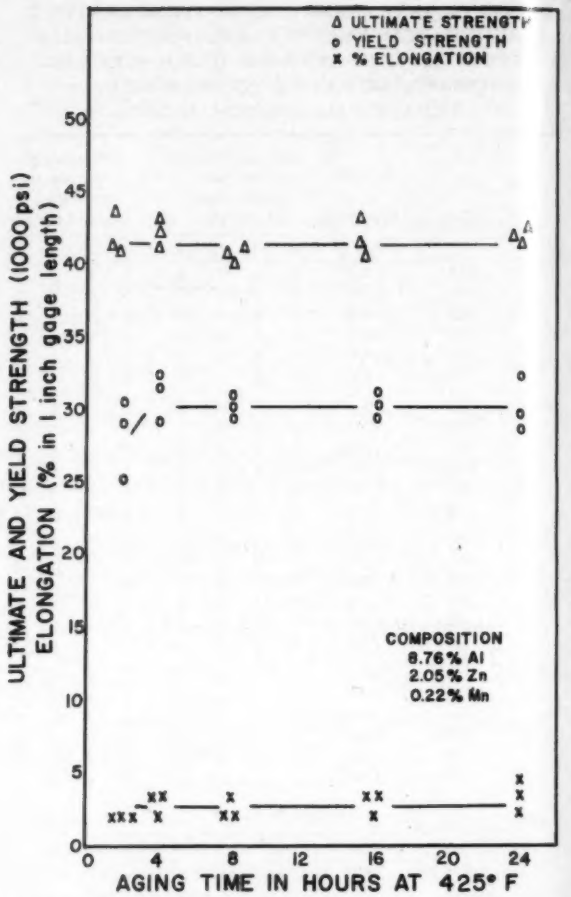



Fig. 5 — Tensile properties of investment cast AZ92A, solution treated and aged at 425 F.

TABLE 3 — INVESTMENT MOLD AND POURING TEMPERATURE EFFECT ON THE TENSILE PROPERTIES OF AZ91C

| Chemical Composition | Mold Temp., F | Pouring Temp., F | Temper | No. of Bars | Avg. Properties | | | Range of Properties | | |
|---|---|---|---|---|---|---|---|---|---|---|
| | | | | | TS, ksi | YS, ksi | E, % | TS, ksi | YS, ksi | E, % |
| A | 650 | 1360 | T6 | 4 | 38.1 | 20.9 | 4.5 | 36.7-39.8 | 18.4-24.2 | 4.0-5.0 |
| A | 650 | 1240 | T6 | 4 | 33.3 | 20.5 | 3.0 | 31.4-36.9 | 18.2-24.4 | 2.0-4.0 |
| A | 400 | 1240 | T6 | 4 | 35.9 | 20.0 | 3.0 | 34.9-37.0 | 17.4-21.8 | 2.0-4.0 |
| A | 70 | 1360 | T6 | 3 | 35.7 | 23.6 | 3.0 | 34.2-37.1 | 20.0-26.2 | 2.0-4.0 |
| A | 70 | 1240 | T6 | 3 | 36.0 | 22.7 | 3.6 | 35.6-36.2 | 19.1-26.3 | 3.0-4.0 |
| A | 650 | 1360 | T4 | 4 | 35.4 | 13.1 | 9.0 | 35.3-35.5 | 11.8-14.4 | 8.0-10.0 |
| A | 650 | 1240 | T4 | 4 | 35.0 | 11.7 | 9.0 | 34.8-35.2 | 10.5-12.9 | 9.0 |
| A | 400 | 1360 | T4 | 4 | 35.6 | 13.9 | 8.0 | 34.9-36.4 | 13.8-14.1 | 7.0-9.0 |
| A | 400 | 1240 | T4 | 4 | 34.9 | 14.1 | 9.0 | 34.2-35.8 | 13.9-14.4 | 9.0 |
| A | 70 | 1360 | T4 | 3 | 37.3 | 14.4 | 10.3 | 36.7-38.0 | 13.5-15.3 | 10.0-11.0 |
| A | 70 | 1240 | T4 | 3 | 37.7 | 14.1 | 11.6 | 35.8-38.9 | 13.9-14.4 | 11.0-13.0 |
| A | 70 | 1240 | F | 4 | 28.7 | — | 2.5 | 27.7-29.2 | — | 2.0-3.0 |

Heat treatment: T4, 16 hr at 775 F.
T6, 16 hr at 775 F and 16 hr at 335 F.

| Chemical Composition | |
|---|---|
| | A |
| Al, % | 8.64 |
| Zn, % | 0.56 |
| Mn, % | 0.22 |

TABLE 4 — INVESTMENT MOLD AND POURING TEMPERATURE EFFECT ON THE TENSILE PROPERTIES OF AZ81A

| Chemical Composition | Mold Temp., F | Pouring Temp., F | Temper | No. of Bars | Avg. Properties | | | Range of Properties | | |
|---|---|---|---|---|---|---|---|---|---|---|
| | | | | | TS, ksi | YS, ksi | E, % | TS, ksi | YS, ksi | E, % |
| A | 650 | 1360 | T4 | 9 | 33.6 | 15.4 | 9.4 | 30.7-35.7 | 11.2-17.9 | 9.0-15.0 |
| A | 650 | 1240 | T4 | 7 | 34.4 | 11.6 | 10.8 | 33.2-35.8 | 10.8-12.1 | 9.0-14.0 |
| A | 400 | 1360 | T4 | 4 | 36.5 | 13.2 | 12.7 | 36.3-36.8 | 12.8-14.0 | 12.0-13.0 |
| A | 400 | 1240 | T4 | 4 | 35.5 | 13.3 | 11.5 | 34.6-36.1 | 12.5-15.3 | 11.0-12.0 |
| A | 70 | 1360 | T4 | 5 | 36.5 | 14.5 | 14.0 | 35.9-36.8 | 13.2-15.7 | 13.0-15.0 |
| A | 70 | 1240 | T4 | 5 | 36.1 | 13.5 | 12.0 | 35.1-36.7 | 11.7-15.0 | 11.0-14.0 |
| B | 650 | 1360 | F | 4 | 26.3 | — | 2.1 | 25.4-27.4 | — | 1.5-3.0 |
| B | 650 | 1240 | F | 4 | 28.5 | — | 3.1 | 27.2-29.6 | — | 2.5-4.0 |
| B | 400 | 1360 | F | 4 | 23.3 | — | 1.6 | 22.1-24.3 | — | 1.0-2.0 |
| B | 400 | 1240 | F | 4 | 32.1 | — | 5.0 | 31.4-32.6 | — | 4.5-5.5 |
| B | 70 | 1360 | F | 4 | 28.3 | — | 4.0 | 25.7-30.0 | — | 2.0-5.0 |
| B | 70 | 1240 | F | 4 | 31.3 | — | 4.9 | 29.7-33.0 | — | 4.0-5.5 |

Heat treatment: T4, 16 hr at 775 F.

| Chemical Composition | | |
|---|---|---|
| | A | B |
| Al, % | 7.70 | 7.76 |
| Zn, % | 0.65 | 0.80 |
| Mn, % | 0.24 | 0.24 |

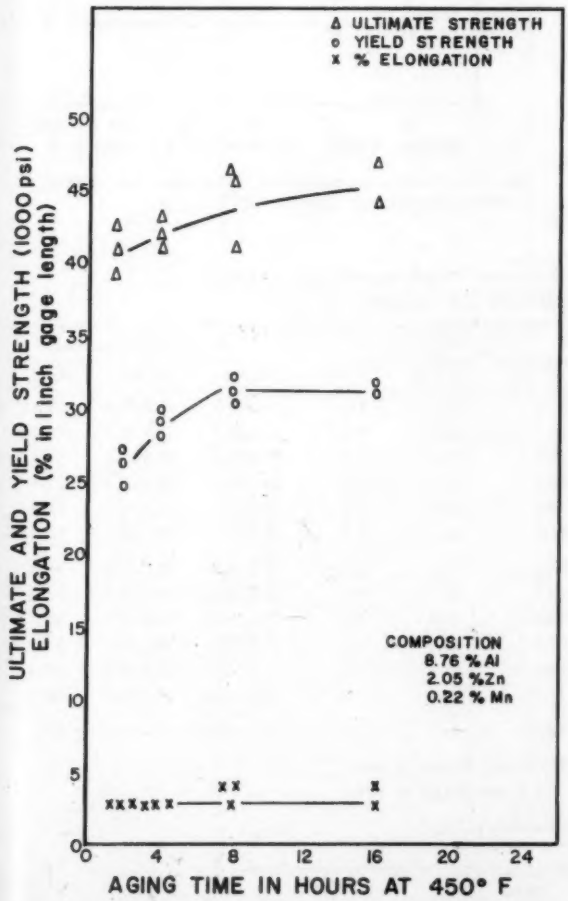



Fig. 6 — Tensile properties of investment cast AZ92A, solution treated and aged at 450 F.

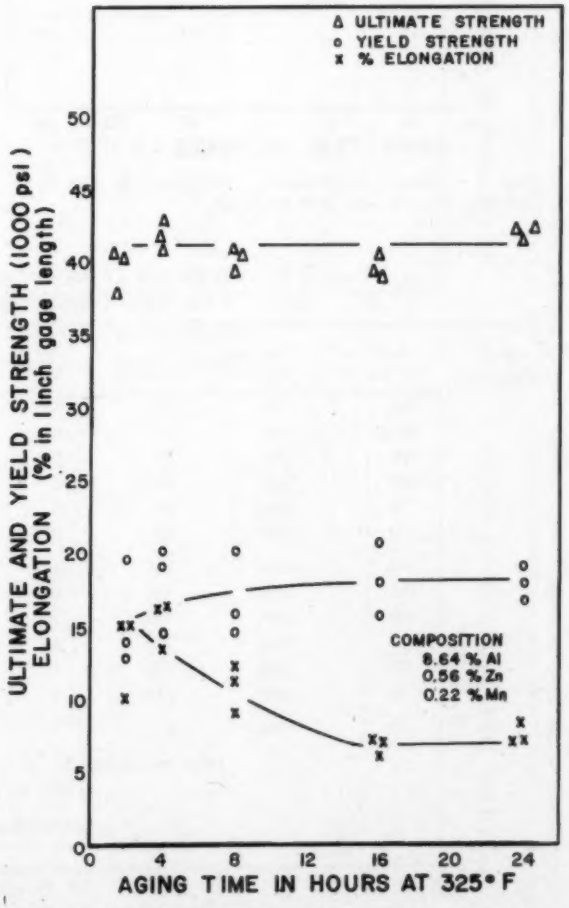



Fig. 7 — Tensile properties of investment cast AZ91C, solution treated and aged at 325 F.

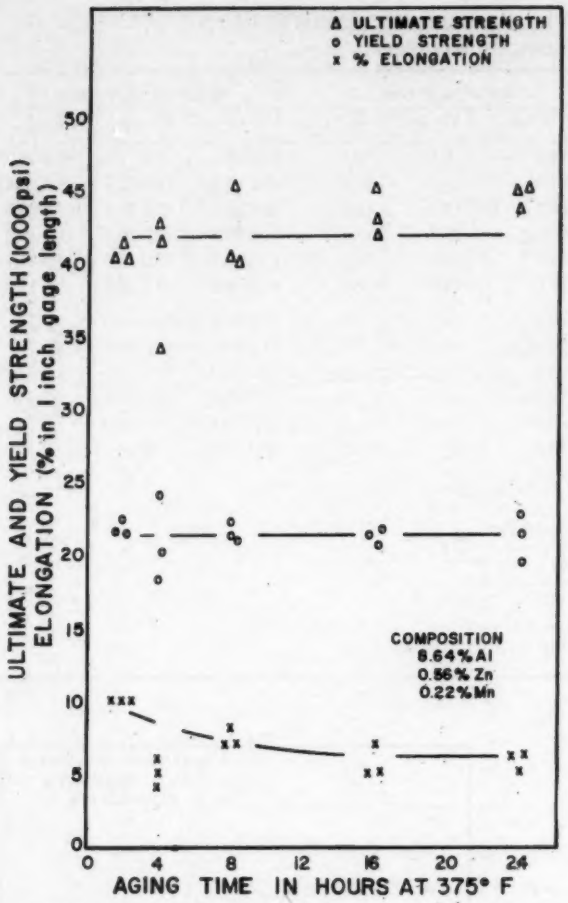



Fig. 8 — Tensile properties of investment cast AZ91C, solution treated and aged at 375 F.

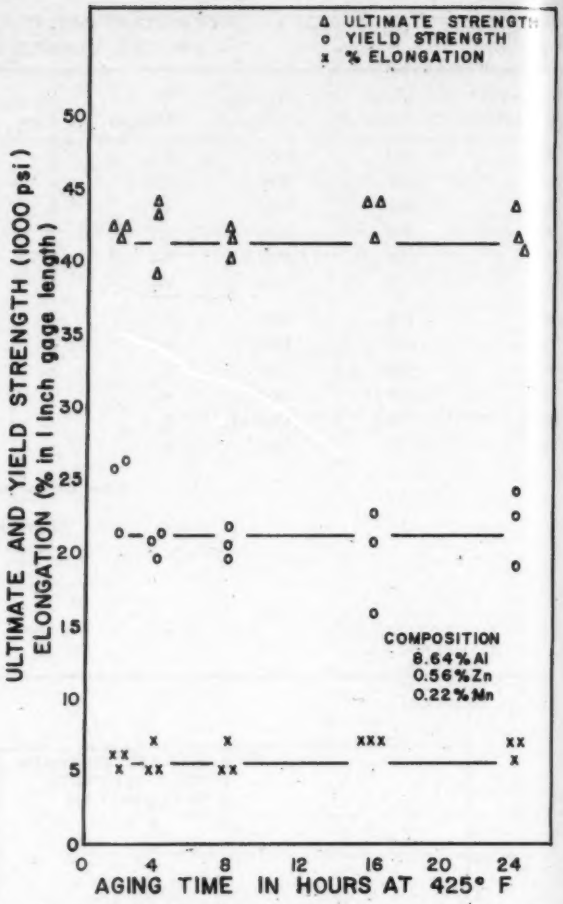



Fig. 9 — Tensile properties of investment cast AZ91C, solution treated and aged at 425 F.

TABLE 5 — INVESTMENT MOLD AND POURING TEMPERATURE EFFECT ON THE TENSILE PROPERTIES OF AM100A

| Chemical Composition | Mold Temp., F | Pouring Temp., F | Temper | No. of Bars | Avg. Properties | | | Range of Properties | | |
|---|---|---|---|---|---|---|---|---|---|---|
| | | | | | TS, ksi | YS, ksi | E, % | TS, ksi | YS, ksi | E, % |
| A | 650 | 1360 | T6 | 5 | 27.8 | 19.0 | 3.0 | 24.3-29.5 | 15.9-20.9 | 2.0-4.0 |
| A | 650 | 1240 | T6 | 6 | 34.8 | 20.3 | 4.3 | 31.5-36.1 | 19.1-21.1 | 4.0-5.0 |
| A | 400 | 1360 | T6 | 6 | 30.3 | 27.6 | 3.3 | 29.2-31.6 | 26.6-29.2 | 3.0-5.0 |
| A | 400 | 1240 | T6 | 6 | 30.3 | 25.7 | 2.7 | 28.5-31.7 | 23.1-28.4 | 2.0-3.0 |
| A | 70 | 1360 | T6 | 6 | 34.0 | 27.9 | 3.3 | 30.7-34.9 | 23.6-30.7 | 1.0-5.0 |
| A | 70 | 1240 | T6 | 6 | 35.5 | 28.2 | 3.7 | 31.5-37.6 | 26.1-30.6 | 3.0-4.0 |
| A | 650 | 1360 | T61 | 6 | 28.8 | 22.4 | 3.7 | 27.2-30.0 | 17.7-26.4 | 3.0-4.0 |
| A | 650 | 1240 | T61 | 6 | 34.5 | 27.1 | 3.7 | 33.6-36.1 | 21.7-30.3 | 3.0-4.0 |
| A | 400 | 1360 | T61 | 4 | 30.0 | 25.6 | 2.0 | 29.5-31.0 | 24.0-27.4 | 1.0-3.0 |
| A | 400 | 1240 | T61 | 6 | 33.5 | 26.9 | 4.0 | 31.2-35.4 | 24.2-28.7 | 3.0-5.0 |
| A | 70 | 1360 | T61 | 6 | 33.4 | 27.8 | 5.3 | 31.3-35.2 | 25.1-30.9 | 4.0-7.0 |
| A | 70 | 1240 | T61 | 6 | 36.7 | 25.9 | 4.5 | 33.0-36.8 | 21.8-29.7 | 3.0-7.0 |
| A | 70 | 1240 | F | 4 | 28.2 | — | 3.0 | 26.7-20.3 | — | 3.0 |

Heat treatment: T6, 16 hr at 775 F and 10 hrs at 400 F.
T61, 16 hr at 775 F and 24 hr at 400 F.

| Chemical Composition | |
|---|---|
| | A |
| Al, % | 10.00 |
| Zn, % | 0.06 |
| Mn, % | 0.24 |

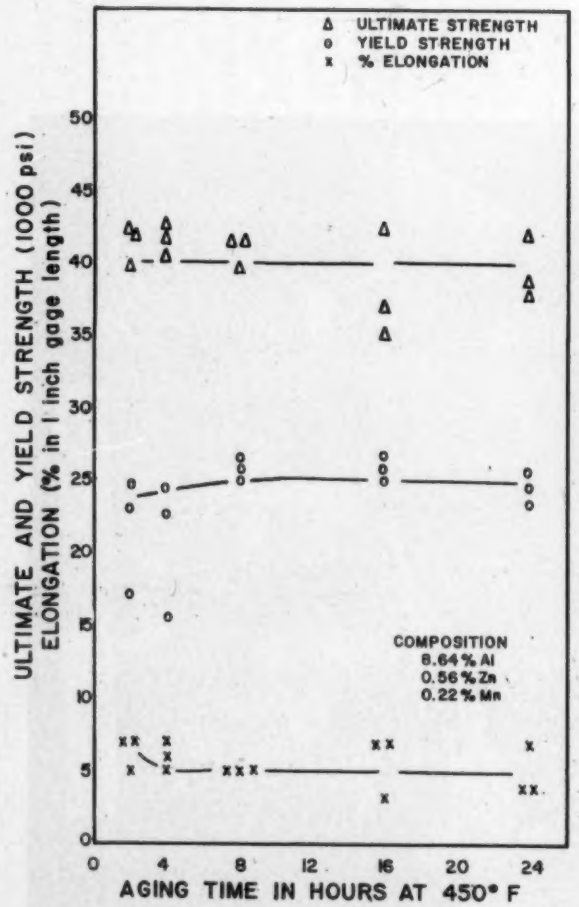



Fig. 10 — Tensile properties of investment cast AZ91C, solution treated and aged at 475 F.

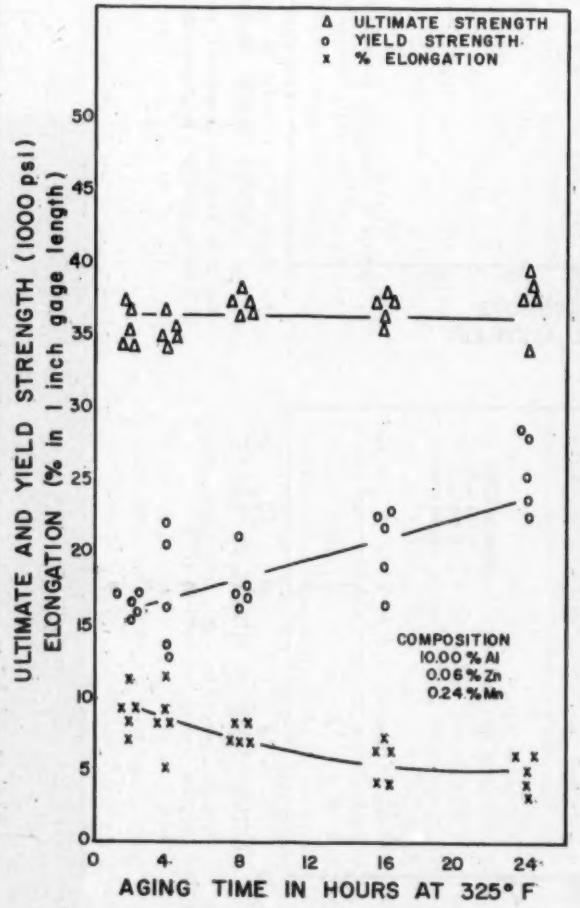



Fig. 11 — Tensile properties of investment cast AM100A, solution treated and aged at 325 F.

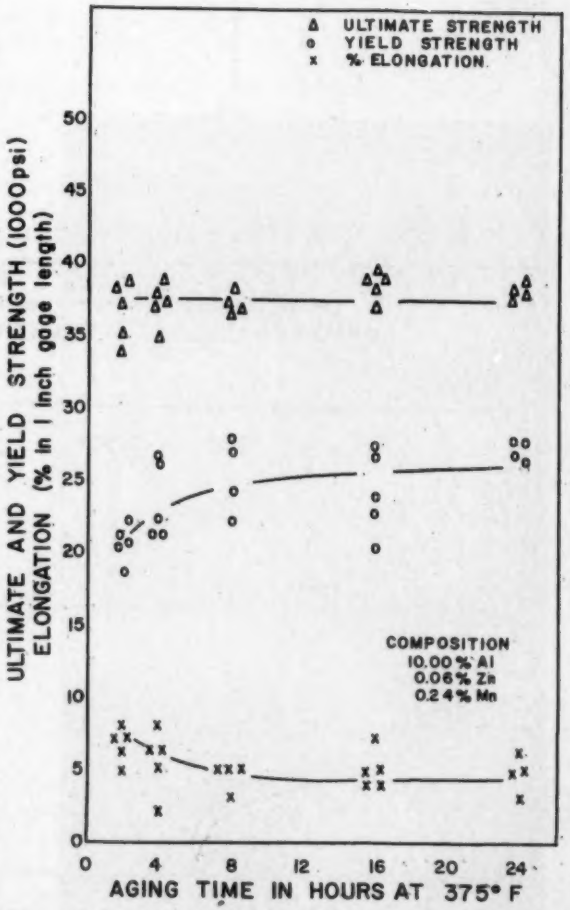



Fig. 12 — Tensile properties of investment cast AM100A, solution treated and aged at 375 F.

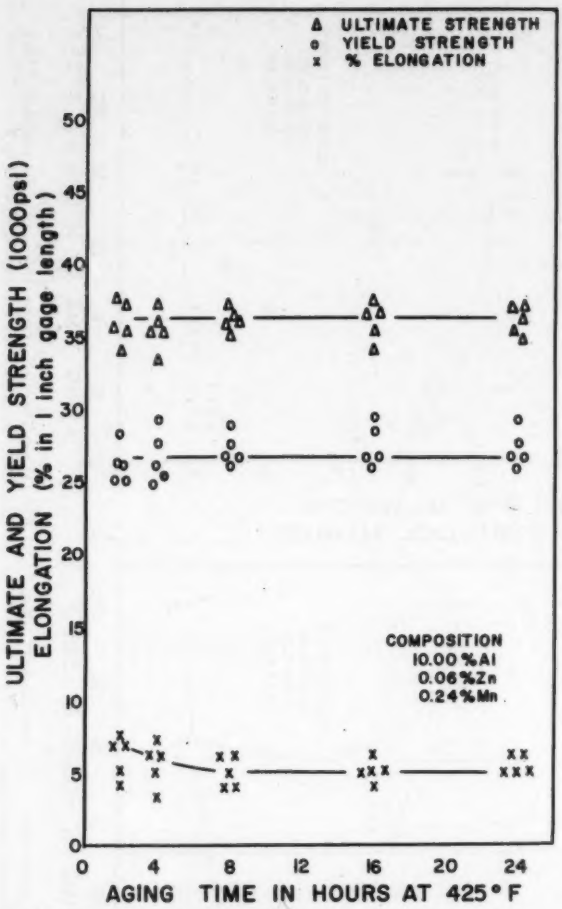



Fig. 13 — Tensile properties of investment cast AM100A, solution treated and aged at 425 F.

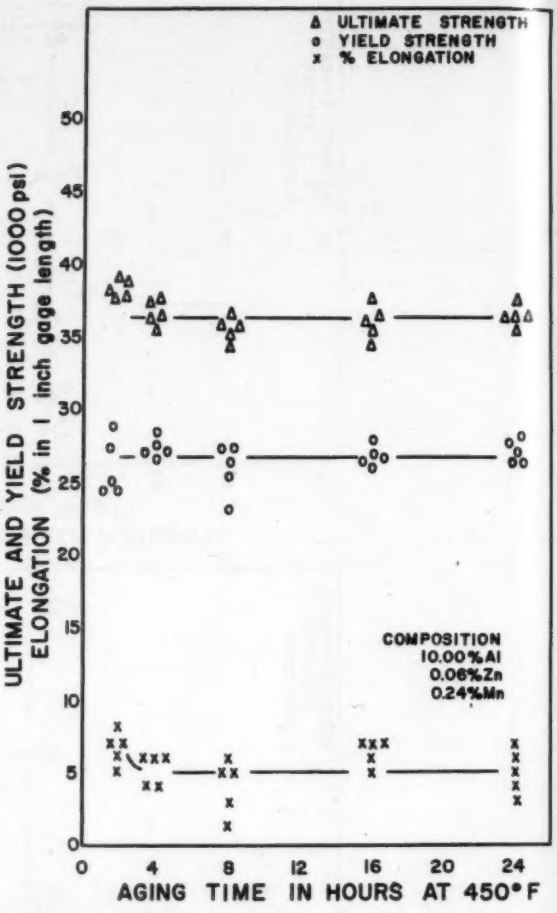



Fig. 14 — Tensile properties of investment cast AM100A, solution treated and aged at 450 F.

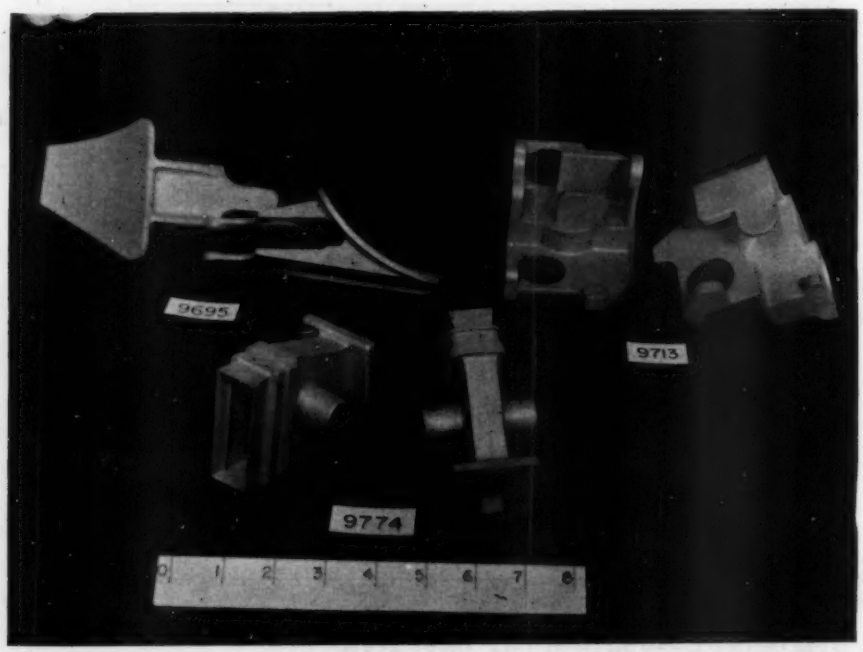



Fig. 15 — Investment cast magnesium alloy parts 9774, 9713 and 9695.

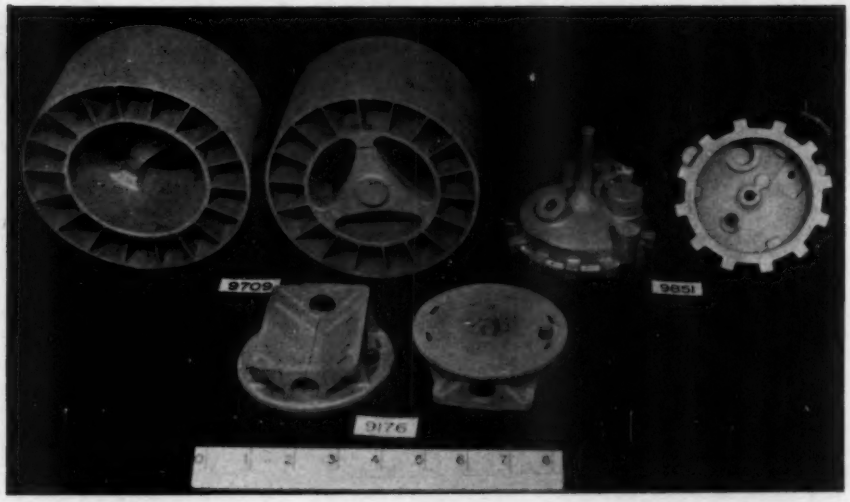



Fig. 16 — Investment cast magnesium alloy parts 9176, 9709 and 9851.

in. diameter test bar is 0.007 in. The observed level of 0.002 to 0.005 in. average grain diameter would be considered fine grained. This level of grain size was maintained with all solidification conditions studied. The amount of massive compound retained after solution treatment varied considerably depending on the alloy, casting conditions and section size of the specimen.

Increasing the section size resulted in greater increases in massive compound than did changes in investment mold temperature or metal pouring temperature. As might be expected, the least amount of massive compound for all alloys was obtained when the investment mold temperature and metal pouring temperature were low. The least amount of massive compound for all experimental conditions was obtained from alloy AZ81A followed by AZ91C, AM100A and AZ92A, respectively.

Increasing the investment mold temperature from 70 to 650 F and the pouring temperature from 1240 F to 1360 F had a deteriorating effect on the average mechanical properties of all alloys and tempers investigated. This effect was the most serious on alloys AZ92A and AM100A, and was of little, if any, consequence on alloys AZ91C and AZ81A. All test bars were examined radiographically after heat treatment and prior to tensile testing. Test bars with gross defects were discarded prior to testing. Radiography revealed no differences in soundness as a function of casting conditions for any of the alloys.

Variations in property levels as a result of changes made in casting conditions were a function of undissolved compound after heat treatment. The largest decrease in properties occurred when the pouring temperature was increased from 1240 F to 1360 F using a 650 F mold temperature. The microstructure of the heat treated test bars with low properties contained a considerably greater amount of undissolved compound as compared to the test bars with higher properties. Metallographic examination was performed on selected test bars of each alloy representing the lowest and the highest properties obtained. This examination revealed that all structures were fine grained and free of porosity voids, but that the massive compound varied considerably.

The massive compound rating of AM100A test bars with low mechanical properties was five to six, as compared to two to three for test bars with the higher mechanical properties. Massive compound rating of AZ92A test bars with low mechanical properties was seven to eight as compared to three to four for test bars with the higher mechanical properties. No effort was made during the course of this work to improve on this condition through changes in heat treating cycles or other special procedures, as this would distract from the primary objective of direct comparison of the four alloys and the casting variables involved.

Optimum properties of AZ92A and AM100A were obtained after solution treatment and aging 4 hr at 425 F. Optimum properties of AZ91C were obtained after solution treatment and aging 8 hr at 450 F. Properties corresponding to this treatment were:

| | Typical Tensile Properties | | |
|---|---|---|---|
| Alloy | Ultimate, psi | Yield, psi | % Elong. |
| AZ92A | 40,000 | 28,000 | 2.0 |
| AZ91C | 40,000 | 25,000 | 5.0 |
| AM100A | 35,000 | 25,000 | 5.0 |

The ten castings selected for radiographic comparison using the four different alloys represent a cross-section of the type of castings currently being produced in production. It should be noted that in all cases, except part 9709, AZ92A had a greater tendency toward microshrinkage than any of the other three alloys. It would be incorrect to assume that most AZ92A castings would normally fall in the category of number two to number four microshrinkage, as implied from Table 6. Modifications in the rigging and casting conditions would improve the quality of the parts cast in AZ92A.

However, these modifications would not be necessary for alloys AZ91C, AZ81A and AM100A. As a

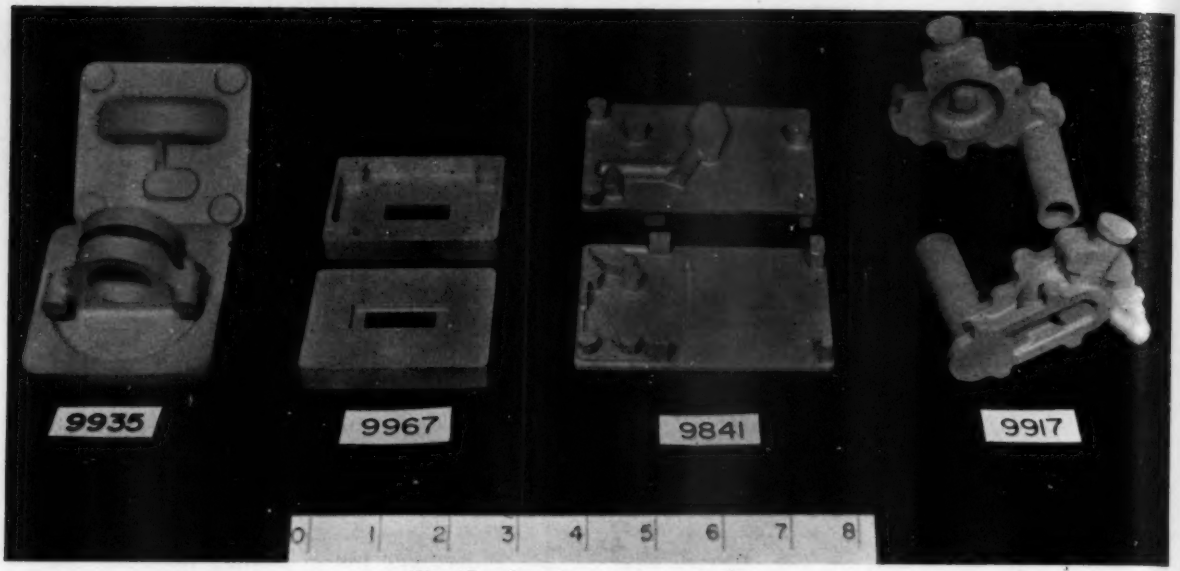



**Fig. 17 — Investment cast magnesium alloy parts 9935, 9967, 9841 and 9917.**

**TABLE 6 — RADIOGRAPHIC RESULTS OF TEN PRODUCTION PARTS GRADED ACCORDING TO MAGNESIUM ALLOY USED AND MICROSHRINKAGE LEVEL**

| Part No. | Pieces x-rayed | No. of Heats | Alloy | Number of parts with degree of microshrinkage per ASTM E98-53T, Par. 2.3, No. 1 through 6 | | | | | | |
|---|---|---|---|---|---|---|---|---|---|---|
| | | | | None | 1 | 2 | 3 | 4 | 5 | 6 |
| 9774 | 4 | 1 | AZ81A | 0 | 4 | 0 | 0 | 0 | 0 | 0 |
| 9774 | 4 | 1 | AZ91C | 0 | 4 | 0 | 0 | 0 | 0 | 0 |
| 9774 | 4 | 1 | AM100A | 3 | 1 | 0 | 0 | 0 | 0 | 0 |
| 9774 | 38 | 4 | AZ92A | 0 | 0 | 15 | 23 | 0 | 0 | 0 |
| 9695 | 4 | 1 | AZ81A | 0 | 1 | 3 | 0 | 0 | 0 | 0 |
| 9695 | 4 | 1 | AZ91C | 3 | 1 | 0 | 0 | 0 | 0 | 0 |
| 9695 | 3 | 1 | AM100A | 2 | 1 | 0 | 0 | 0 | 0 | 0 |
| 9695 | 36 | 2 | AZ92A | 0 | 6 | 30 | 0 | 0 | 0 | 0 |
| 9709 | 3 | 1 | AZ81A | 0 | 3 | 0 | 0 | 0 | 0 | 0 |
| 9709 | 3 | 1 | AZ91C | 0 | 3 | 0 | 0 | 0 | 0 | 0 |
| 9709 | 3 | 1 | AM100A | 0 | 3 | 0 | 0 | 0 | 0 | 0 |
| 9709 | 24 | 6 | AZ92A | 0 | 24 | 0 | 0 | 0 | 0 | 0 |
| 9935 | 6 | 1 | AM100A | 0 | 6 | 0 | 0 | 0 | 0 | 0 |
| 9935 | 68 | 4 | AZ92A | 0 | 0 | 27 | 41 | 0 | 0 | 0 |
| 9713 | 8 | 2 | AM100A | 0 | 8 | 0 | 0 | 0 | 0 | 0 |
| 9713 | 119 | 15 | AZ92A | 0 | 0 | 0 | 119 | 0 | 0 | 0 |
| 9851 | 4 | 1 | AM100A | 0 | 4 | 0 | 0 | 0 | 0 | 0 |
| 9851 | 6 | 2 | AZ92A | 0 | 0 | 0 | 6 | 0 | 0 | 0 |
| 9841 | 8 | 2 | AM100A | 3 | 5 | 0 | 0 | 0 | 0 | 0 |
| 9841 | 604 | 38 | AZ92A | 80 | 23 | 450 | 51 | 0 | 0 | 0 |
| 9917 | 4 | 1 | AM100A | 0 | 0 | 4 | 0 | 0 | 0 | 0 |
| 9917 | 92 | 17 | AZ92A | 0 | 0 | 10 | 21 | 61 | 0 | 0 |
| 9176 | 44 | 4 | AZ91C | 8 | 32 | 4 | 0 | 0 | 0 | 0 |
| 9176 | 6 | 2 | AZ92A | 0 | 0 | 0 | 4 | 2 | 0 | 0 |
| 9967 | 45 | 2 | AZ91C | 12 | 22 | 11 | 0 | 0 | 0 | 0 |
| 9967 | 607 | 11 | AZ92A | 0 | 0 | 316 | 81 | 210 | 0 | 0 |

result of careful review of all radiograph film, alloy AM100A appeared to have the least tendency toward microshrinkage. Alloys AZ91C and AZ81A occupy an intermediate category followed by AZ92A which had the greatest tendency toward microshrinkage. This classification is in agreement with work done on sand castings by Hanawalt, Nelson and Busk.[6]

## CONCLUSIONS

On the basis of the experimental evidence it is concluded that:

1. Magnesium alloys AZ92A, AZ91C, AZ81A and AM-100A can be cast satifactorily in a variety of shapes common to the investment casting process for other nonferrous alloys.
2. Tensile properties of all four alloys exceed the minimum requirements of applicable Government and Aeronautical Material Specifications.
3. Of the four alloys studied, AZ91C had the best combination of mechanical properties and castability for use with the investment casting process.

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6. J. D. Hanawalt, C. E. Nelson and R. S. Busk, "Properties and Characteristics of Common Magnesium Alloys," AFS TRANSACTIONS, *vol. 53*, p. 77 (1945).

# THE CARBON DIOXIDE PROCESS

by G. D. Haley and J. L. Leach


## ABSTRACT

**The purpose of this investigation was to determine the interrelated effect of the major variables in the carbon dioxide process. The variables that must be considered are gassing time, gassing pressure, grain fineness of the sand and amount of binder. An important part of this investigation has been the proposal of a standard method of testing the carbon dioxide hardened sands.**


## EXPERIMENTAL PROCEDURE

### *Standard Test*

The first step in the investigation was the selection of a method of testing the specimens during the course of study. While many types of testing machines were available in this laboratory, the selection of a machine found in most foundries posed a difficult problem. After selecting the specific testing machine, it was decided to investigate the standard tension and compression tests that were specified for use with it.

The carbon dioxide hardened tensile specimens broke at low strengths, and some broke while loading the specimen into the machine. The specimens proved strong in compression, and most of them were beyond the range of the machine. The material to be tested was therefore weak in tension and extremely strong in compression. It was decided to modify the compression test. By reducing the diameter of the specimen to one in., four times the stress could be applied to the specimen.

Thus, almost 1200 psi could be applied to the specimen. This was found sufficient to break all of the cores except the strongest. Tests indicated that testing a one in. diameter specimen in compression would satisfy the requirement of consistency and reproducibility.

### *Major Variables Effect*

The effect of four major variables on the properties of carbon dioxide cores were investigated. These variables were gassing time, gassing pressure, per cent binder and sand fineness. The standard test, described in the preceeding section, was used to study these effects.

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Gas pressure was investigated by gassing specimens with different pressures from 4 to 55 psi. Five different sands were tested, which ranged from 38 to 126 Grain Fineness number. The gassing times investigated were 30, 60 and 120 sec. Three binder contents were investigated for each sand – 2, 3 and 4 per cent. The binder which was used was a commercial grade suitable for molds containing no sugar and is available on the open market.

The grain size of each sand was checked by the standard method when it was received from the supplier, and the Grain Fineness number was calculated. In all cases at least 75 per cent of the sample weight accumulated on 3 consecutive screens. The sands were not dried prior to mixing. The $CO_2$ gas was of a commercial grade and was not dried prior to use.

## APPARATUS

The special equipment used in these studies is pictured in Figs. 1-2. A view of the apparatus for gassing the specimens is shown in Fig. 1. Gas from the tank is released through two pressure reduction valves into a ¼-in. diameter tube, then into a warming coil. From the coil the gas flows to the cut-off valve on top, which is controlled by the timer.

After leaving the valve the gas enters the top of the die holder, passes through the specimen and exits below the die. During gassing, the moveable head, shown in Fig. 1 in the down position, holds the die in place to insure all of the gas passing through the specimen. After gassing, the head is raised to allow the removal of the die. The head is moved by an air cylinder which is controlled by a switch on the front of the machine. Figure 2 shows the testing machine with the adapters to hold the one in. diameter specimen used in this investigation in place.

## DISCUSSION

In Figs. 3-7, an attempt has been made to graphically show the interrelationships of the variables studied on the strength of hardened cores.

In Fig. 3, the maximum compressive strength is plotted against the mesh size of the sand for each of the binder contents investigated. The values of compressive strength plotted were the maximum strengths obtained for the given sand and binder without regard to the time and pressure of gassing. The curves

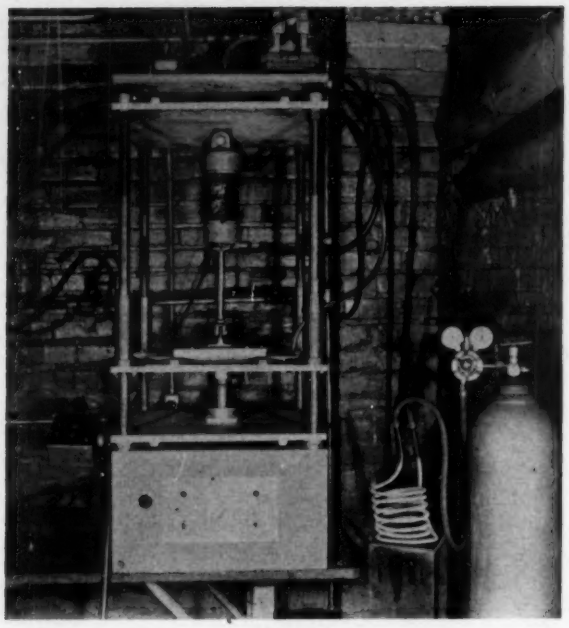


Fig. 1 — Apparatus used for gassing the specimens. Gas from tank is released through two pressure reduction valves into a ¼-in. diameter tube, then into a warming coil. From the coil the gas flows to the cut-off valve on top, which is controlled by the timer.

thus show only the effect of amount of binder and mesh size of the sand.

It is immediately apparent from Fig. 3 that an increase in binder content increases the compressive strength of the hardened cores. This effect appears greater when increasing the per cent of binder from 2 to 3 per cent for all of the sands tested except the finest one. The finest sand requires 4 per cent binder before there is a great change in compressive strength. It is believed that 2 per cent binder forms a thin film around the sand grains, and in many places the continuity of the film may be broken causing low compressive strength.

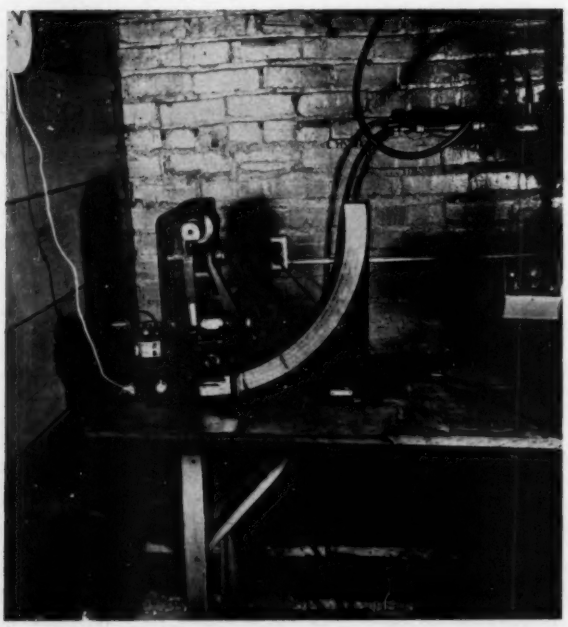


Fig. 2 — Testing machine used on the specimens. Adapters were used to hold the one in. specimen in place.

Additional binder (3 per cent) gives a thicker film, and apparently covers all of the grains. The addition of still more binder only gives a thicker film, and the increase in strength is not as great. The finest sand requires 4 per cent binder before the film is continuous, so there is not any appreciable rise in strength until 4 per cent binder is used.

Another point of interest in Fig. 3 is the effect of grain size on strength of the cores at a constant per cent of binder. It is noted that there is a rise in strength as the grain fineness of the sand is increased until the finer sands are tested. The drop associated with 126 mesh sand is greatest for the lowest per cent binder, and the effect decreases as the per cent increases. This would indicate that, if additional higher percents of binder were tested, there would not be a drop. The strength increases with finer sands because there is more bonding surface present in the finer sands.

It is proposed that the increase in strength is counteracted at 126 mesh by the inability of the binder to form a continuous bond, and hence the strength drops. At 4 per cent binder, there is almost enough binder to form a continuous film and the drop in strength is not as great.

### *Maximum Strength vs. Mesh Size*

In Fig. 4, pressures necessary to produce maximum strength at each gassing time are plotted as a function of the mesh size of the sand. Only the pressures necessary to produce optimum strength at 4 per cent binder are shown. The curves for 2 and 3 per cent binder show the same trend, only to a lesser extent, and are not plotted. Figure 4 shows that shorter gassing times require higher gassing pressures. Using 126 mesh sand and 60 sec gassing time at 15 psi gas pressure will give the maximum strength possible with this gassing time. Shorter gassing times (30 sec) require 50 psi gas pressure, which is more than three times the gas pressure times waste gas.

In addition, the curves show that higher pressures are required to gas fully the finer mesh sands. This effect is greatest for the short gassing time. This is probably due to the finer sands offering more resist-

ance to the passage of gas. Unfortunately, no equipment was available to measure the volume of gas passing through the specimen, but observation was made of the flow rate at higher pressures.

Figure 5 is the same type of plot as Fig. 4, except that time is held constant and the three binder contents are plotted, rather than holding the binder constant and plotting the three different times. The curves now show the pressure necessary to produce maximum strength at different per cent of binder using constant time. All three times produce the same type of curves, but the effect of a 30 sec time is the greatest and is the one plotted.

Figure 5 shows that higher per cent of binder require higher gassing pressures. Since the gassing time is constant, the amount of gas required by the higher binder contents is greater. Again, as in Fig. 4, it is demonstrated that finer sands require higher gassing pressures.

### *Gassing Time and Pressure Relationship*

Figures 4 and 5 show a close relationship between gassing time and pressure. The product of these two variables is a rough approximation of the amount of gas passed through the specimens. Figure 6 is a plot of compressive strength versus this product of gas pressure and time. One sand is considered at different binder contents. A great number of points are plotted, and they show a considerable spread of values for compressive strength. The extreme points are connected so they encompass all of the points for each binder content. Only the sand with mesh size 109 is shown, but the relations hold for all the sands.

Figure 6 shows that the higher binder contents give higher ranges of compressive strength. This was pointed out in Fig. 3, and is further supported by Fig. 6. The peak strength for higher binder contents occurs with larger amounts of gas. This supports Fig. 5, which shows that to harden higher percentages of binder additional gas is needed. Figure 6 shows that maximum strengths are obtained with a relatively small amount of gas compared to the total range of amounts tested.

It was found that the high pressures and short times gave lower values of compressive strengths than long times and low pressures, even though the product was the same in both cases. The decrease in strength at the higher amounts of gas is due to the overgassing effect, and also due to the formation of channels in the specimens caused by the gas passing through at the high velocity associated with high pressure.

It is possible to draw an average curve through one set of points in Fig. 6. This was done for all of the sands and all of the binder contents. Figure 7 shows the curves for all of the sands tested containing 4 per cent binder. It shows that the maximum strength occurs at increasingly higher binder contents as the fineness of the sand increases. All of the maximums of the average curves fall within a narrow range. None of the sands tested required a large amount of

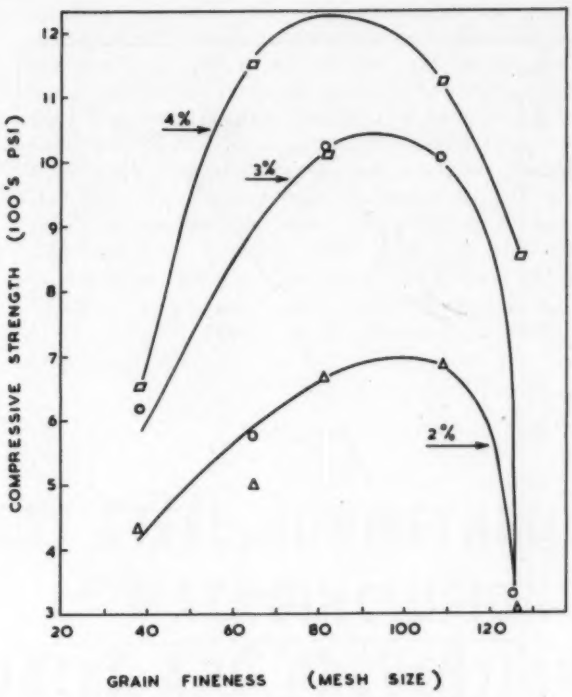



Fig. 3 — Maximum compressive strength obtained from 2, 3 and 4 per cent binder as a function of the grain size of the sand. Gassing pressures and time are not specified on this graph.

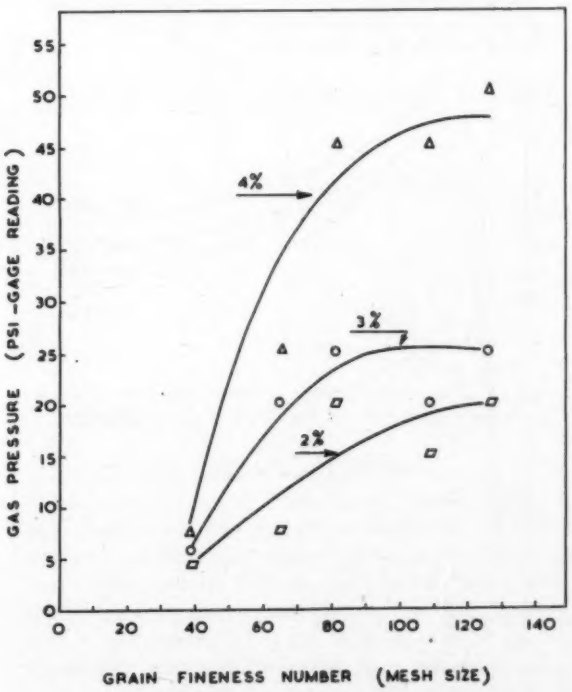



Fig. 4 — Gas pressure to produce maximum compressive strength as a function of grain size of the sand for 2, 3 and 4 per cent binder with 30 sec gassing time.

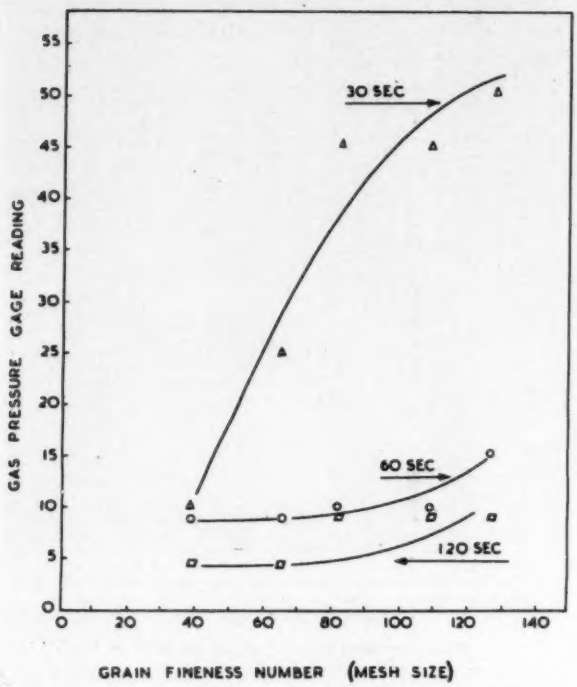



Fig. 5 — Gas pressure to produce maximum compressive strength as a function of grain size of the sand with 4 per cent binder for 30, 60 and 120 sec gassing time.

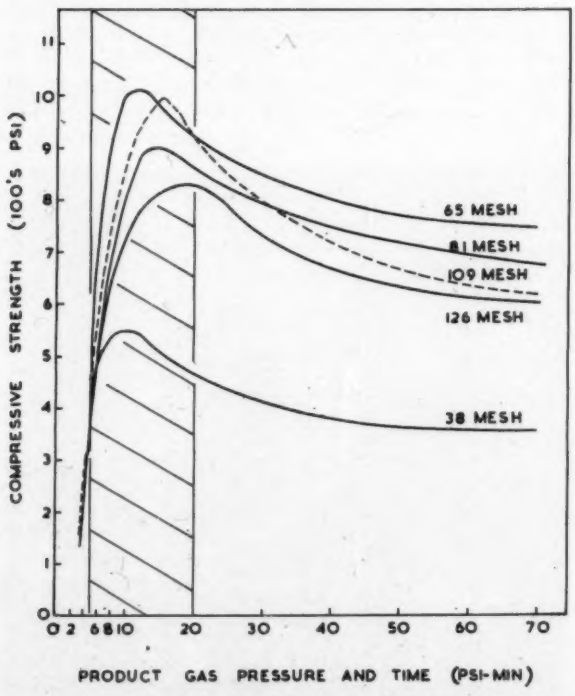



Fig. 7 — Average compressive strength as a function of the amount of gas passed through the specimen for all mesh size sands tested, the sand having 4 per cent binder.

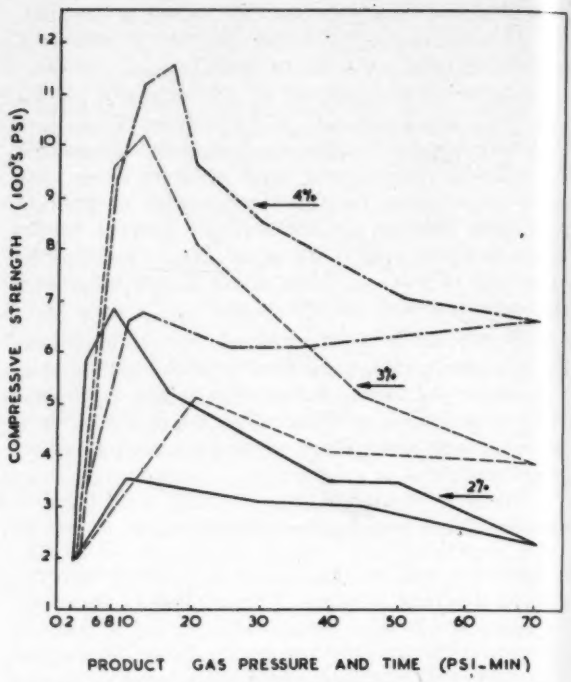



Fig. 6 — Range of compressive strengths as a function of amount of gas passed through the specimen for 2, 3 and 4 per cent binder, the sand being of 109 mesh size.

gas for full hardening. All the curves show the characteristics loss in strength with large amounts of gas.

## CONCLUSIONS

1. Within the range tested, higher strengths are obtained with higher percentages of binder.
2. The rate of increase in strength decreases with higher percentages of binder.
3. More binder requires a greater volume of gas to develop high strengths.
4. The amount of binder required increases as the mesh size of the sand decreases. Coarse sands do not form strong mixtures. Finer sands form harder mixtures if sufficient binder is present.
5. Shorter gassing time requires higher pressures to obtain high strengths.
6. Higher percentages of binder require higher gassing pressures at constant gassing time.
7. For a specific amount of gas, longer gassing times and lower pressures give higher strengths than shorter times and higher pressures.
8. There is a critical amount of gas that will give the maximum strength for each sand mixture. This amount of gas is higher for the finer sands.

It is the opinion of the authors that these conclusions apply to the $CO_2$ process in general, and can be proved under most conditions.

*The two papers which follow—"Cast Steel Nonmetallic Macroinclusions Sources and Prevention" by R. A. Flinn, W. B. Pierce and L. H. Van Vlack and "Refractories as Sources of Macroinclusions, An evaluation" by L. H. Van Vlack, J. E. Brokloff and R. A. Flinn are a report of the work which has been done during the past year at the University of Michigan sponsored by the AFS Training and Research Institute and under the direction of the Research Committee of the AFS Steel Division. These studies are a logical continuation of those which the Committee has been directing toward a solution of the problem of macroinclusions in steel castings. These inclusions are those which are called by such names as snotters, ceroxides, etc.*

*The work which is described in these papers demonstrates with more certainty the mechanism of formation of the defects, and discloses two leads for the prevention of the defects. These are ladle refractories and gating methods. During the coming year, work is planned on field studies to determine the effectiveness of changes in refractories and gating. The Committee is confident that much useful information has been obtained, and is hopeful that the field tests will complete the development of information which will be of considerable benefit to the steel foundryman. John A. Rassenfoss, Chairman, G. A. Colligan, A. J. Kiesler, W. A. Koppi, E. A. Lange, C. H. Lorig, E. Punko, W. R. Punko, D. N. Rosenblatt and J. Zotos.*

# CAST STEEL NONMETALLIC MACROINCLUSIONS SOURCES AND PREVENTION

Progress Report
Steel Division Research
Sponsored by
AFS Training & Research Institute

by R. A. Flinn, W. B. Pierce and L. H. VanVlack

## Part I


### ABSTRACT

**The nature, causes and prevention of macroinclusions in steel castings have been the subject of a three-year investigation employing laboratory and field research. In previous reports, confirmed by the present work, it has been determined that the most serious type of macroinclusion is composed of corundum crystals in a glassy silicate groundmass. Although this material may be formed (1) during deoxidation of the steel, (2) by reaction with ladle refractories and (3) in the mold itself, reaction (2) is the most serious source.**

**Present data indicate the inclusions may be eliminated by use of alumina or magnesia refractories, and by a whirl gate trapping system of new design. These indications will be evaluated by further field research.**


R. A. FLINN is Prof. of Met. Engrg., L. H. VAN VLACK is Prof. of Ceramics and Met. and W. B. PIERCE is with Dept. of Chemical and Met. Engrg., University of Michigan, Ann Arbor.

### INTRODUCTION

For several years the American Foundrymen's Society has sponsored research to determine the causes of large nonmetallic macroinclusions in steel castings. These usually take the form of shallow slag-like masses on cope or vertical surfaces, are usually about one in. in diameter but up to 3 in. and as deep as ¼-in. (Fig. 1). These inclusions have also been known as "ceroxides," and by other more descriptive, less delicate terms.

To indicate the position of this paper and its companion[1] in the overall research program, it may be helpful to review briefly the three phase plan deal-

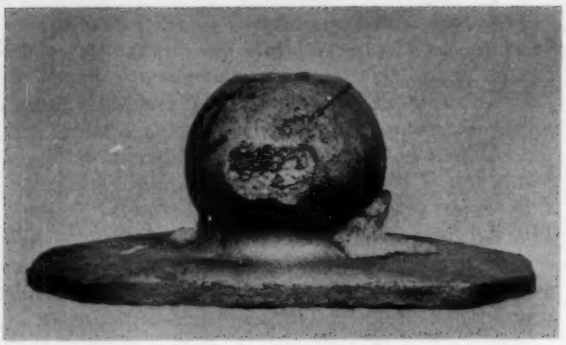


Fig. 1 — Typical nonmetallic macroinclusion on steel casting. The inclusion has been chipped away and moved down disclosing scar on casting. Scale approximately ½ natural size.

ing with macroinclusions which has been followed from the beginning:

1. Determine the nature of the macroinclusions.
2. Determine and evaluate the relative importance of the causes.
3. Develop methods for eliminating the defect.

### *Nature of Macroinclusions*

Since this type of macroinclusion is most prevalent in aluminum deoxidized, green sand castings, this occurrence has received major attention. In other words, since aluminum is widely used to prevent pinholing in green sand castings it seemed important to begin with these inclusions. In castings of this type, the macroinclusions have been identified[2] as silicate glass containing corundum ($Al_2O_3$) and occasionally hercynite ($FeO \cdot Al_2O_3$) crystals (Fig. 2). Fayalite ($Fe_2SiO_4$) is often found in the inclusion. However, this is formed during cooling and is not present at the time of casting.

### *Inclusion Sources*

There are three possible sources of the macroinclusions:

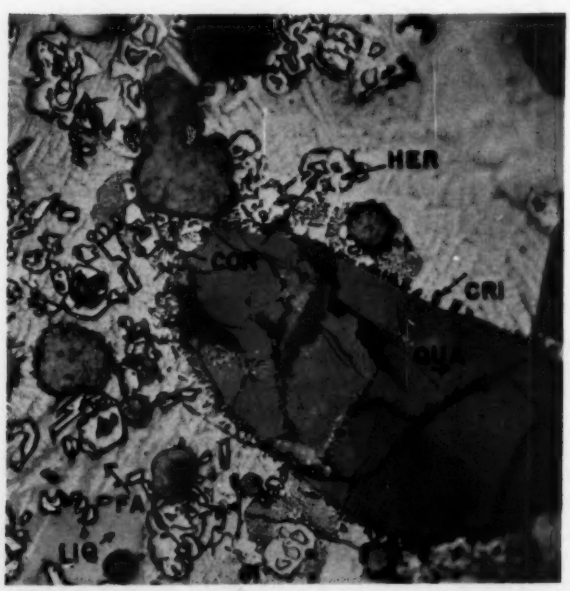



Fig. 2 — Macroinclusion. COR — corundum ($Al_2O_3$), HER — Hercynite ($FeO \cdot Al_2O_3$), FAY — Fayalite ($Fe_2SiO_4$), LIQ — Liquid (now glass), QUA — Quartz ($SiO_2$) from sand, CRI — Cristobalite ($SiO_2$).

1. Formation during deoxidation of the steel.

   $(\underline{Al} + \underline{O} \rightarrow Al_2O_3$ plus silicates, etc.)

2. Reaction between liquid steel and ladle refractories.

   $(\underline{Al} + SiO_2 \rightarrow Al_2O_3 + \underline{Si})$

   Corundum

   Other oxides may be formed concurrently (MnO, FeO, etc.).

3. Reaction between liquid steel and mold dirt and sand.

   $(Al + SiO_2 \rightarrow Al_2O_3 +$ silicates, etc.)

   excess

It has been demonstrated repeatedly[2] that the ladle scum rising after deoxidation, the reaction products on nozzles and other refractories and the macroinclusions in castings are identical in their constituents. Of course the inclusions in the cope surface of the casting will often contain quartz grains from the sand and alteration products as well as the phases just discussed.

After extensive investigation it appears that the inclusions resulting from reaction with ladle refractories are the most troublesome. In tests involving bottom pour ladles, for example, the severity of inclusions increases rather than decreases as a function of holding time in the ladle before pouring. Observation of the metal stream from a teapot or open ladle shows increasing amounts of slag accompanying the metal as time passes.

Attack of the refractory proceeds in the absence of aluminum, by other components of the steel such

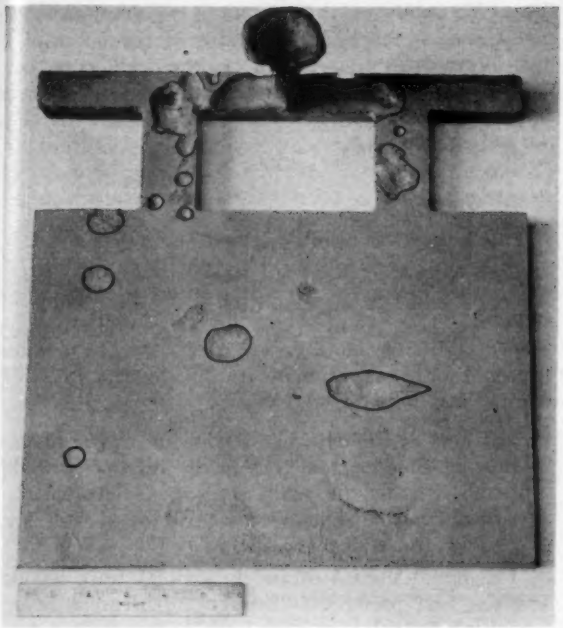


Fig. 3 — Cross gated test casting 10 x 14 x ½-in. showing synthetic inclusions (outlined).

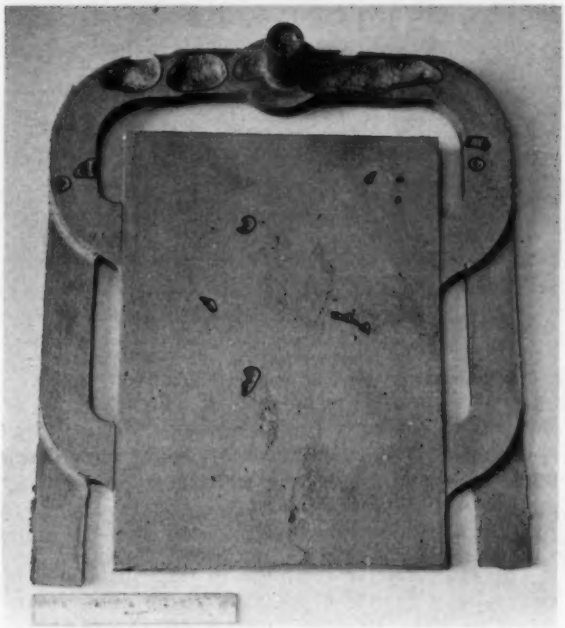


Fig. 4 — Streamline gated test casting 10 x 14 x ½-in. showing synthetic inclusions (outlined).

as iron and manganese. However, the severe inclusions accompanying the presence of aluminum may be related to the effect of the corundum crystals and the alumina in the glassy groundmass of the inclusion in raising its freezing point. Thus, when the inclusion reaches its position in the mold, it is viscous and is not absorbed by the sand. Instead, it protrudes into the metal requiring weld repair.

### *Elimination of Macroinclusions*

As a result of the previous investigation it was decided that three principal courses were open to eliminate this type of macroinclusion.

*(A) Eliminate aluminum as a deoxidant.* When the aluminum is eliminated or reduced to low levels, $<0.02$ per cent, inclusions are drastically reduced. Unfortunately, at present aluminum is a specific for pinhole control in green sand castings. Until other deoxidants are developed and generally accepted which do not attack refractories but eliminate pinholes this course must be deferred.

*(B) Develop and employ refractories which resist attack.* During the past year this phase received major attention as reported in a companion paper.[1] The roles of refractory type, aluminum and manganese contents, time of contact and temperature of liquid steel were investigated. High alumina or magnesia appear superior to fireclay or silica under the more erosive conditions.

It should be pointed out however that, if more resistant refractories are employed, steps should be taken to prevent furnace slag and deoxidation products from building up in the ladle. These can flux the ladle refractories or develop serious accumulations.

*(C) Design and test gating systems to trap the inclusions.* Since these inclusions are relatively large, it appeared that there was adequate opportunity to trap them in the gating system.

Because of the detailed nature of the work under (*B*) this is reported separately. We can proceed to discuss progress on phase (*C*) as a portion of this paper.

## GATING SYSTEMS DEVELOPMENT FOR MACROINCLUSIONS ELIMINATION

Many gating systems have been devised in the past for trapping inclusions, dross and dirt. In this investigation it seemed advisable to select three prin-

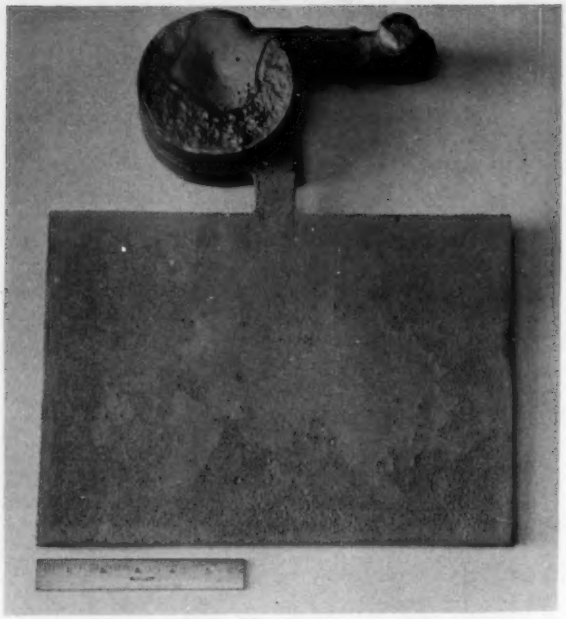


Fig. 5 — Whirl gated test casting 10 x 14 x ½-in. showing trapping of synthetic inclusions (outlined).

cipal types (1) the cross gate, (2) the settling basin with flat gates and (3) the whirl gate or riser.

### *Cross Gate*

The cross gate is intended to trap inclusions by employing a change in metal direction. This is supposed to make it more difficult for the inclusions to follow the metal stream into the mold. A large number of variations in this design have been described in the literature. Among the significant differences are the vertical dimension of the cross gate (runner) compared to the ingots, placing of the runner in the cope ingate in the drag and conversely. The simplest form was used in the preliminary experiments, Fig. 3.

### *Settling Basin With Flat Gates*

In other work sponsored by the American Foundrymen's Society[3] a flat streamlined gating system was developed to prevent dross in aluminum, and has been applied with success to ferrous alloys. It would be expected that a system of this type might trap these macroinclusions in steel. This system is illustrated in Fig. 4. The pertinent details are:

1. A basin at the base of the tapered downsprue with an area five times that of the bottom of the downsprue.
2. Extension of the downsprue below the gate greater than gate height, omitted in the figure.
3. Total gate area from settling basin 4-6 times downsprue area.
4. Casting ingates above the runner.
5. Reduction in runner area on passing ingates.
6. Runner extension beyond last ingate.

The philosophy of this system is to decelerate the metal once it reaches the bottom of the downsprue and allow dross to separate on the wide cope surfaces of the gates. The runner extension is to trap the dross in the first metal passing through the runner system.

### *Whirl Gate*

This system has been described in foreign and U.S. publications, and relies upon centrifugal action to whirl the inclusions to the center of the whirl gate, Fig. 5.

Several design features are important as demonstrated by the experiments to be described. These features are:

1. The area of the exit, i.e., the ingate to the casting must be smaller than that of the inlet so that the metal rises in the whirl and the inclusions are carried above the gate area. A ratio of 1 : 1.5 in area was found desirable.
2. The metal should rotate through at least 180 degrees, and preferably through 270 degrees before passing the exit.
3. To provide additional rotation for the first metal before it enters the mold, the cylindrical section may be extended below the level of the exit.

## TEST PROCEDURES

Test castings of several sizes were tried at different stages of the investigation. The change toward larger dimensions was made in the effort to develop a more critical test casting. The sizes used were flat rectangular plates:

A. 5½ x 6½ x ⅜-in.
B. 1 x 4 x 6 in.
C. 8 x 8 x 1 in.
D. 14 x 10 x ½-in.

The effect of size will be discussed under Results.

### *Synthetic Macroinclusions Use*

To provide for rapid testing of the different gating systems, 100 lb 3000 cycle induction furnace heats were made of Grade B steel using the standard rammed magnesia lining. Lip pour ladles lined with fireclay and magnesia were tried in the initial experiments. It was soon determined that it was difficult to produce characteristic large macroinclusions with these techniques, because the time of contact with the lining was insufficient to produce sufficient reaction.

Because of the size of the heat, the time in the ladle could only be of the order of one min compared with 10 to 50 times this period encountered in production heats. Only a few small inclusions were obtained even when the gating system was altered to enter the casting by a simple gate from the base of the downsprue, as in Fig. 6.

It was decided that the deliberate introduction of synthetic macroinclusions in the form of small glass

balls would provide a more severe test to gating traps than the occasional small inclusions encountered in the laboratory. Experiments to be described later indicate that this technique also provides a more severe test than production conditions.

Accordingly, a stream of glass beads about ¼-in. in diameter is introduced with the metal at the top of the downsprue at the rate of about 15 beads/sec. The beads are of a low melting point sodium glass with a low density comparable to the actual macroinclusions. The beads soften rapidly and pass through the system. They can be found on the cope surfaces, Figs. 3, 4 and 5.

## LABORATORY RESULTS

The various gating systems were tried with different plate sizes with these results:

| 1 x 4 x 6 in. (Type B) | |
|---|---|
| (1) Cross gate | Inclusions in casting and runner. |
| (2) Flat gate | No macroinclusions in casting, but present in the runner. |
| (3) Whirl gate | |
| 270°* | Inclusions trapped completely, casting clean. |
| 180° | Most inclusions trapped, traces of inclusions. |
| 90° | Poor trapping, inclusions in casting. |

*Rotation of metal inlet to exit.

The results for the type C plate were essentially the same as for the type B. Finally it was decided to use a larger plate to provide more cope surface and also to permit the use of two ingates per side when the flat gating system was tried, Fig. 4. These castings showed the effects:

| Type D (14 x 10 x ½-in. plate) | |
|---|---|
| Cross gate | Inclusions in gates and casting. |
| Flat gate | Inclusions in gates and casting. |
| Whirl gate | |
| 270° | Inclusions trapped in whirl, none in castings. |
| 90° | Inclusions in gates and castings. |

From these data it was concluded that the 270 degree whirl gate provided a successful trap in all cases. The flat gate was satisfactory for smaller castings but allowed some inclusions to reach the larger casting.

## FIELD TESTS

At the present time tests have been made at only one producer, and therefore only tentative conclusions may be drawn.

In the first series of tests molds were made of the type D pattern using the three different gating systems just described. A standard jolt rollover machine and a typical synthetic green sand were employed. Aluminum additions totalling 0.2 per cent were made, 0.15 per cent to the bull ladle and 0.05 per cent to the pouring ladle. The important point is that no macroinclusions were encountered in any of the test plates, gated by the three methods described. However, macroinclusions were obtained in all of small type A plates (Fig. 6) which had simple direct ingates, and which were poured for comparison.



Fig. 6 — Simple ingate on test plate 5½ x 6½ x ⅜-in. showing natural inclusions (outlined).

Another heat was poured using simple ingates for the type D plates and characteristic macroinclusions were obtained (no synthetic inclusions were used in these plant tests).

From these tests it may be tentatively concluded:

1. The cross gate, flat gate and whirl gate all are superior to the straight gating system.
2. Under highly critical conditions, simulated by the synthetic macroinclusions, the results from the whirl gate appear best at the present time.

## ACKNOWLEDGMENT

The authors wish to acknowledge the many helpful suggestions and continued support of the A.F.S. Steel Division Research Committee: J. Rassenfoss, *Chairman*, G. A. Colligan, A. J. Kiesler, W. Koppi, E. Lange, C. Lorig, S. C. Massari, E. Punko, W. Punko and J. Zotos.

We also appreciate the cooperation of Mr. S. F. Lukacek, Vice-President, The Wehr Steel Co. in performing the plant experiments.

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# REFRACTORIES AS SOURCES OF MACROINCLUSIONS

## *An evaluation*

Progress Report
Steel Division Research
Sponsored by
AFS Training & Research Institute

by L. H. Van Vlack, J. E. Brokloff and R. A. Flinn

## Part II

### ABSTRACT

A companion paper[1] summarizes the work which has been done to date in determining the (a) nature, (b) source and (c) control of macroinclusions in steel castings. One of the important sources is from metal-refractory interaction, and this paper is concerned with the details of this interaction.

This phase of the investigation establishes a test procedure to evaluate refractories for their stability in the presence of molten steel. In addition, consideration is given to the relative effects of (a) metal composition, (b) time and (c) temperature upon the molten metal-refractory reactions. Pertinent conclusions are presented within the report.

Four principal types of refractories were selected for evaluation: silica, fireclay, alumina and magnesia. Although the latter two types are more expensive per pound, it is considered possible that their greater stability might lead to a lower cost per pound of castings shipped.

### INTRODUCTION

Molten metals react with many of the commonly available refractories. The consequences are evident in two forms—(a) The refractories are eroded rapidly and therefore lead to high operating costs and production difficulties and (b) The reaction products can be sources of casting imperfections which decrease quality unless specific and expensive repair procedures are used.

The purpose of this work, sponsored by the American Foundrymen's Society Training and Research Institute under the direction of the AFS Steel Division, has been to provide a simple test to analyze the nature and the extent of the reactions between molten steel and foundry refractories.

L. H. VAN VLACK is Prof. of Ceramics and Met., J. E. BROKLOFF is Graduate Student and R. A. FLINN is Prof. of Met. Engrg., Dept. of Chemical and Met. Engrg., University of Michigan, Ann Arbor.

### PROCEDURE

1. Steels of the desired base composition are melted in a small (60 lb) induction furnace.
2. Final alloy additions are made and the surface slag removed.
3. Small (0.5 x 0.5 in. cross-section) refractory samples are immersed into the steel for a period of 5 min (The furnace power is reduced so as to maintain only slow stirring during the test).
4. The characteristic reaction is observed at the surface below the metal line.
5. The necessary precautions are only those that would be anticipated on the basis of common metallurgical practices such as—melting, oxidation, slag-off and power supply.

As previously reported,[2,3] it is possible to observe reactions between refractories and molten steel on the basis of the laboratory test scheme, shown in Fig. 1.

The refractory sample is prepared by:

1. The refractory to be tested is cut from a brick, nozzle, or other source so that the dimensions of 0.5 x 0.5 x 4 in. are available.
2. These are mounted into slots cut out of a soft insulating brick holder. A fireclay mortar slurry is used for cementing purposes.
3. Two or more samples (e.g., the current refractory for reference, plus new refractories) may be tested concurrently. The holder and cement do not contact the melt.

The results of testing are shown in Fig. 2. The severity of the reaction is classified from 0 (none)

to 5 (complete erosion). The nature of the reaction varies from a liquid product which is easily absorbed by a porous ladle or mold refractory, through a semi-viscous glass, to a solid oxidation product which accumulates on the refractory surface. The notations of (p), (v) and (a) are given in Fig. 2 for the products which are respectively absorbed into the pores (p) as a fluid liquid, carried with the moving metal as a viscous (v) liquid or accumulated (a) on the surface as a solid accretion. The region of greatest significance is the tip or the part of the test sample that was most deeply immersed.

The requirements and precautions of the test involve a specified time, avoidance of excessive surface accumulations on the molten metal and of excessive oxidation. Five min exposure has proved to be satisfactory, because the 0.5 x 0.5 in. refractory is heated sufficiently for the reaction to proceed to a satisfactory extent. Additional time reveals little additional information. Surface accumulations on the molten metal of fluxed refractory or oxidation products must be removed during the course of the test.

This is particularly true when more than one type of sample is being tested, because reaction products from one type of sample may mask the surface of the neighboring sample. Excessive oxidation of the melt during the test will produce extra FeO at the surface, and can mislead the inexperienced analyst. Therefore, it is recommended that the surface of the molten steel be covered with tank nitrogen during the course of the test from the time of alloy additions until completion.

## RESULTS

The results of experiments using the test procedure just described corroborate information which has been observed in previous plant experience. In addition, it has led to an indication of the relative significance of several contributing factors in refractory erosion. These will be summarized in the balance of the paper. Table 1 summarizes the test data and Table 2 condenses the results.

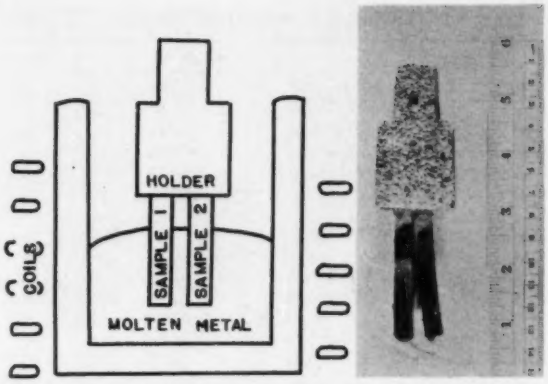



Fig. 1 — Experimental procedure. *Left* — schematic *right* — sample after testing. Direct contact was made between molten metal and refractory samples. Preference is given to the use of a nitrogen blanket over the metal. However, air may be used if samples are deeply immersed in metal so surface oxides are avoided.[3]

### *Aluminum Effect*

The interaction of dissolved aluminum and refractories was described previously.[3] In review, fireclay refractories are most severely affected. A viscous reaction product is formed which is subject to erosion by the metal stream (Fig. 3). Silica refractories are less severely affected. The product is more fluid and is absorbed by the refractory. Refractories with alumina higher than that found in fireclay form a solid reaction product which is not eroded by the metal stream, and may even be subject to surface accumulation.

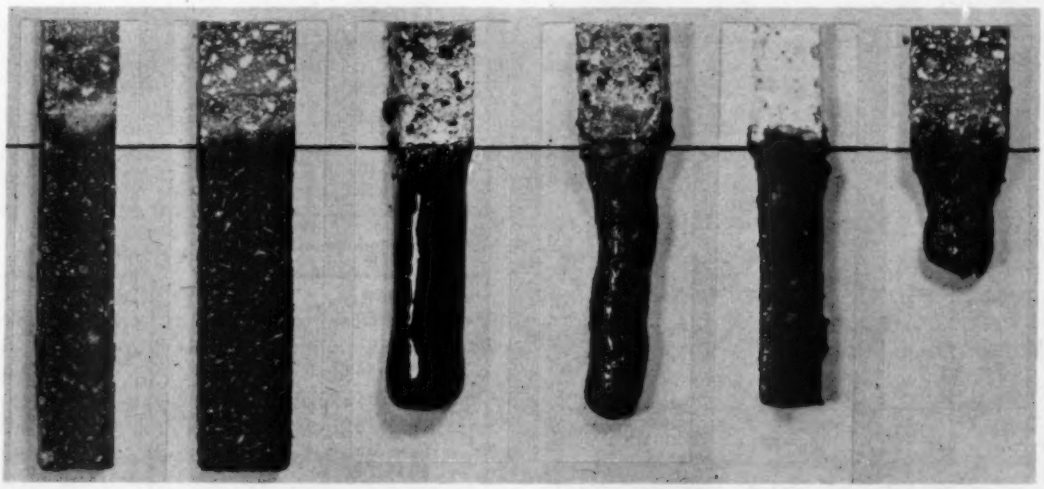


Fig. 2 — Rating of refractory reaction in molten metal. Line in figure represents the metal line. *Left to right* — *1p* — slight fluxing, fluid liquid is absorbed into refractory pores; *2p* — fluxing, fluid liquid is absorbed into refractory pores; *2v* — fluxing, a viscous liquid product is formed; *3v* — extensive reaction and fluxing, a viscous liquid is formed; *3a* — extensive reaction and accumulation of a solid reaction product; *4v and 5v* — severe fluxing and erosion.

TABLE 1 — MELTING DATA

| Heat | Temp., F | Time, min | Analysis, % C | Si | Mn | Al | Aim, % Al |
|---|---|---|---|---|---|---|---|
| F-42 | 2830 | 10 | 0.38 | 0.36 | 0.37 | — | — |
| F-43 | 2970* | 10 | 0.36 | 0.38 | 0.42 | — | 0.2 |
| F-44 | 2850 | 5 | 0.33 | 0.41 | 0.36 | — | — |
| F-45 | 2845 | 5 | 0.31 | 0.45 | 0.38 | 0.21 | 0.2 |
| F-46 | 2960 | 5 | 0.28 | 0.49 | 0.35 | 0.21 | 0.2 |
| F-47 | 2840* | 5 | 0.31 | 0.22 | 0.25 | — | — |
| F-48 | 2850* | 5 | 0.31 | 0.33 | 0.30 | — | — |
| F-49 | 2880* | 5 | 0.29 | 0.51 | 0.36 | — | — |
| F-52 | 2845* | 5 | 0.33 | 0.41 | 3.28 | — | — |
| F-53 | 2850* | 5 | 0.30 | 0.47 | 3.26 | >0.04 | 0.2 |
| F-54 | 2960* | 5 | 0.28 | 0.51 | 3.28 | 0.04 | 0.2 |
| F-55 | 2845* | 10 | 0.34 | 0.41 | 2.79 | — | — |
| F-56 | 2950* | 10 | 0.30 | 0.45 | 2.71 | 0.05 | 0.2 |

*Temperatures by immersion thermocouple, others by optical pyrometer.

TABLE 2 — SUMMARY OF RESULTS (Figs. 2 and 4)

| | I | II | III | IV | V |
|---|---|---|---|---|---|
| Temperature, F | 2850 | 2950 | 2950 | 2850 | 2850 |
| Time of contact, min | 5 | 5 | 10 | 5 | 10 |
| Aluminum | yes | yes | yes | no | no |
| Refractory — Manganese (%) | | | | | |
| S Silica — Lo Mn( 0.5) | 1* | 1 | 3 | 1 | 2 |
| S Silica — Hi Mn( 2.75) | 4-5 | 4-5 | 4-5 | 4-5 | 4-5 |
| F Fireclay — Lo Mn( 0.5) | 1a | 1a | 4 | 2 | 3 |
| F Fireclay — Hi Mn( 2.75) | 3a | 4-5 | 4-5 | 4-5 | 4-5 |
| A Alumina — Lo Mn( 0.5) | 0 | 0 | 0a | 0 | 0 |
| A Alumina — Hi Mn( 2.75) | 0a | 1 | 2a | 1a | 1 |
| M Magnesia — Lo Mn( 0.5) | 0 | 0 | 0a | 0 | 0 |
| M Magnesia — Hi Mn( 2.75) | 0a | 1a | 1 | 0a | 0a |

* See Fig. 2 for notation.

### *Manganese Effect*

The role of manganese varies with other factors. These are presented in Figs. 5-9 and summarized in Fig. 4. However, each will be given individual attention.

*Figure 5 — Effect of manganese (with no aluminum, 2850 F and standard time).* Fireclay refractories are affected more than other types at low manganese levels, while silica refractories are affected most severely at high manganese levels. Alumina and magnesia refractories are subject to negligible reaction at levels up to 3-4 per cent manganese. In contrast, they will accumulate metal oxidation products on their surfaces.

*Figure 6 — Effect of manganese (with no aluminum, 2850 F and longer times).* The reactions are increased as compared with standard times, otherwise the relative relationships described in the previous paragraph are applicable.

*Figure 7 — Effect of manganese (with aluminum deoxidation, 2950 F and standard time).* Manganese in-

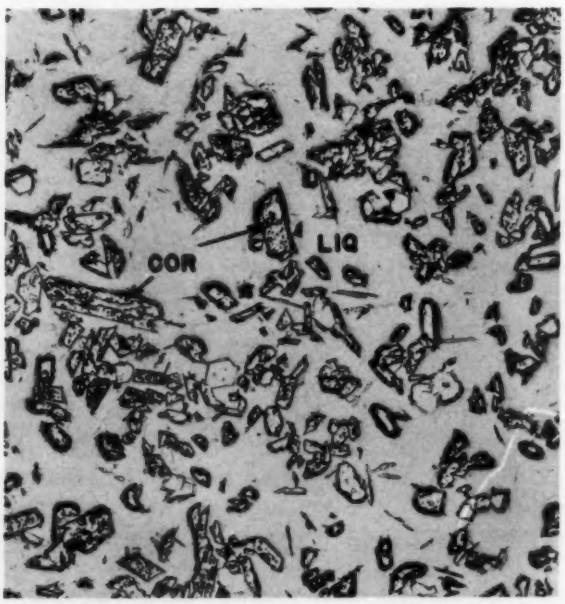



Fig. 3 — Corundum (COR), $Al_2O_3$, is formed as a reaction product between the aluminum in the molten steel and the oxygen in the refractories and steel. Liquid (LIQ) is now glass.[3]

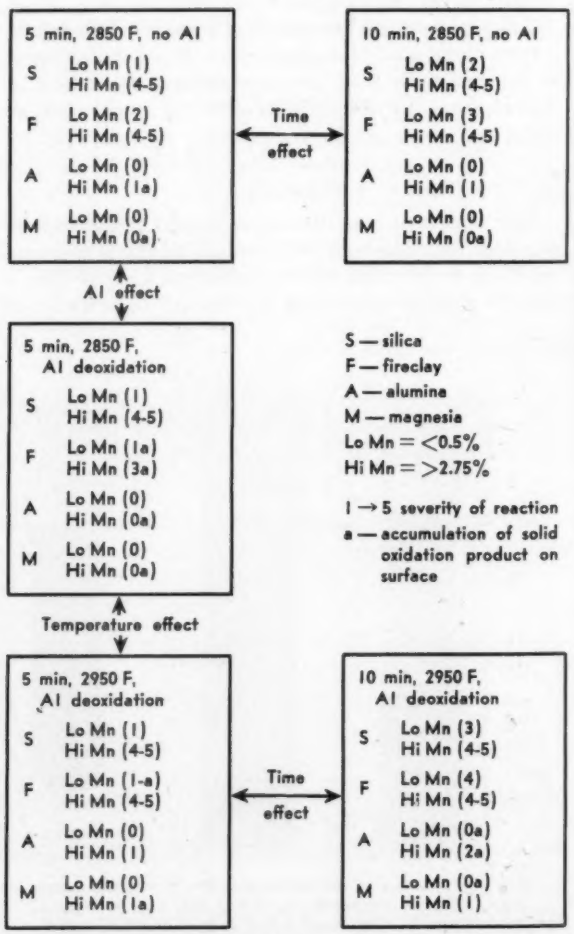



Fig. 4 — Summary of the role of manganese as it varies with other factors, as presented in Figs. 5 - 9.

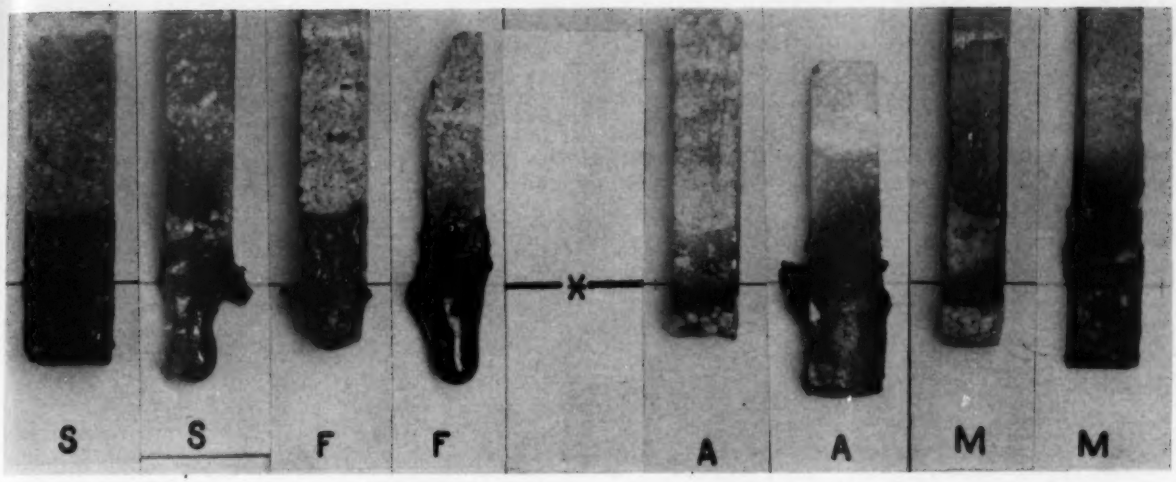
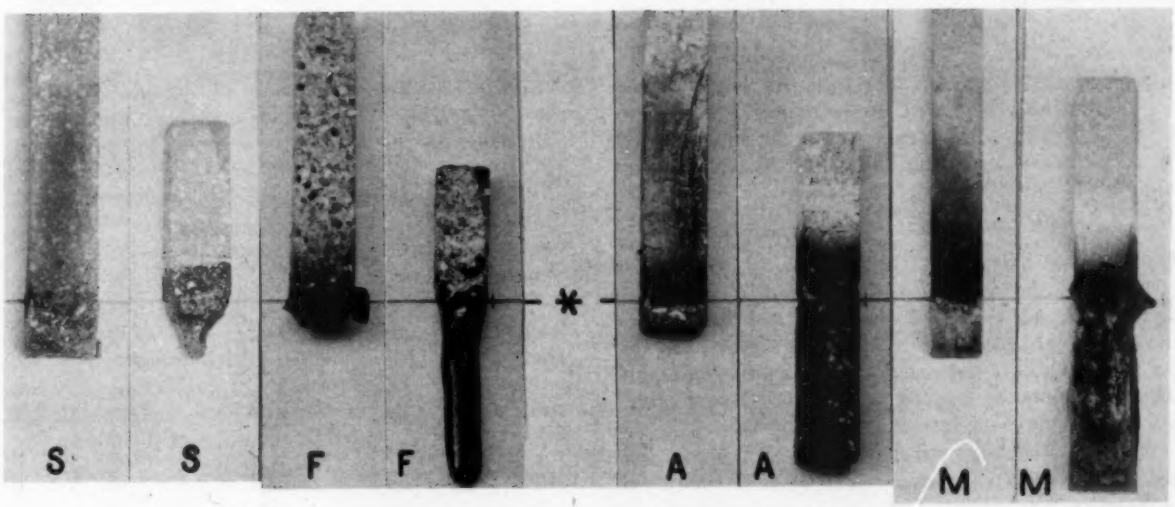



Fig. 5 — Manganese content variable, temperature (2850 F) and aluminum content (none) fixed (Table 1 presents exact data). Contact time — 5 min. * = metal line. a = accumulation of oxidation product on sample surface.

| Position / % Mn / Test / Refractory | Left 0.36 F-44 | Right 3.28 F-52 |
|---|---|---|
| S - silica | 1 | 4-5 |
| F - fireclay | 2 | 4-5 |
| A - alumina | 0 | 1-a |
| M - magnesia | 0 | 0-a |

Fig. 6 — Manganese content variable, temperature (2850 F) and aluminum content (none) fixed (Table 1 presents exact data). Contact time — 10 min. * = metal line. a = accumulation of oxidation product on sample surface.

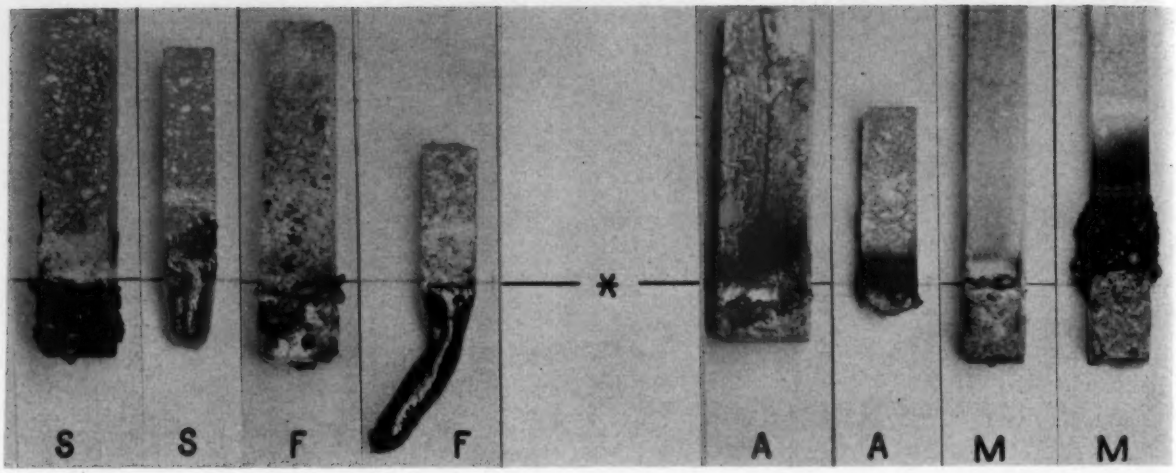



creases the reaction. At low manganese levels, a solid reaction product is formed as a result of the aluminum deoxidizer.

*Figure 8 — Effect of manganese (with aluminum deoxidation, 2950 F and longer times).* The increased time increases total attack, otherwise the relative relationships are comparable.

*Figure 9 — Effect of manganese (with aluminum deoxidation, 2850 F and standard time).* The reaction gives less liquid and more solid (with corundum and spinel) than shown in Fig. 5 without aluminum

| Position / % Mn / Test / Refractory | Left 0.37 F-42 | Right 2.79 F-55 |
|---|---|---|
| S - silica | 2 | 4-5 |
| F - fireclay | 3 | 4-5 |
| A - alumina | 0 | 1 |
| M - magnesia | 0 | 0-a |




| Position / % Mn / Test / Refractory | Left 0.35 F-46 | Right 3.28 F-54 |
|---|---|---|
| S - silica | 1 | 4-5 |
| F - fireclay | 1-a | 4-5 |
| A - alumina | 0 | 1-x |
| M - magnesia | 0 | 1-a |

**Fig. 7 — Manganese content variable, temperature (2950 F) and aluminum deoxidation fixed (Table 1 presents exact data). Contact time 5 min. * = metal line. a = accumulation of oxidation product on sample surface. x = sample broke during test.**

deoxidation. Otherwise the results are comparable. Aluminum must be added for corundum to be present. There is no solid accumulation on magnesia refractories unless both aluminum and manganese are added.

### *Temperature Effect*

The role of temperature was examined at 2850 F and 2950 F. Above 3000 F the silica and fireclay refractories melt from the temperature alone, and do not require additional reactions for failure. Although in normal practice tapping temperatures are greater, these represent characteristic ladle temperatures.

*Figure 10 — Effect of temperature (with aluminum deoxidation; low manganese and standard time).* An increase in temperature from 2850 F to 2950 F shows little effect when the manganese is low (<0.5 per cent).

*Figure 11 — Effect of temperature (with aluminum deoxidation and standard time).* Temperature has little effect on the attack on the silica in the 2850 F to 2950 F range. It does have some effect upon fireclay, indicating that the solid reaction products are being fluxed.

### *Time of Contact Effect*

The effect of the time of contact is one of aggravating the reaction. No new nor change in reactions is found.

*Figure 12 — Effect of time of contact (with no aluminum, 2850 F and low manganese).* Only minor differences in the amount of reaction for periods up to 10 min.

*Figure 13 — Effect of time of contact (with no aluminum, 2850 F and high manganese).* The time has appreciably less effect than the increase in the amount of manganese.

*Figure 14 — Effect of time of contact (with aluminum deoxidation, 2950 F and low manganese).* A slight effect of the added time can be observed.

*Figure 15 — Effect of time of contact (with aluminum deoxidation, 2950 F and high manganese).* The effect of time is masked by the effect of the manganese.

### *Alloy Source*

*Figure 16 — Effect of Ca-Mn-Si alloy as a source of manganese.* As long as the amount of manganese stays the same, the presence of calcium and silicon as alloy additions provides no change in the amount of reaction between the metal and any of the refractories. As previously noted, the variation between refractories is evident.

### *Reaction Products*

A glassy reaction product is formed when silica brick is attacked by steel. Either aluminum or manganese will accentuate the amount of glass. In the case of firebrick, the reaction product which is formed is more affected by the composition of the steel than

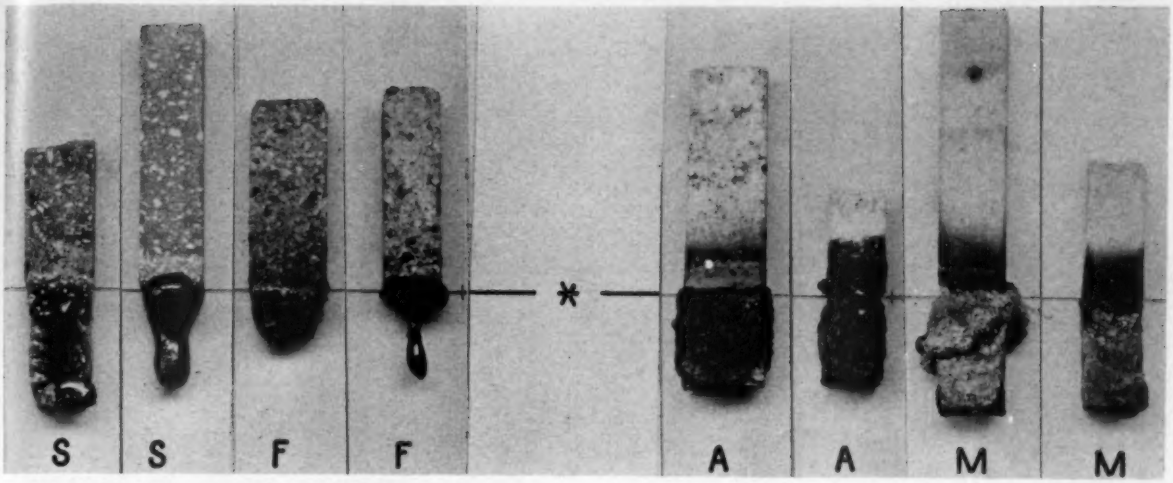
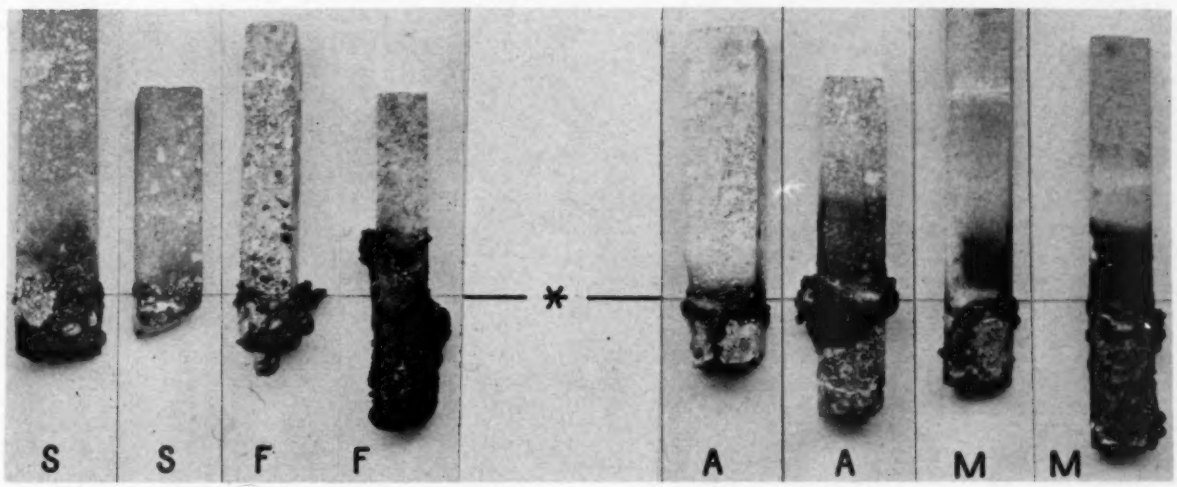



Fig. 8 — Manganese content variable, temperature (2950 F) and aluminum deoxidation fixed (Table 1 presents exact data). Contact time — 10 min. * = metal line. a = accumulation of oxidation product on sample surface.

| Position / % Mn / Test / Refractory | Left 0.42 F-43 | Right 2.71 F-56 |
|---|---|---|
| S - silica | 3 | 4-5 |
| F - fireclay | 4 | 4-5 |
| A - alumina | 0-a | 2-a |
| M - magnesia | 0-a | 1 |

Fig. 9 — Manganese content variable, temperature (2850 F) and aluminum deoxidation fixed (Table 1 presents exact data). Contact time — 5 min. * = metal line. a = accumulation of oxidation product on sample surface.




| Position / % Mn / Test / Refractory | Left 0.38 F-45 | Right 3.26 F-53 |
|---|---|---|
| S - silica | 1 | 4-5 |
| F - fireclay | 1-a | 3-a |
| A - alumina | 0 | 0-a |
| M - magnesia | 0 | 0-a |

in the case of the silica-steel reaction. Manganese will produce a low melting glass. Aluminum produces a more viscous glass which contains corundum ($Al_2O_3$) crystals (Fig. 3). The latter are not encountered in the absence of aluminum additions, because they are formed by aluminum reacting with oxygen in the refractories and in the steel.

Higher alumina brick (more than the typical 35 per cent $Al_2O_3$ which is present in fireclay refractories) are not subject to alteration by molten steel unless high amounts of manganese are present. In

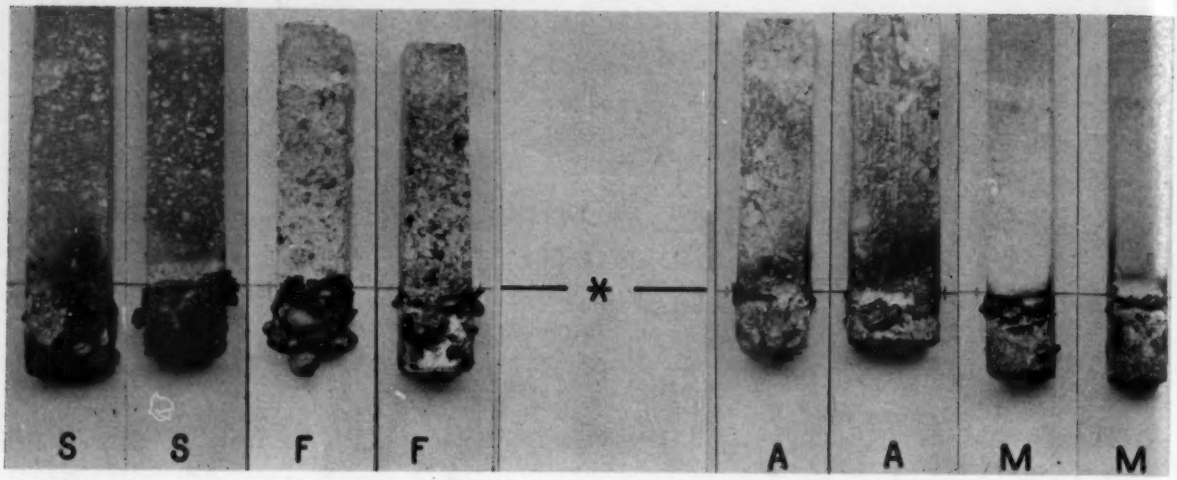
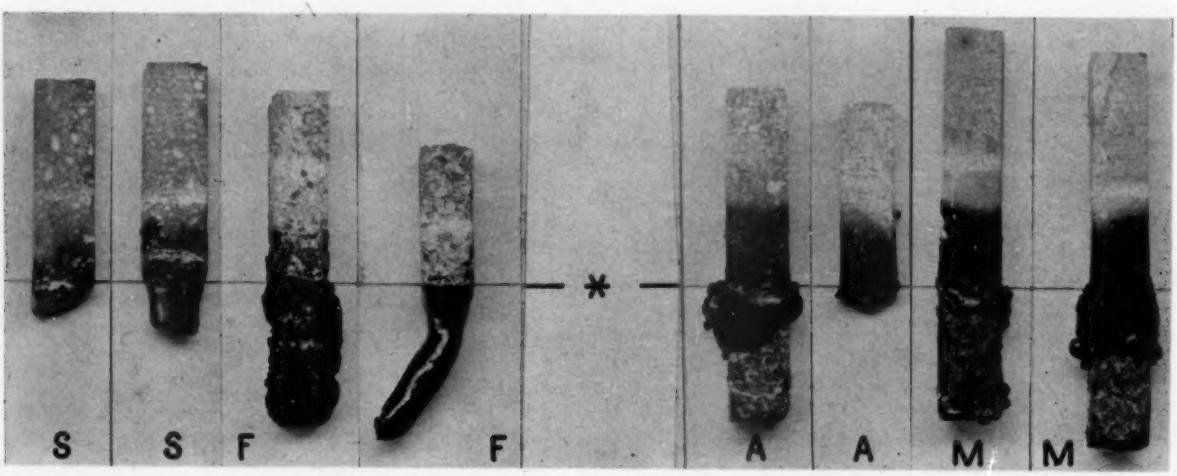



Fig. 10 — Temperature variable, low manganese (<0.5%) and aluminum deoxidation fixed (Table 1 presents exact data). Contact time — 5 min. * = metal line. a = accumulation of oxidation product on sample surface.

| Position / Temperature / Test / Refractory | Left 2850°F F-45 | Right 2950°F F-46 |
|---|---|---|
| S - silica | 1 | 1 |
| F - fireclay | 1-a | 1-a |
| A - alumina | 0 | 0 |
| M - magnesia | 0 | 0 |

Fig. 11 — Temperature variable, high manganese (>2.75%) and aluminum deoxidation fixed (Table 1 presents exact data). Contact time — 5 min. * = metal line. a = accumulation of oxidation product on sample surface. x = sample broke during test.




| Position / Temperature / Test / Refractory | Left 2850°F F-53 | Right 2950°F F-54 |
|---|---|---|
| S - silica | 4-5 | 4-5 |
| F - fireclay | 3-a | 4-5 |
| A - alumina | 0-a | 1-x |
| M - magnesia | 0-a | 1-a |

this case, galaxite ($MnO \cdot Al_2O_3$) is encountered. This mineral is a member of the spinel family and similar in appearance to hercynite ($FeO \cdot Al_2O_3$) previously described.[3]

Magnesia refractories are not altered by MnO nor FeO. In an isolated instance, the mineral, spinel ($MgO \cdot Al_2O_3$) was observed as the result of a solid reaction with the alumina deoxidation product.

When aluminum and manganese are both present in the steel, galaxite ($MnO \cdot Al_2O_3$) is encountered as a deoxidation product. This will form as an accretion on both magnesia and high alumina refractories.

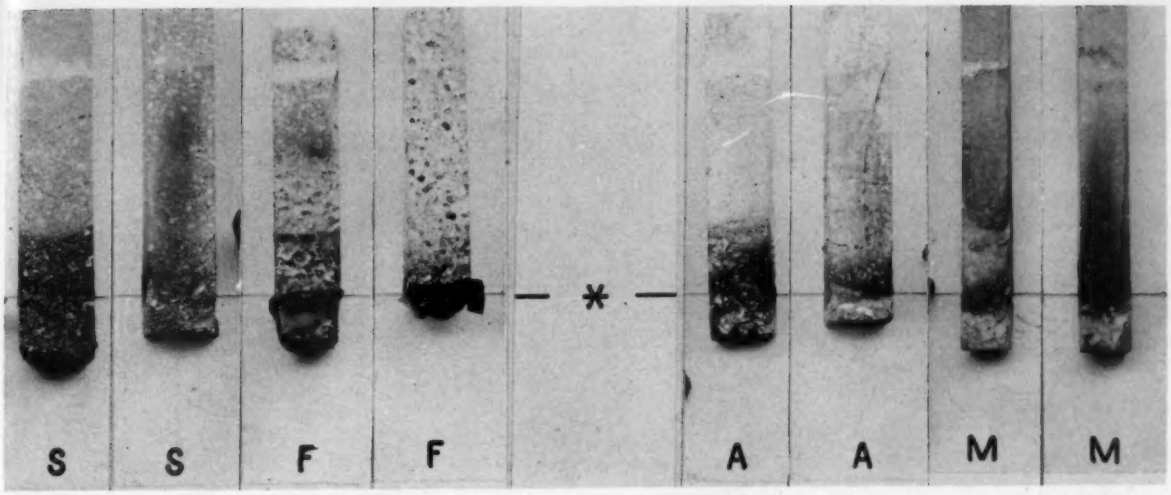
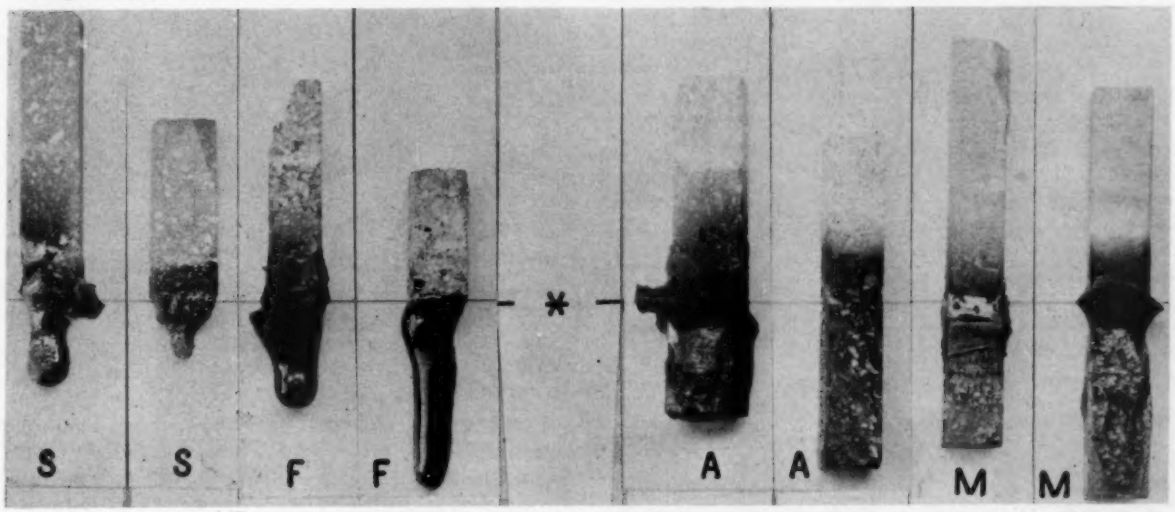



Fig. 12 — Time of contact variable, temperature (2850 F), aluminum content (none) and low manganese (<0.5%) fixed (Table 1 presents exact data). * = metal line.

| Position / Time / Test / Refractory | Left 5 min. F-44 | Right 10 min. F-42 |
|---|---|---|
| S - silica | 1 | 2 |
| F - fireclay | 2 | 3 |
| A - alumina | 0 | 0 |
| M - magnesia | 0 | 0 |

Fig. 13 — Time of contact variable, temperature (2850 F), aluminum content (none) and high manganese (>2.75%) fixed (Table 1 presents exact data). * = metal line. a = accumulation of oxidation product on sample surface.




| Position / Time / Test / Refractory | Left 5 min. F-52 | Right 10 min. F-55 |
|---|---|---|
| S - silica | 4 | 5 |
| F - fireclay | 4 | 4 |
| A - alumina | 1-a | 1 |
| M - magnesia | 0-a | 0-a |

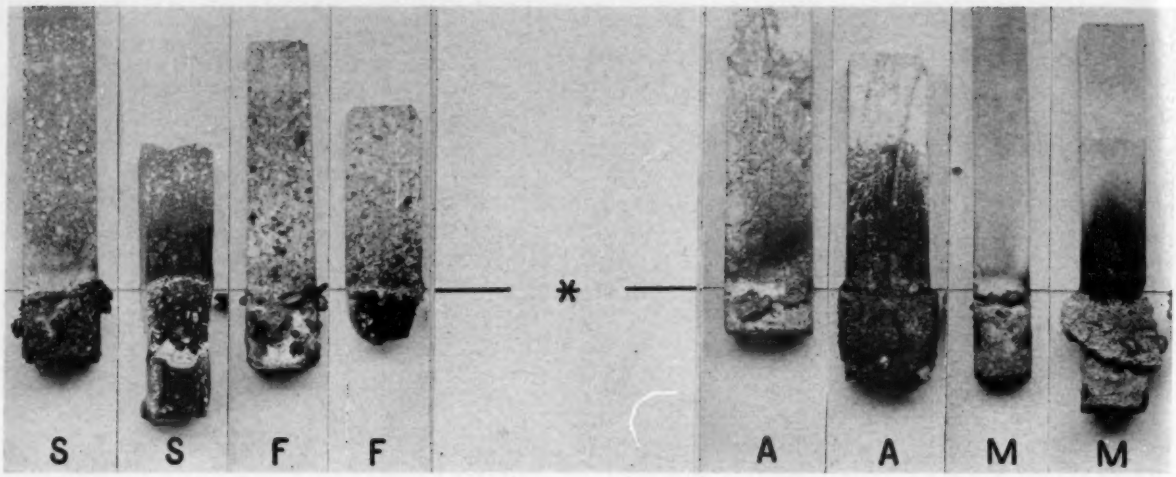
The reaction product from Ca-Mn-Si deoxidation is a glass.

## CONCLUSIONS

A test procedure for refractory-metal reactions is presented.

The conclusions concerning reactions are summarized by Fig. 4 and Table 2. These include:

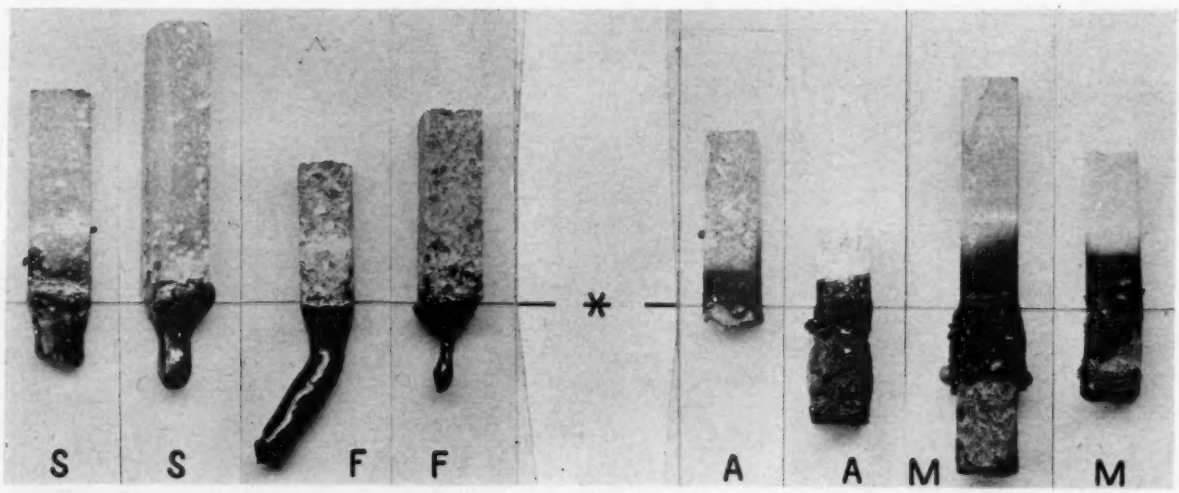
1. Aluminum produces a reaction product on fireclay refractories that is identical with the macroinclu-




| Position / Time / Test / Refractory | Left 5 min. F-46 | Right 10 min. F-43 |
|---|---|---|
| S - silica | 1 | 3 |
| F - fireclay | 1-a | 4 |
| A - alumina | 0 | 0-a |
| M - magnesia | 0 | 0-a |

Fig. 14 — Time of contact variable, temperature (2950 F), aluminum deoxidation and low manganese (<0.5%) fixed (Table 1 presents exact data). * = metal line. a = accumulation of oxidation product on sample surface.

Fig. 15 — Time of contact variable, temperature (2950 F), aluminum deoxidation and high manganese (>2.75%) fixed (Table 1 presents exact data). * = metal line. a = accumulation of oxidation product on sample surface. x = sample broke during test.




| Position / Time / Test / Refractory | Left 5 min. F-54 | Right 10 min. F-56 |
|---|---|---|
| S - silica | 4-5 | 4-5 |
| F - fireclay | 4-5 | 4-5 |
| A - alumina | 1-x | 2-a |
| M - magnesia | 1-a | 1 |

sions which cause the casting imperfections. This may temporarily protect the refractory at low temperatures until it is removed by erosion into the metal stream.

2. Manganese accentuates the fluxing reaction at the refractory surface. It does not produce corundum unless aluminum is also added to the steel.
3. Silica and fireclay refractories are both subject to severe fluxing reactions at high manganese levels (>2.75 per cent). Alumina and magnesia refractories are essentially immune to the reactions under the range of the conditions tested.

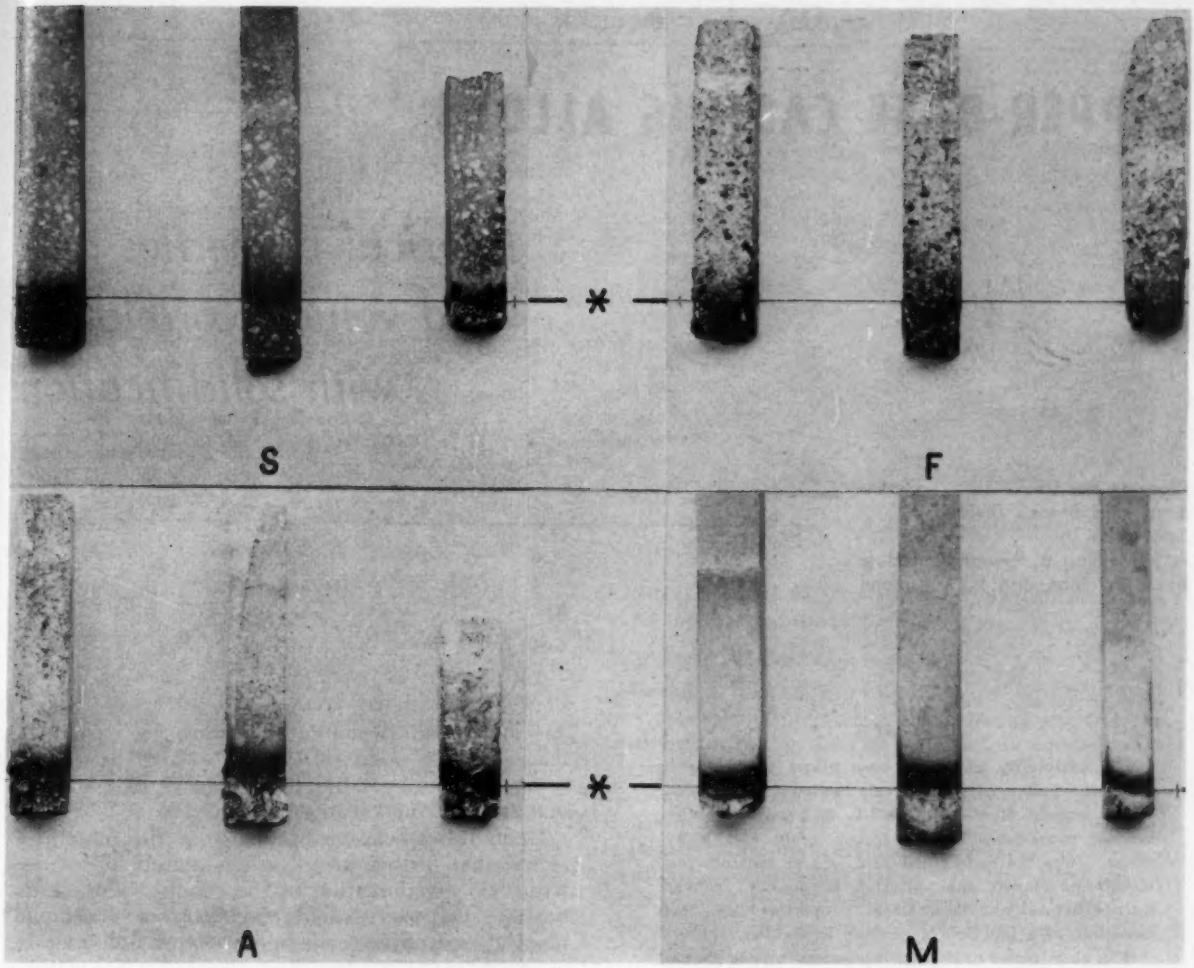



Fig. 16 — Ca-Mn-Si addition variable, temperature (2850 F), aluminum content (none) and low manganese (<0.5%) fixed (Table 1 presents exact data). Contact time — 5 min. * = metal line.

| Position / Ca-Mn-Si / Test / Refractory | Left none F-47 | Middle 3 lb/T F-48 | Right 9 lb/T F-49 |
|---|---|---|---|
| | ← Equal Mn → | | |
| S - silica | 1 | 1 | 1 |
| F - fireclay | 2 | 2 | 2 |
| A - alumina | 0 | 0 | 0 |
| M - magnesia | 0 | 0 | 0 |

4. At low manganese levels (<0.5 per cent), fireclay refractories are more subject to reaction with molten steel than are silica refractories. At high manganese levels, at 2850 F, the initial attack of the silica is greater.
5. Temperature variation in the range of 2850-2950 F has relatively minor effect upon the reactions. There is one exception. The reaction product (Conclusion 1) is fluxed more severely at higher temperatures and therefore the refractories are more easily eroded (temperatures in excess of 3000 F for fireclay and silica would alter this conclusion because these refractories could be melted without any reaction).
6. Time of contact has a minor but noticeable effect at low manganese levels. At high manganese levels, reaction is severe even in brief intervals of contact.

## ACKNOWLEDGMENT

The authors wish to acknowledge the many helpful suggestions and continued support of the A.F.S. Steel Division Research Committee: J. Rassenfoss, *Chairman*, G. A. Colligan, A. J. Kiesler, W. Koppi, E. Lange, C. Lorig, S. C. Massari, E. Punko, W. Punko and J. Zotos.

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# COPPER-BASE CASTING ALLOYS

## *Physical properties and void volume correlation with solidification*

Progress Report
Brass and Bronze Division
Sponsored by
AFS Training & Research Institute
by R. A. Flinn and H. Kunsmann

### ABSTRACT

**The castability of copper-base alloys with a wide range of liquidus-solidus intervals has been investigated. These include 80-10-10, 85-5-5-5, 88-8-0-4, aluminum bronze, manganese bronze, silicon brass and nickel silver. Using the end-chilled 12 x 2 x 2 in. test bar previously developed, the variation in density, leakage, microstructure and mechanical properties has been measured as a function of position in the bar.**

**The aluminum bronze and manganese bronze showed little variation in density, no leakage or microshrinkage and little change in mechanical properties. Surprisingly, the 80-10-10 and the silicon brass, materials with the longer liquidus-solidus intervals, also showed similarly good castability. The other mushy-type freezing alloys exhibited leakage, low density, microshrinkage and poorer mechanical properties at the stations in the bar well away from the chill. A mechanism relating the large volume of low melting point liquid in the silicon bronze and the 80-10-10 to the morphology of solidification is proposed.**

### INTRODUCTION

In previous work in this investigation sponsored by the AFS Training and Research Institute under the direction of the AFS Brass and Bronze Division Research Committee,[1] the immediate objective of the research program was to determine the conditions affecting the pressure tightness of 85-5-5-5 castings. The initial effort was directed toward this single objective because of the predominant use of this alloy and the widespread interest in the variables affecting leakage. As a result of this work it was established[1] that, to produce pressure tight sections, a strong thermal gradient (60 F/in.) was required during the final stages of solidification of the matrix. A variety of shapes, section sizes, chill and riser designs was investigated to determine the thermal gradients and soundness of the castings.

Upon the conclusion of this work the committee agreed that further investigations should follow two paths—(1) determination of the conditions of solidification necessary for development of maximum physical properties (including pressure tightness) in other copper-base alloys and (2) application of the solidification data to problem castings to illustrate their use to the foundryman. This paper is concerned only with phase one; the work dealing with problem castings is underway and will be reported at a later date.

### PROCEDURE

Most of the procedure follows the methods developed and described previously.[1] Important procedural additions were made, however.

#### *Alloys Investigated*

In order to provide a wide range of alloys with different solidification intervals these analyses were selected for the first phase of the work:

| B30 A.S.T.M. Class | Composition, % Cu | Sn | Al | Pb | Fe | Zn | Ni | Si | Metal Type | Freezing Range, F |
|---|---|---|---|---|---|---|---|---|---|---|
| 1 B | 88 | 8 | — | — | — | 4 | — | — | Modified G Bronze | 270 |
| 3 A | 80 | 10 | — | 10 | — | — | — | — | High Leaded Tin Bronze | 302 |
| 4 A | 85 | 5 | — | 5 | — | 5 | — | — | Leaded Red Brass | 260 |
| 8 A | 58 | — | 1 | — | 1 | 39 | — | — | Manganese Bronze | 30 |
| 9 B | 89 | — | 10 | — | 1 | — | — | — | Aluminum Bronze | 15 |
| 11 A | 64 | 4 | — | 4 | — | 8 | 20 | — | Nickel Silver | 250 |
| 13 A | 81 | — | — | — | — | 15 | — | 4 | Silicon Brass | |

R. A. FLINN is Prof., Met. Engrg., Cast Metals Lab. and H. KUNSMANN is F.E.F. Scholarship Student, University of Michigan, Ann Arbor.



61-103

TABLE 1 — CHEMICAL ANALYSES OF THE ALLOYS[1]

| A.S.T.M. | 3 A 80-10-10 | 1 B 88-8-0-4 | 4 A 85-5-5-5 | 8 A Mn Bronze | 9 B 89-1-10 | 11 A Ni Silver | 13 A Si Brass |
|---|---|---|---|---|---|---|---|
| Cu | 79.24 | 86.98 | 84.2 | 58.45 | 88.66 | (64.67) | 82.85 |
| Sn | 9.45 | 8.07 | 4.7 | 0.45 | — | (3.97) | — |
| Pb | 9.97 | 0.11 | 5.1 | 0.23 | — | (3.83) | Tr. |
| Zn | bal (0.6) | bal (4.20) | bal (5.08) | bal (38.67) | — | bal (6.36) | bal (13.0) |
| Fe | 0.015 | 0.08 | 0.16 | 0.92 | 1.15 | 0.92 | 0.08 |
| Al | — | — | <0.01 | 0.38 | 10.00 | — | — |
| Mn | — | — | 0.0 | 0.30 | — | — | — |
| Ni | 0.39 | 0.55 | 0.54 | — | — | (20.08) | — |
| Sb | 0.27 | — | 0.20 | — | — | 0.17 | — |
| P | — | 0.011 | 0.01 | — | — | — | 4.02 |
| Liquidus, F | 1705[3] | 1840[4] | 1830 | 1690[3] | 1915[4] | 2100[5] | 1683[3] |
| Solidus,[2] F | 1403[3] | 1570[4] | 1570 | 1660[3] | 1900[4] | 1850[5] | 1510[3] |
| Range, F | 302 | 270 | 260 | 30 | 15 | 250 | 173 |

1. Average of analyses from heats 50 and 52.
2. Solidus of $\alpha$ phase, not of lead rich eutectic.
3. From Brass and Bronze Ingot Institute data for alloys of this type.
4. From binary diagram approximation.
5. From International Nickel Co. literature.

The ingot analyses are listed in Table 1.

### *Melting Procedure*

The alloys were melted in a gas fired crucible furnace, with the exception of several induction melts. The melts were superheated to 300 F above their respective liquidus temperatures, and poured at 200 F above the liquidus. Standard deoxidation practices were followed, and no evidence of gas porosity was encountered.

### *Test Casting*

The standard 12 x 2 x 2 in. bar chilled at one end face, which had been developed in previous work, was poured in dry sand molds. In determining solidification patterns, thermocouples were placed at the centerline of the bar at 1, 3, 5, 7, 9 and 11 in. from the chilled end. A six point recorder with a one sec full scale response was employed.

In order to supplement the thermocouple data a number of castings were drained at a predetermined time during solidification. This was accomplished by tilting the mold 90 degrees and allowing the liquid to escape through the riser.

### *Sectioning of Test Castings*

**Leakage test.** A longitudinal slice $\frac{1}{32}$ x 2 x 12 in. was machined from the center plane of the bar for pressure testing, as in Fig. 1. The leak rate of compressed air at 100 psi through a ½-in. diameter orifice held at the centerline was determined at different locations along the bar.

**Density.** Density specimens were taken at various locations, as shown in Fig. 1. In all cases the surfaces were ground free of sand and scale. The density was determined using an analytical balance and distilled water containing aerosol.

**Tensile specimens.** These specimens were cut just beneath the center plane of the bar, as shown in Fig. 1. The dimensions of the specimen are shown in Fig. 2.

**Microspecimens.** The locations of the specimens are shown in Fig. 1. The specimens were mounted in bakelite and polished using metallographic papers and diamond paste. Because of the necessity of dis-

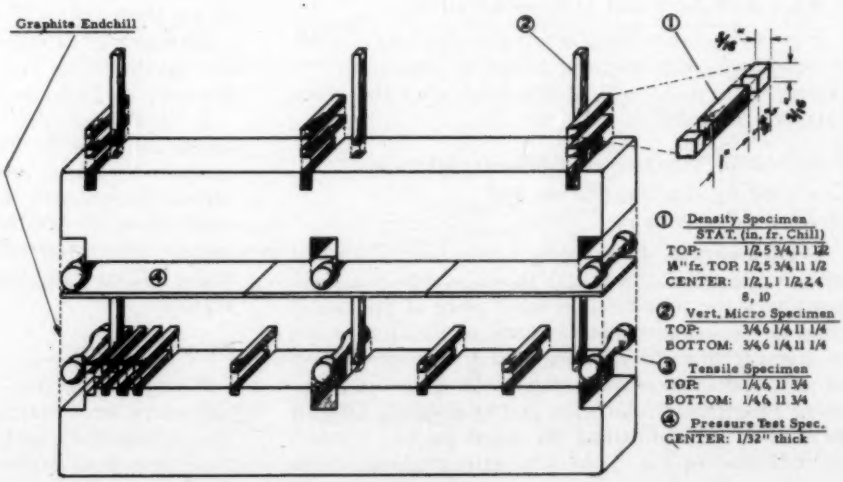



Fig. 1 — Exploded view of 2 x 2 x 12 in. standard test bar to indicate specimen locations.

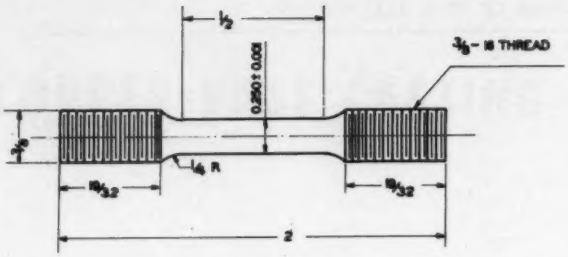



Fig. 2 — 0.250 in. tensile specimen.

tinguishing between lead and shrinkage voids, special care was taken to use low nap polishing wheels and a minimum of fluid in polishing.

## EXPERIMENTAL DATA

Because of some of the complex relations among the different sets of data, it is advisable to present the data separately under these divisions and then discuss their interrelations:

1. Variation in physical properties including pressure tightness as a function of position in the test bar.
2. Solidification patterns of the different alloys as disclosed by cooling curves and drainout experiments.
3. Metallographic examination.

### *Variation in Properties as a Function of Position in the Test Bar*

The data for all the alloys are summarized in Fig. 3. By employing a common abscissa for mechanical properties, density and leak rate the following general relations are made evident. The compositions may be divided into two principal groups according to the results:

1. No leakage, no appreciable variation in density and only minor variations in mechanical properties along the test bar. These alloys are—3 A—80-10-10, 8 A—Mn bronze, 9 B—Al bronze and 13 A—Si brass.
2. Leakage in portions of the test bar away from the chill and corresponding decreases in density and mechanical properties. These alloys are—1 B—88-8-0-4, 4 A—85-5-5-5 and 11 A—nickel silver.

It is obvious that there is not the usual correlation of soundness with freezing range as commonly expected. This point will be discussed after the other data are presented.

### *Solidification Patterns of Different Alloys as Disclosed by Cooling Curves and Drainout Experiments*

**Cooling curves.** The cooling curves may be divided into two general classes—(1) those in which a large portion of the solidification takes place at practically constant temperature, particularly in the final stages of freezing and (2) those in which a large portion of the solidification takes place over a wide temperature range, particularly in the final stages. On the basis of this classification, the alloys may be divided, as indicated in Fig. 5, by schematic cooling curves:

*Class 1.*
(3 A—80-10-10).
8 A—Mn bronze.
9 B—Al bronze.
13 A—Si brass.

*Class 2.*
1 B—88-8-0-4.
4 A—85-5-5-5.
11 A—nickel silver.

It will be noted that in this arrangement the alloys of Class 1 are the same as those showing no leakage, while those of Class 2 show a decrease of properties away from the chill (placing the 3A alloy in Class 1 is controversial, as discussed later, since the data for the freezing of the lead-tin phase are not available).

Cooling curves provide interesting data relevant to solidification morphology, but do not generally indicate the time for initiation of freezing at given thermocouple locations. Consider, for example, the beginning of the arrest at the No. 3 thermocouple in the 89-1-10 bar, Fig. 4c. At this location the heat transfer is essentially unidirectional because of the nearness of the end chill. When this station reaches the liquidus temperature there is a thermal block to heat transfer, because the metal between it and the chill is also at the liquidus. The No. 2 couple, for example, is also at the liquidus temperature and does not complete freezing for 120 sec.

Only after the temperature at No. 2 falls below the liquidus can heat transfer, and hence freezing take place at No. 3. Another procedure was therefore used to provide additional information as described in the drainout experiments.

**Drainout experiments.** To supplement the cooling curve data, it was decided to determine the flowability of the metal at a given time at different distances from the solidifying interface. In other words, it was hypothesized that if, in a given alloy, fluid metal existed close to the finally solidifying interface this would lead to a sound surface compared to the case where an extensive mushy region extended between the interface and fully fluid metal. The word "fluid" is used rather than 100 per cent liquid, since the drainout does not distinguish between these.

Photographs of typical bars from these experiments are presented in Fig. 6. These data may be summarized, as indicated in Table 2.

It is evident that only two of the four alloys listed as Class 1 show smooth front freezing as compared with mushy freezing. In the 89-1-10 aluminum bronze and the manganese bronze there was only 0.3 in. of dendritic material ahead of the completely solid material. By contrast the other alloys in both Class 1 and Class 2 showed much wider mushy regions.

### *Microstructures*

To corroborate the density and leakage data, microspecimens were examined from the important centerline locations. In addition, a survey of microstructure was made from cope to drag surface at the important

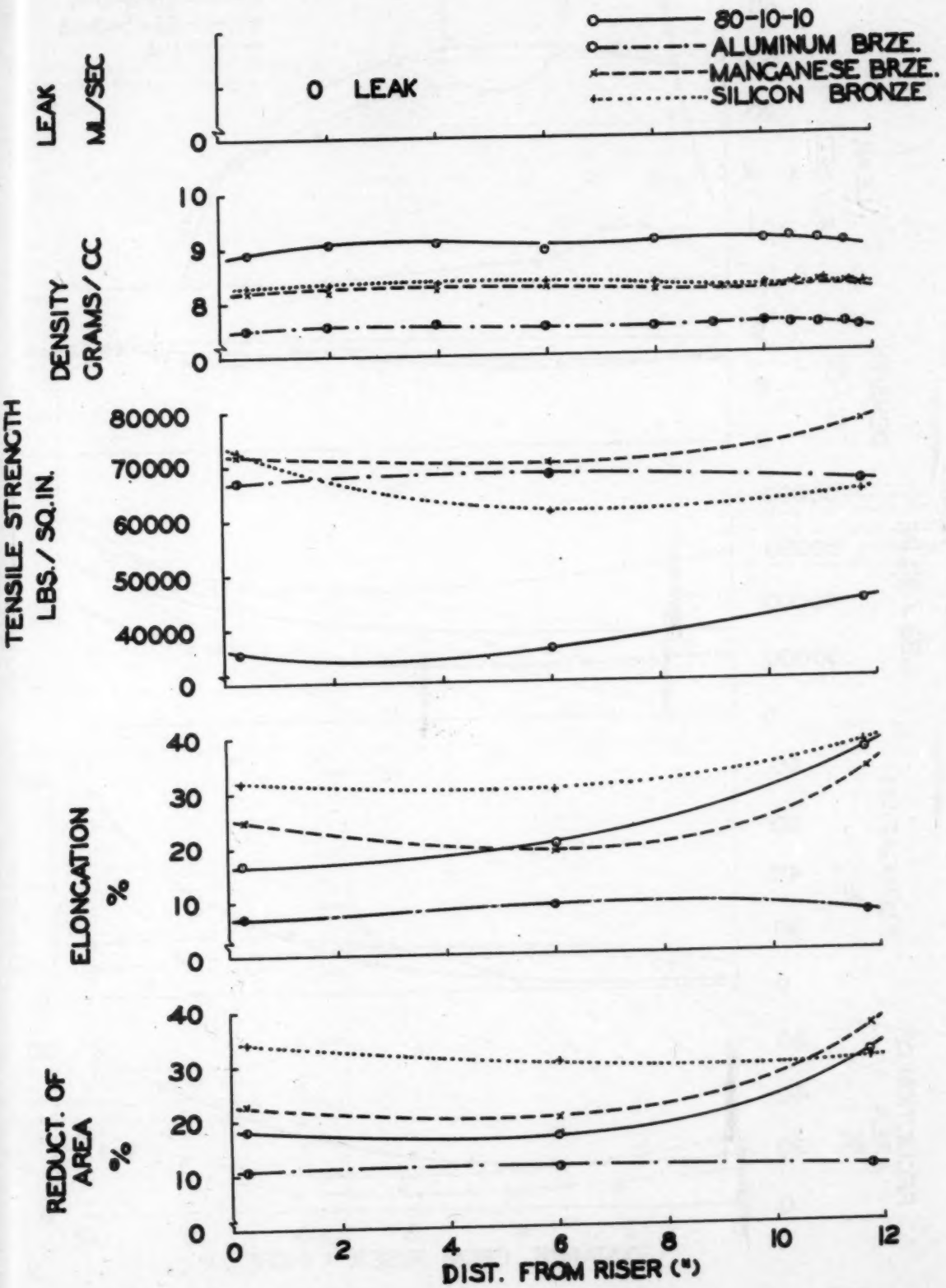



Fig. 3a — Mechanical properties of the non-leaking alloys group.

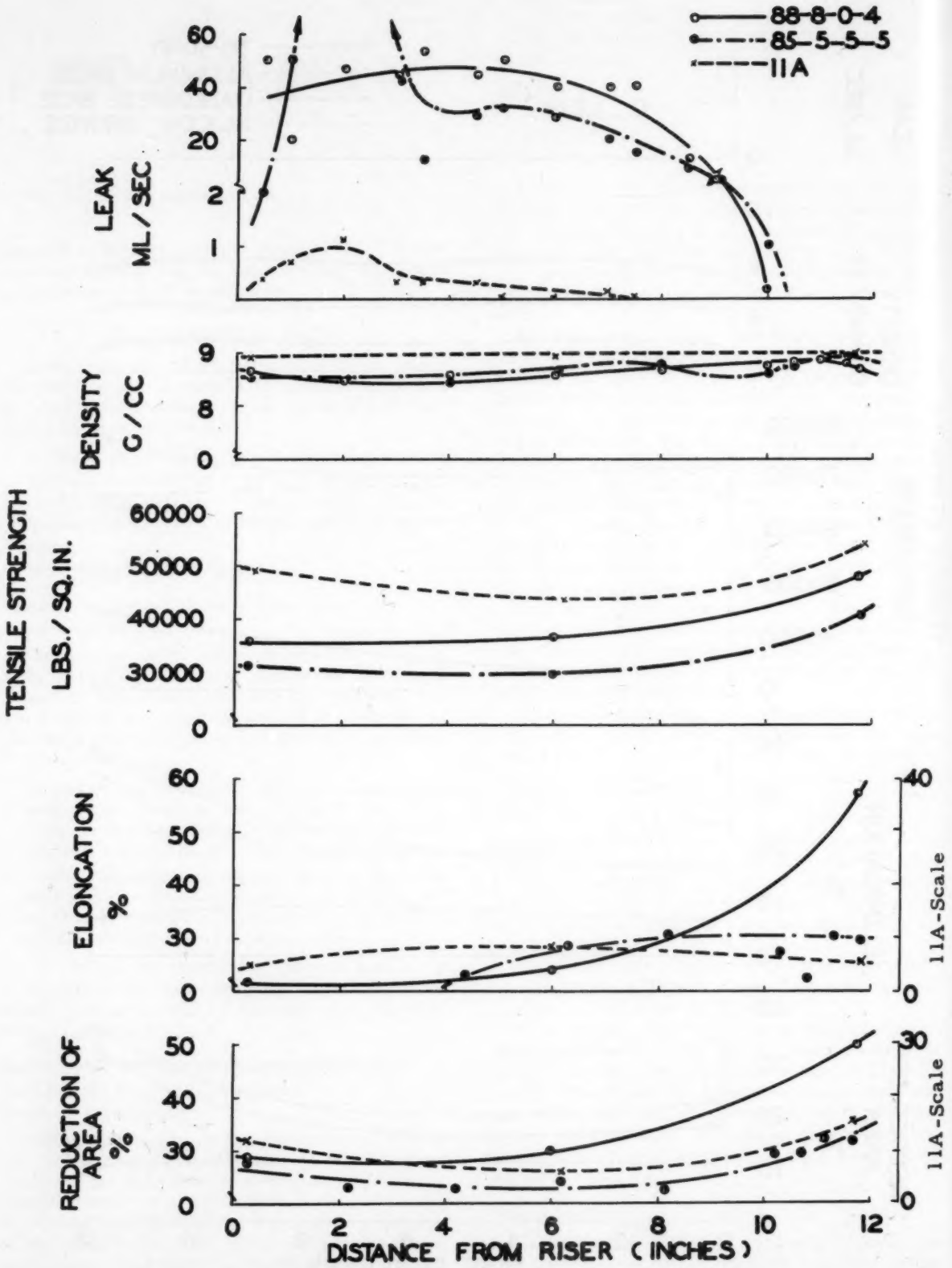



Fig. 3b — Mechanical properties of the alloy group exhibiting leakage.

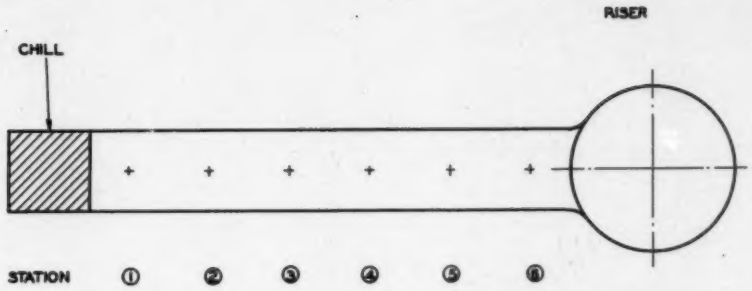



Fig. 4a — Standard test bar showing thermocouple station arrangement.

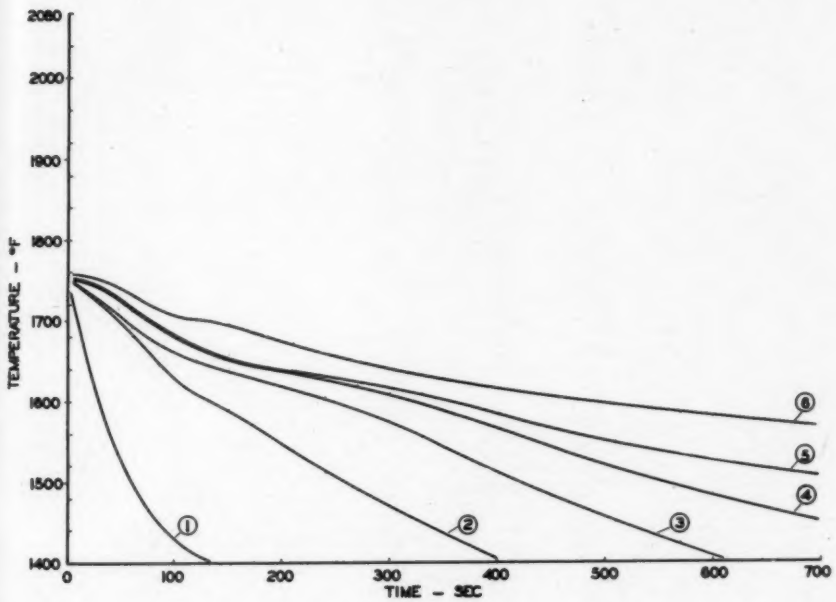



Fig. 4b — Time-temperature cooling curves for 80-10-10 (3A) alloy. Induction furnace, P-Cu deoxidation, 1905 F pouring temperature, dry sand, end chill.

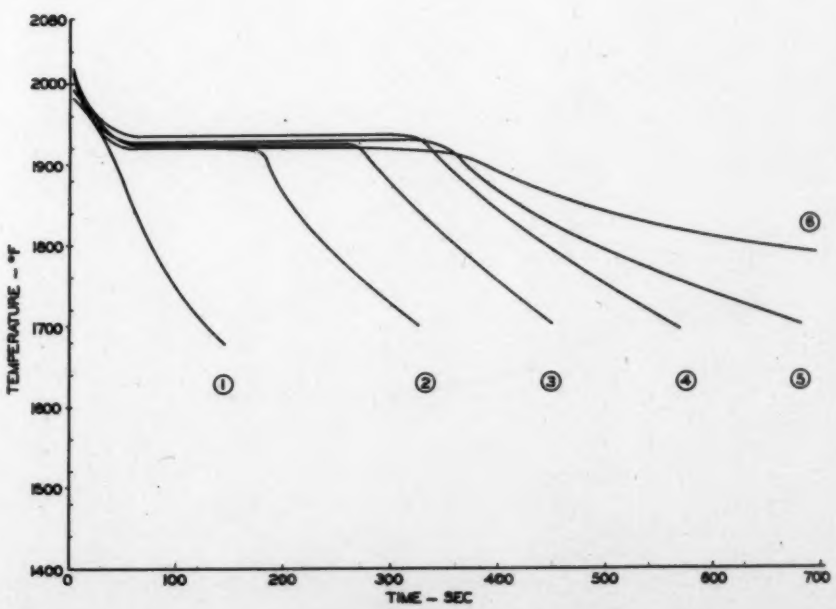



Fig. 4c — Time-temperature cooling curves for aluminum bronze (9B). Induction furnace, no. 77 alloy deoxidation, 2115 F pouring temperature, dry sand, end chill.

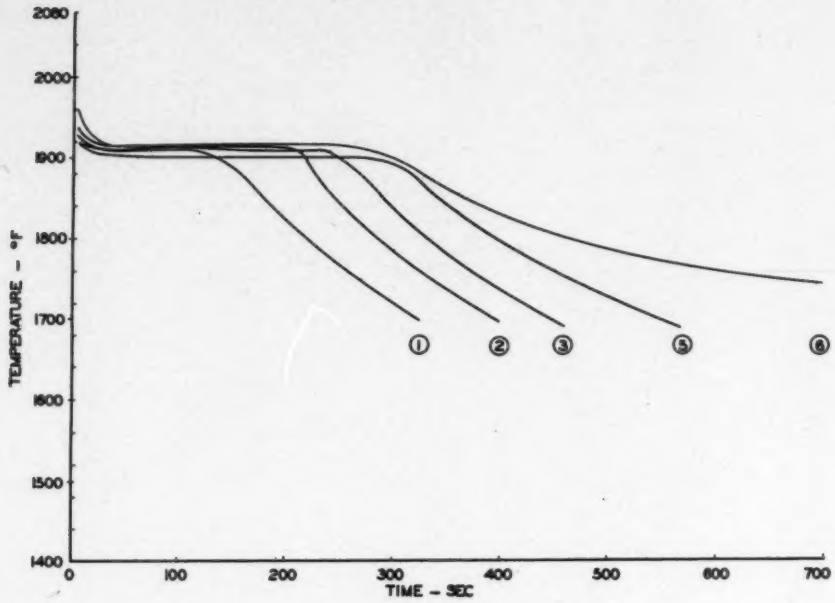



Fig. 4d — Time-temperature cooling curves for aluminum bronze (9B). Induction furnace, no. 77 alloy deoxidation, 2115 F pouring temperature, dry sand, no chill.

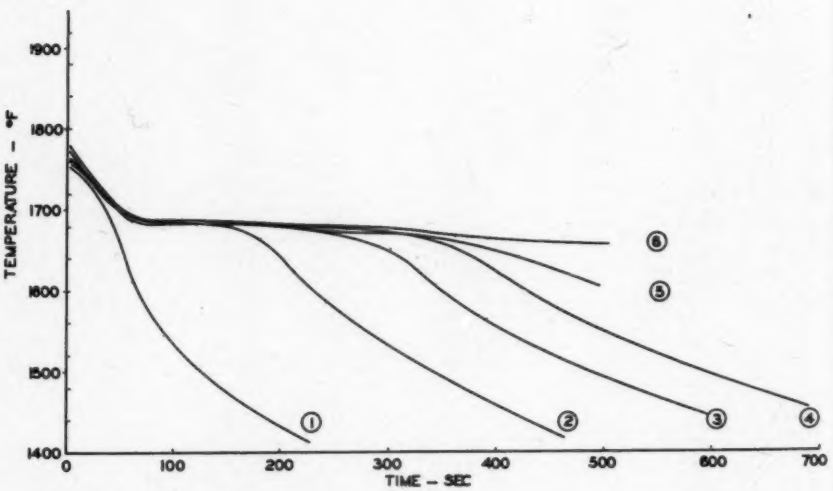



Fig. 4e — Time-temperature cooling curves for manganese bronze (8A). Induction furnace, no deoxidation, 1890 F pouring temperature, dry sand, end chill.

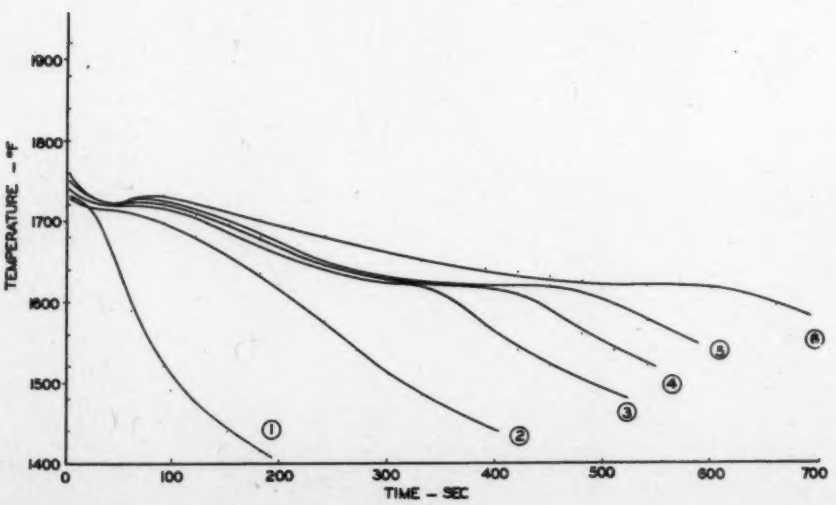



Fig. 4f — Time-temperature cooling curves for silicon bronze (13A). Induction furnace, no deoxidation, 1885 F pouring temperature, dry sand, end chill.

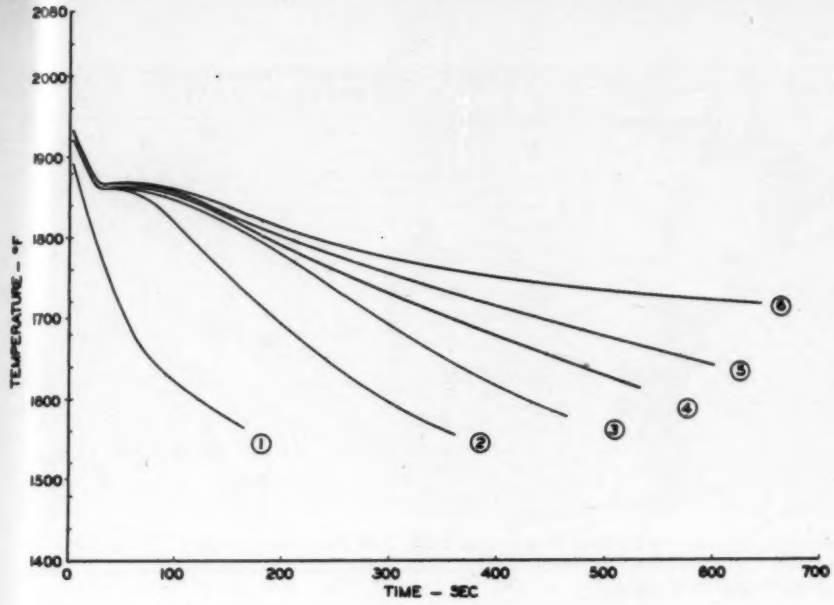



Fig. 4g — Time-temperature cooling curves for 88-8-0-4 (1B). Induction furnace, P-Cu deoxidation, 2050 F pouring temperature, dry sand, end chill.

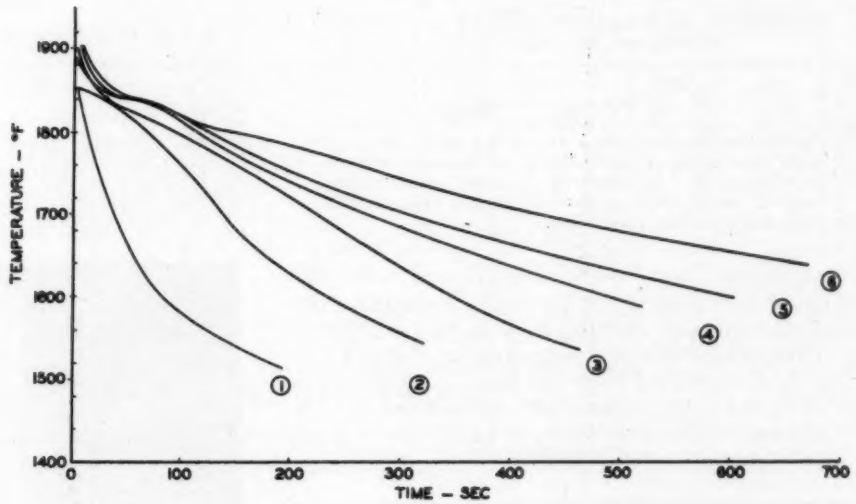



Fig. 4h — Time-temperature cooling curves for 85-5-5-5 (4A). Gas furnace, P-Cu deoxidation, 2025 F pouring temperature, dry sand, end chill.

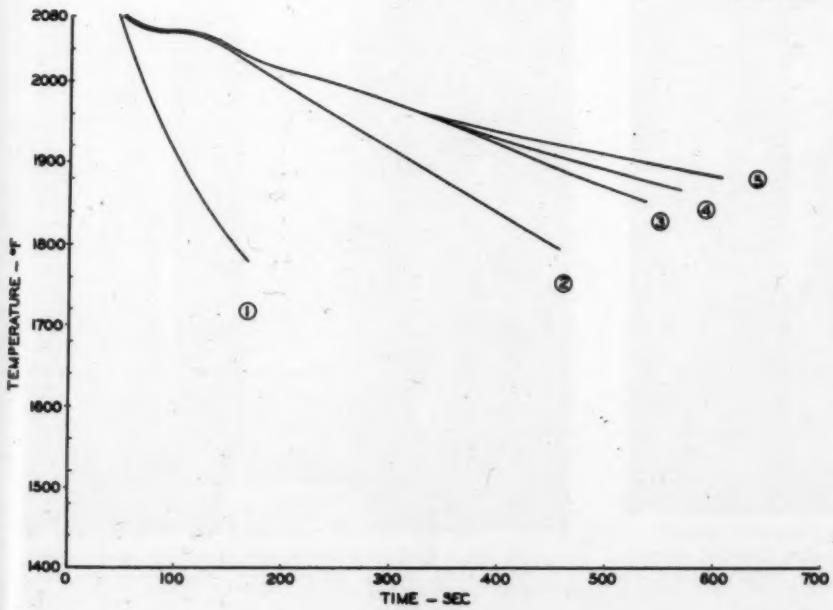



Fig. 4i — Time-temperature cooling curves for nickel silver (11A). Induction furnace, no. 66 alloy plus Mg sticks deoxidation, 2300 F pouring temperature, dry sand, end chill.

Figure 4 I

Time Temperature Cooling Curves

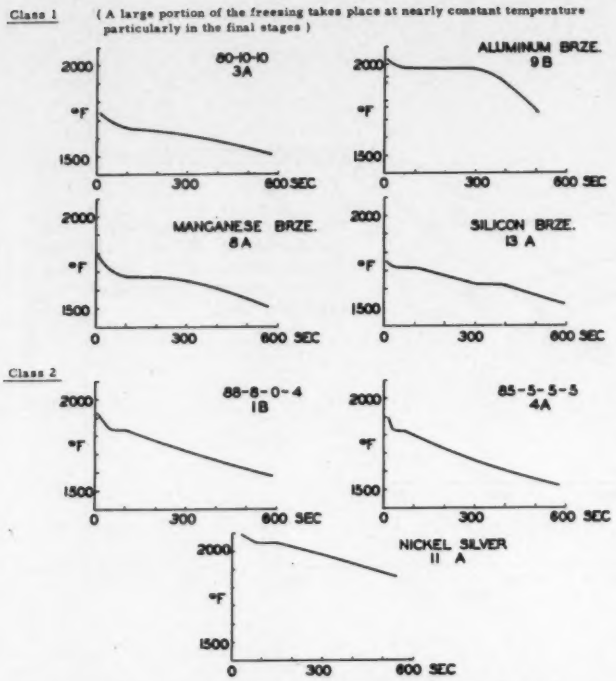



Fig. 5 — Schematic cooling curves for alloys showing differences in final solidification characteristics. Data from station 4, 5 in. from riser. A large portion of the freezing takes place at nearly constant temperature, particularly in the final stages.

locations, i.e., next to the riser (11¼-in. from the chill), at the center of the bar (6¼-in. from the chill) and near the end chill (¾-in. from the chill face).

These results may be summarized in Table 3 and Fig. 7. It is evident that the Class 1 alloys, 3 B — 80-10-10, 8 A — Mn bronze, 9 B — Al bronze and 13 A — Si brass are free from microshrinkage in the specimens examined at 100 diameters.

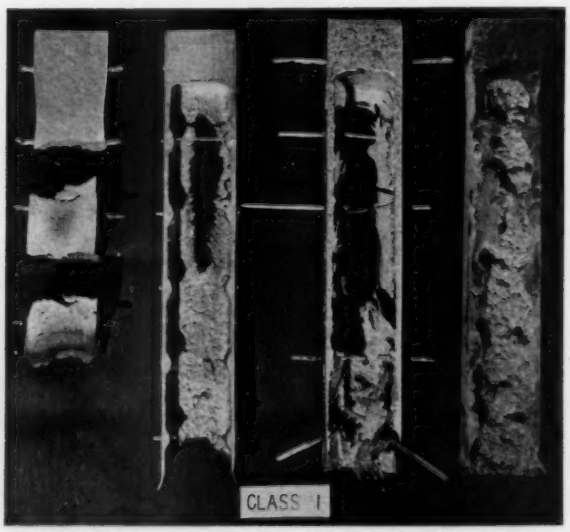



Fig. 6a — Class one drainout results. *Left to right* — 80-10-10 (3A), 30 sec; aluminum bronze (9B), 60 sec; manganese bronze (8A), 60 sec; silicon bronze (13A), 60 sec.

TABLE 2 — DRAINOUT CHARACTERISTICS OF COPPER-BASE ALLOYS

| Alloy | 3 A 80-10-10 | 1 B 88-8-0-4 | 4 A 85-5-5-5 | 8 A Mn Bronze | 9 B 89-1-10 | 11 A Ni Silver | 13 A Si Brass |
|---|---|---|---|---|---|---|---|
| Time of Drainout (sec. after pouring) | 30 | 30 | 30 | 60 | 60 | 120 | 60 |
| Distance from chill to point of 100% drainout | 1.6 | 1.75 | 2.5 | 1.4 | 1.4 | 6.9 | |
| Distance from chill to point of 100% solid | «1 | <1.0 | <1 | 1.1 | 1.1 | 1.0 | 4.0 |
| Mushy zone | »0.6 | »0.75 | >1.5 | 0.3 | 0.3 | 5.9 | 2.4 |

## DISCUSSION

Before taking up the principal points of the discussion it may be generally helpful to recall some of the predictions made at the beginning of this investigation. This is done not to discredit the predictions, which represent present metallurgical thinking, but to guide the reader more quickly to the interesting anomalies in the data.

At the beginning of this year's work, it had been

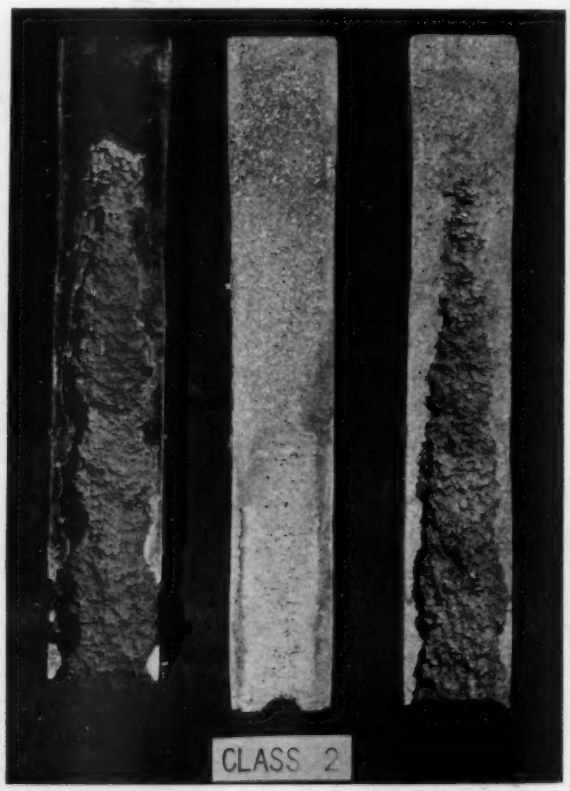



Fig. 6b — Class 2 drainout results. *Left to right* — 88-8-0-4 (1B), 30 sec; nickel silver (11A), 120 sec; 85-5-5-5 (4A), 30 sec.

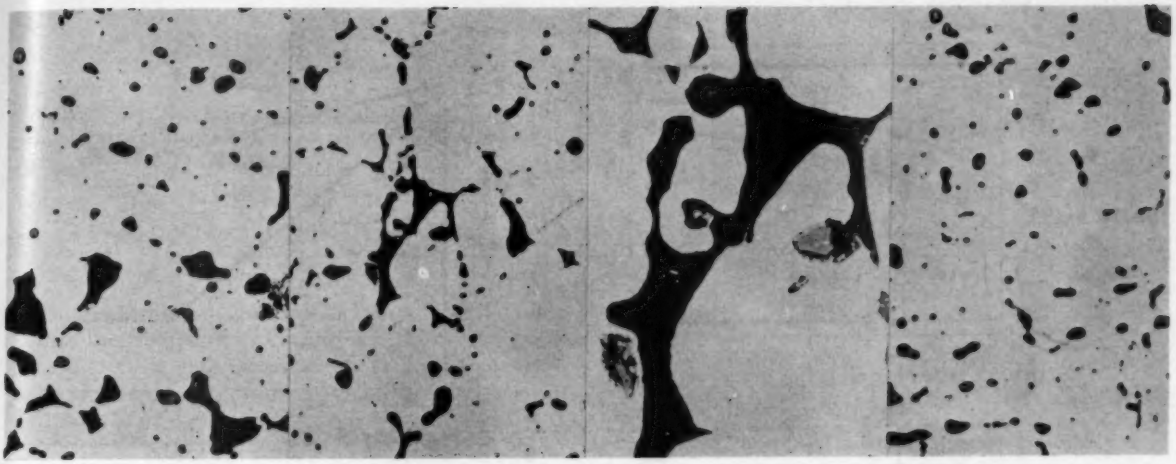


Fig. 7 — Specimen 3007. Microshrinkage in 85-5-5-5 bar at ½-in. from riser and at designated location. *Left to right* — ¼-in. from cope surface (shrink), 100 ×; at center of bar (shrink), 100 ×; center of bar (shrink), 500 ×; ¼-in. above drag surface (no shrink), 100 ×.

established that a certain thermal gradient (60 F/in.) at the solidus was required to insure soundness in 85-5-5-5 alloy castings. It seemed natural then to investigate other copper-base alloys using the same test bar and determine a simple number, the thermal gradient, required to produce sound castings in these other analyses. A representative group was selected with different liquidus-solidus intervals ranging from 15 to 302 degrees. It was generally predicted that the alloys with a short interval would require a smaller gradient while others, such as 80-10-10 and silicon brass with longer intervals, would probably give unsound bars in all cases.

The data just presented show, on the contrary, that the two alloys with relatively long freezing ranges produced bars as sound as those with the shortest freezing ranges. It is evident therefore that some new thought is required.

To summarize the data in another way, there are four alloys in Class 1 which show no leakage or microshrinkage and exhibit little variation in mechanical properties at various locations in the test bar. Although the liquidus-solidus distance varies considerably in these alloys, they all have the common feature of a flat cooling curve at final solidification (assumed in the case of 80-10-10). Apparently two different mechanisms are needed to explain the soundness of the aluminum and manganese bronzes with narrow freezing ranges compared to the silicon brass and the 80-10-10 alloy.

### *Mechanism 1 — Soundness in Short Freezing Range Alloys*

To explain the soundness of these alloys it is useful first to review briefly the role of constitutional supercooling in the morphology of solidification. It is necessary to discuss the freezing of pure metals first and then of alloys.

To summarize the conditions prevailing during the solidification of a pure metal, let us now review what occurs when a metal of freezing temperature $T_E$ freezes when poured into a mold. If poured

TABLE 3 — MICROSHRINKAGE AT DIFFERENT LOCATIONS IN TEST BAR

| Alloy[1] | | 3 A 80-10-10 | 1 B 88-8-0-4 | 4 A 85-5-5-5 | 8 A Mn Bronze | 9 B 89-1-10 | 13 A Si Brass |
|---|---|---|---|---|---|---|---|
| riser end | C | O | X | X | O | O | O |
| | C.L. | O | X | X | O | O | O |
| | D | O | O | O | O | O | O |
| mid-section | C | O | O | O | O | O | O |
| | C.L. | O | X | X | O | O | O |
| | D | O | O | O | O | O | O |
| chill end | C | O | O | O | O | O | O |
| | C.L. | O | O | O | O | O | O |
| | D | O | O | O | O | O | O |

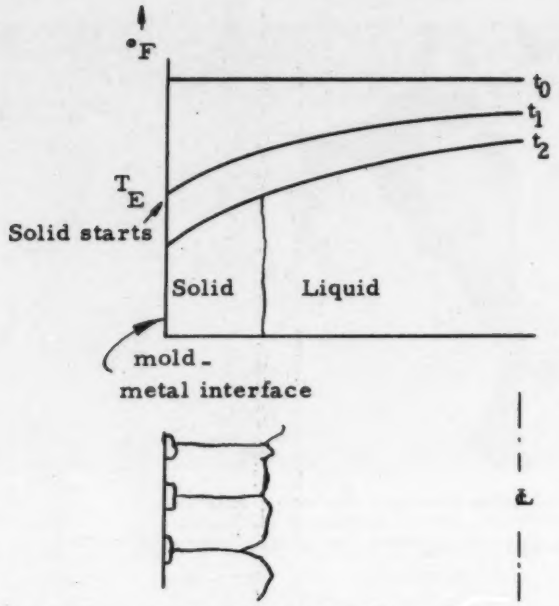
1. C = ¼-in. beneath cope surface.
C.L. = at centerline.
D = ¼-in. above drag surface.
X = Microshrinkage present.
O = No microshrinkage present.

rapidly there is no appreciable gradient at time = $t_o$ (Fig. 8).

Due to the heat transfer at the interface, a gradient develops in the metal. The interface temperature falls slightly below $T_E$ and random crystallization begins at $t_1$ at the interface (heterogeneous nucleation, Fig. 8a). To digress a moment, it should be recalled here that there is a direction of preferred growth in a crystal. A familiar example is the growth of ice crystals on a window pane.

Returning now to the metal, those nuclei which happen to be formed with the preferred growth direction perpendicular to the interface grow more rapidly than crystals of other orientations. This leads to the growth of columnar grains to the center of the ingot. The final structure therefore consists of a thin layer of randomly oriented grains at the mold surface and columnar grains extending to the centerline. There are no dendrites or randomly oriented grains at the center of the ingot.

**Nucleation and growth in alloys.** The three principal differences between solidification of pure metals and their alloys are:




Figs. 8a and 8b — Solidification of a pure metal.

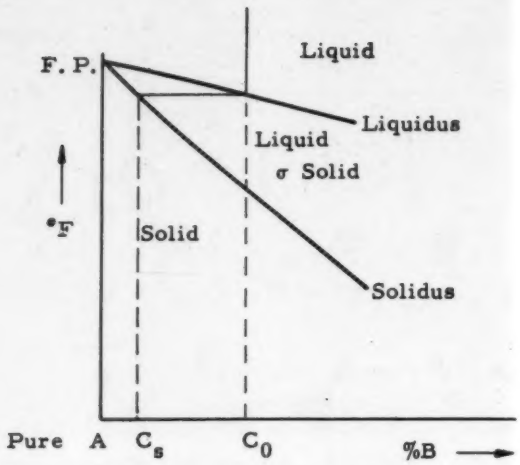



Figure 9

Schematic Phase Diagram

Fig. 9 — Schematic phase diagram.

1. Usually freezing occurs over a temperature range.
2. The composition of the solid which separates first is different from that of the liquid.
3. There may be more than one solid phase crystallizing from the liquid.

**Solid solution alloys.** If we dissolve metal *B* in liquid metal *A* to form a liquid alloy, we can describe the crystallization by a phase diagram, as shown in Fig. 8b. In most cases the freezing point of the pure metal is depressed, as shown by the liquidus line. In addition the alloy, of composition $C_0$ for example, does not freeze at a single temperature but instead freezes over a temperature range.

Another important effect is that as the metal solidifies the composition of the solid is not the same as that of the parent liquid, but is richer in metal *A*. We shall discuss now how these effects lead to a solidification pattern different from those described for the pure metal.

If we begin with a homogeneous melt of composition $C_0$, the first crystals to precipitate are of analysis $C_s$ (Fig. 9). If solidification is fairly rapid and no diffusion occurs, the liquid at the interface becomes richer in the solute *B* than the liquid away from the interface. The variation of per cent *B* with distance is shown in Fig. 10.

**Freezing mechanism modification.** Let us see how these effects modify the freezing mechanism compared with that of a pure metal.

Assume the mold is filled rapidly, as in the previous case, giving no temperature gradient at the start where time $t = 0$ (Fig. 11). Consider the liquidus temperature as $T_E$. As previously mentioned, we establish a thermal gradient and supercool somewhat before crystallization begins at time $t_1$. Consider the situation at $t_2$ (after appreciable solid has separated)—a temperature gradient similar to that of the pure metal is assumed.

The difference between this situation and that of the pure metal is that the liquidus temperature of the remaining liquid now varies with distance from the interface, as shown by the dashed line. Near

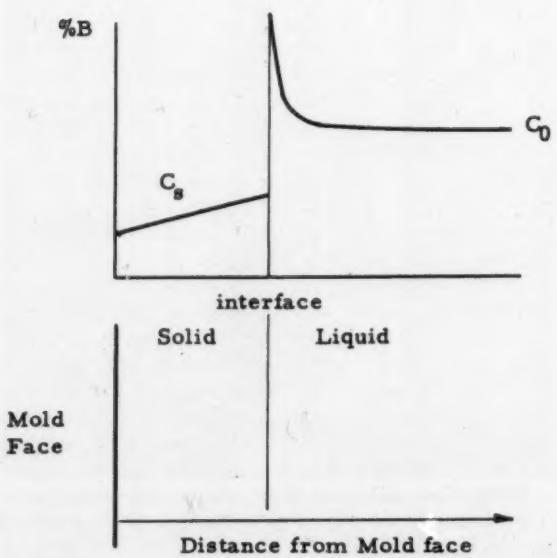



Fig. 10 — Variation of per cent B with distance.

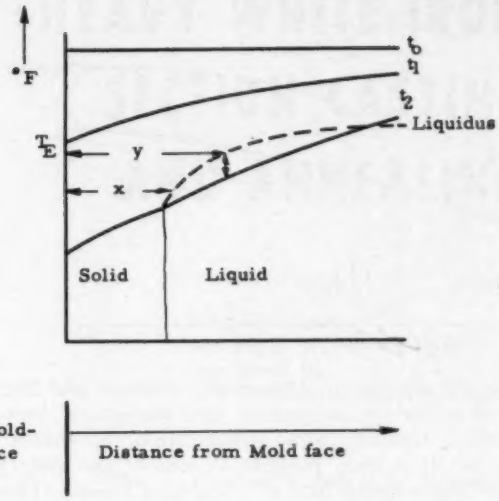



Fig. 11 — Temperature conditions during constitutional supercooling. Solid lines show actual temperature in the mold at $t_0$, $t_1$ and $t_2$.

the interface, due to buildup of component *B* in the liquid, the freezing temperature is considerably lower than $T_E$. Well away from the interface the equilibrium freezing temperature is still $T_E$, assuming no diffusion.

Notice now that a region of liquid at point *x* has a higher freezing point than the liquid at the interface. It is true that the actual metal temperature at *x* is higher than at the interface. However, the difference in the liquidus temperature between metal at *x* and metal at the interface is greater than the difference between their actual metal temperatures. Therefore, the metal at *x* is said to be constitutionally supercooled. The effects of this supercooling upon the crystallization are of three types, depending upon the degree of supercooling:

1) If there is only minor supercooling, certain preferred regions of the interface will protrude into the undercooled region and, once started, will grow more rapidly than neighboring regions, Fig. 12. This will happen both because of the greater driving force for freezing in the supercooled region and because these spikes will reject solute at their sides, delaying freezing of the side regions. These spikes result in formation of a regular honeycomb structure.
2) If supercooling is greater, the spikes tend to form side arms resulting in a dendritic structure, Fig. 13.
3) Finally, in the case of maximum supercooling, the driving force, $T_{liquidus}$-$T_{actual}$, which is a maximum at distance *y*, (Fig. 11) can lead to independent crystallization. In this way randomly oriented (equiaxed) grains are encountered. A case of this type is shown in Fig. 14. It should be mentioned that the relative solubility of the solute is shown by the factor *k* where:

$$k = \frac{C_s}{C_o}$$

*k* can be 0.5, as shown in the sketch to $10^{-2}$. When *k* is low, and the thermal gradient low, even the metal at the center of the casting can be undercooled.

It should be noted on the other hand that the greater the thermal gradient, the greater the possibility for columnar growth. This avoids random crystallization ahead of the freezing interface and facilitates metal flow for feeding of solidification shrinkage. In other words, the greater the thermal gradient, as in chilled castings, the simpler the feeding problem.

**Mass feeding.** There is one other effect called mass feeding which should be mentioned for completeness. In a section in which equiaxed crystals are nucleated randomly, the tendency of these crystals is to sink because of the greater density of the solid. This effect results in less porosity at the bottom portion of a section than in the upper part.

From this discussion it is evident that the longer the liquidus solidus distance and the greater the distribution coefficient *K*, the more likely is the formation of random equiaxed crystals away from the freezing front. This is particularly noticeable in the Class 2 alloys such as 85-5-5-5 (Fig. 6b). By con-

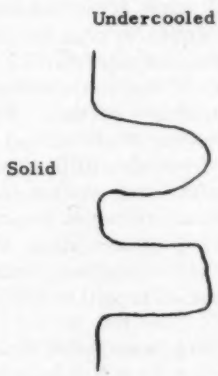



Fig. 12 — Minor supercooling effect.

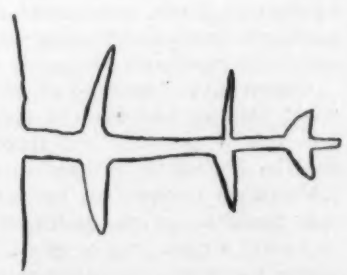


Fig. 13 — Spike formation due to progressively severe supercooling conditions.

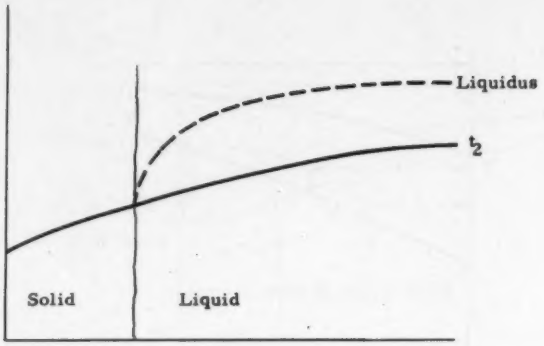



**Fig. 14 — Liquidus and actual temperature variations as a function of distance from the mold face during maximum supercooling conditions.**

trast, the shorter liquidus-solidus distance and higher $K$ value for the aluminum and manganese bronzes promote smooth front solidification, as shown in Fig. 6a. It is even possible to remove the end chill and obtain sound bars of aluminum bronze. This is because the sand cast end surface provides enough end cooling to develop directional solidification.

*Mechanism 2*

It is now necessary to consider another explanation for the soundness of the 80-10-10 and the silicon bronze which are admittedly long freezing range alloys. It is proposed tentatively that although the primary crystallization in both these materials proceeds by formation of a network of copper rich $\alpha$ crystals, the final stages of solidification are similar to the narrow range liquids just discussed.

**Silicon bronze.** In the case of the silicon bronze, the cooling curves are quite different from the alloys which exhibit shrinkage, such as 85-5-5-5. The marked difference is in the long flat portion of the curve at the period of final solidification. This indicates that a substantial amount of liquid is present. The cooling curves at the different stations drop off from this final plateau just as those of the narrow range materials drop off from the single plateau. This signifies there is an orderly movement of the end of freezing wave down the bar from the chill to the riser, and that there is no constitutional supercooling.

It is assumed, therefore, that after the original $\alpha$ network has formed there is adequate space and time for orderly motion of the remaining liquid toward the chilled end of the bar, as required to avoid shrinkage.

The microstructure of the silicon brass shows clearly the primary $\alpha$ crystals surrounded by the lower melting point material. It is proposed that the same mechanism leads to a sound structure in 80-10-10 although thermocouple data were not obtained for the low temperature range.

**Copper-lead.** The copper-lead phase diagram shows slight solubility of lead in the solid copper phase, and this is probably not increased appreciably by the tin. Therefore, at least 10 per cent of the total alloy (lead) is liquid to low temperatures. Data are not available for the partition of the tin between the solid copper rich $\alpha$ phase and the liquid lead phase. However, since the maximum solubility of the tin in the $\alpha$ is only 10 per cent, and unlimited in liquid lead, it is possible that at least half of the tin should be dissolved in the liquid lead phase. In other words the bar may contain 15 per cent liquid to rather low temperatures.

For example, the freezing of a 66 per cent lead 34 per cent tin alloy (following the proportion of 10 Pb 5 Sn) would result in the freezing of an $\alpha$ Pb phase from 600 F to 361 F and of the Pb-Sn eutectic at 361 F. Under the conditions assumed, half of this lead alloy would crystallize at the eutectic. Probably the tin content of the lead alloy would be greater, and therefore the eutectic amount would be greater. To determine if the lead content alone was responsible for the better density of the 80-10-10, a 5 per cent lead addition was made to the 85-5-5-5 alloy giving approximately 80-5-10-5. This alloy showed some improvement over the 85-5-5-5. Further critical experiments are obviously indicated.

## PRACTICAL APPLICATIONS

While this paper constitutes a progress report, and further testing is required, the following applications may be suggested:

When risering is possible, aluminum bronze and manganese bronze provide maximum strength and pressure tightness. From present indications silicon bronze and 80-10-10 also produce sound castings. The thermal gradients required to produce sound castings in these alloys are less than for 85-5-5-5. In intricate castings where risering is difficult, the higher tin-lead ratios may be of value in that selected regions may be made pressure tight by chilling.

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## ACKNOWLEDGMENT

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# HEAVY WHITE IRON SECTION CASTING AND ANNEALING

Progress Report
Malleable Division Research
Sponsored by
AFS Training & Research Institute

by R. W. Heine, T. W. Mueller and J. W. Widmeyer

## ABSTRACT

This is one of the research project reports on casting heavy sections in white iron and then annealing them to malleable iron. Chemical analyses of iron and additions to the melt for casting heavy sections as white iron with no mottle spots on a fractured surface, and time-temperature requirements for annealing the heavy sections to produce malleablized iron were studied.

A drastic decrease in the number of graphite nodules developed during annealing as the as-cast section increased. When Bi, Te or Ce are used for prevention of mottling, there is no improvement in the number of nodules during annealing. Those factors which are known to increase nodule number in lighter sections have less effect in heavy sections. Increasing the silicon percentage produces an increase in nodule number.

Cerium is revealed as another element capable of producing white fractures at C and Si percentages where mottle or gray fractures would normally occur.

## INTRODUCTION

The third annual report of a research project aimed at casting heavy sections in white iron and then annealing them to malleable iron is presented herewith. The project is sponsored by the AFS Training & Research Institute under the direction of the AFS Malleable Division at the University of Wisconsin. Subjects selected for study were:

1. Chemical analyses of iron and additions to the melt for casting heavy sections as white iron so that no mottle spots appear on a fractured surface.
2. Temperature-time requirements for annealing the heavy sections to produce a malleablized iron.

References 1, 2 and 3 present the results of work completed earlier in this project, comprising the first two annual progress reports.

## CASTING WHITE

Figure 1, from a reference[1] shows the per cent carbon and per cent silicon limits for white vs. mottled fractures under various conditions of melting and metal additions for a 4 x 4 x 8 in. block casting risered on one end. Details of melting practice involving 100 lb induction furnace heats are described in references,[1,2] as are the methods of making additions. Figure 1 shows the base composition curve for melting under a dry atmosphere, and also for melting under an atmosphere containing 12 grains of moisture/cu ft of gas. Also shown on Fig. 1 are composition limit curves for additions of Bi and Te to the melt. Compositions above the respective curves are mottled, and below the curves are white.

Composition limit curves for a 1⅞-in. D Bar, a 2 x 2 x 8 in. bar, and 3 x 3 x 8 in. bar are shown in Fig. 2. The curve for the 1⅞-in. D bar was also presented.[1] A 2 x 2 x 8 in. D bar should have about the same freezing time as a 1⅞-in. D bar. Therefore, the carbon and silicon limits for mottling are expected to be about the same. Heats were made to compare mottling in these two bars and also the 3 x 3 x 8 in. bar with the earlier work for the 4 x 4 x 8 in. bar. Chemical analyses and fracture data for these heats are reported in Table 1.

The points are plotted on Fig. 2, and show that the 2 x 2 x 8 in. D bars, and even the 3 x 3 x 8 in. D bars, have the same C and Si percentage limits for mottling. This appears to be true for the heats with no additions and with additions of Bi and Te. As in the previous work, additions were made to the melt

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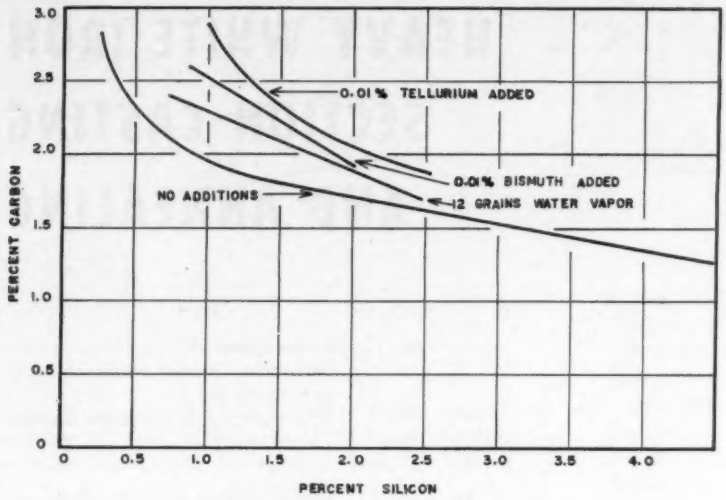



Fig. 1 — Curves showing the limiting carbon and silicon percentages for white iron fractures in a 4 x 4 x 8 in. casting risered on one end, from a reference.[1] The base curve is for irons with no additions melted under standard conditions, while the higher curves are for the addition of 12 grains water/cu ft furnace atmosphere, 0.01 per cent Bi and 0.01 per cent Te, respectively.

by plunging. The Bi or Te additions were wrapped in low carbon steel sheet. The package was wired to a rod for simultaneously plunging into the melt and storing.

A variety of additions were used alone and in combination. Aluminum, boron, zinc and 12 grains water vapor/cu ft* of furnace gases were used as well as Bi and Te. Percentage added varied from 0.003 per cent B, 0.016 to 0.02 per cent Al, 0.01 per cent Zn and 0.01 per cent Bi to 0.01 to 0.20 per cent Te. Table 1 gives the kind and size of additions for each heat. No different results were obtained than have already been described in past reports.[1,3]

## CASTING WHITE WITH CERIUM ADDITIONS

In a reference,[1] the authors hypothesized that certain other elements should have an effect similar to that of Bi and Te in raising the C and Si percentages for mottling. Cerium was suggested as one of these elements, but was found to be ineffective in the range of 0.01 to 0.11 per cent Ce added. This was thought to be due to a side reaction of Ce with S, resulting in the formation of cerium sulfide. If the sulfur content of the iron was lowered, it was thought that cerium would then be effective. To ascertain the truth, heats M59, M60, M63, M78, M79, M80 and M83 were prepared with and without desulfurization.

By means of basic slag, sulfur content was lowered from the range of 0.05 to 0.09 per cent S to a range of 0.015 to 0.025 per cent. Desulfurization was performed with a synthetic slag consisting of lime, calcium carbide, fluorspar, sodium carbonate, magnesia and silica with a basicity ratio of 2·3. The slag materials were placed on the cold metal charge, and remained on the metal until a temperature of 2750-2800 F was reached.

Cerium in amounts of 0.11 to 0.70 per cent was added to the iron by plunging it into the iron at 2750 F. Under these conditions, cerium was found to

*At a flow rate of 0.78 cfm.

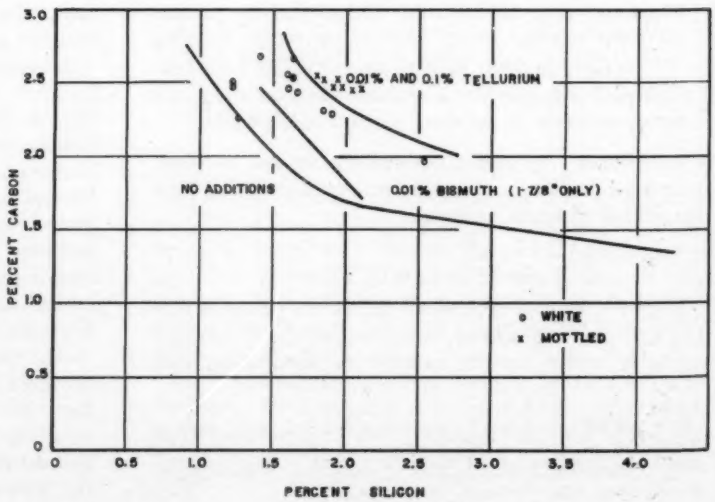



Fig. 2 — Curves showing the limited carbon and silicon percentages for white iron fractures in a 1⅞-in. D bar, a 2 x 2 x 8 in. bar and a 3 x 3 x 8 in. bar risered on both ends. The lower curve is for irons with no additions melted under standard conditions, while the higher curves are for additions of 0.01 per cent Bi and 0.01 per cent Te, respectively.

## TABLE 1 — BASE COMPOSITIONS AND ADDITIONS

| Heat No. | Base Composition, % C | Si | P, est | Addition, % | Casting Size — Fracture, in. 1⅜ | 1⅞ | 2 x 2 x 8 | 3 x 3 x 8 | 4 x 4 x 8 |
|---|---|---|---|---|---|---|---|---|---|
| N 19 | 2.56* | 1.17* | 0.102 | | | M | HM | G | |
| N 21 | 2.51* | 1.14* | 0.102 | | | | LM | LM | |
| N 54 | 2.06 | 1.01 | 0.090 | | | | W | W | |
| N 55 | 1.92* | 1.08* | 0.090 | | | | W | W | |
| N 58 | 2.06 | 1.15 | 0.090 | | | | W | W | |
| N 59 | 2.06 | 1.25 | 0.090 | | | | W | W | |
| M 1 | 2.18* | 1.04* | 0.088 | | | | W | W | |
| M 3 | 2.34 | 0.95 | 0.096 | | | | W | W | |
| M 4 | 2.32* | 1.12* | 0.096 | | | | W | W | |
| M 6 | 2.28* | 1.18* | 0.096 | | | | W | W | |
| M 8 | 2.36* | 1.28* | 0.096 | | | | LM | LM | |
| M 14 | 2.36 | 1.28 | 0.096 | | | | LM | LM | |
| N 22 | 2.44* | 1.15* | 0.102 | 0.01 Bi | | W | W | W | |
| N 23 | 2.48* | 1.24* | 0.077 | 0.01 Bi | | | W | W | |
| N 24 | 2.28 | 1.25 | 0.077 | 0.01 Bi | | | W | W | |
| N 25 | 2.28 | 1.25 | 0.077 | 0.01 Bi | | | W | W | |
| N 26 | 2.48* | 1.23* | 0.077 | 0.01 Bi | | W | W | W | |
| N 28 | 2.51* | 1.24* | 0.102 | 0.01 Te | | | W | W | |
| N 29 | 2.48* | 1.24* | 0.102 | 0.01 Te | | | W | W | |
| N 30 | 2.28 | 1.25 | 0.077 | 0.01 Te | | | W | W | |
| N 31 | 2.11 | 1.50 | 0.088 | 0.01 Te | | | | | W |
| N 34 | 2.11 | 1.50 | 0.088 | 0.01 Te | | | | | W |
| M 2 | 2.34* | 1.85* | 0.096 | 0.01 Te | | | W | W | |
| M 5 | 2.48* | 1.81* | 0.116 | 0.1 Te | | | W | W | |
| M 7 | 2.44* | 1.85* | 0.116 | 0.1 Te | | | W | W | |
| M 9 | 2.54* | 2.02* | 0.116 | 0.1 Te | | | LM | LM | |
| M 10 | 2.50 | 2.00 | 0.116 | 0.01 Te | | | HM | G | |
| M 11 | 2.50 | 2.00 | 0.116 | 0.2 Te | | | W | LM | |
| M 12 | 2.50 | 1.90 | 0.116 | 0.2 Te | | | W | LM | |
| M 13 | 2.50 | 2.00 | 0.116 | 0.01 Te | | | LM | HM | |
| M 16 | 2.51 | 1.60 | 0.090 | 0.01 Te | | | W | W | |
| M 17 | 2.51 | 1.60 | 0.090 | 0.1 Te | | | W | W | |
| M 50 | 2.68 | 1.40 | 0.096 | 0.01 Te | | | W | W | |
| M 51 | 2.52 | 1.80 | 0.090 | 0.01 Te | | | LM | M | |
| M 53 | 2.52 | 1.75 | 0.090 | 0.01 Te | | | M | M | |
| M 54 | 2.37 | 1.90 | 0.089 | 0.01 Te | | | M | M | |
| M 55 | 2.52 | 1.60 | 0.090 | 0.01 Te | | | W | LM | |
| M 56 | 2.41 | 1.60 | 0.091 | 0.01 Te | | | W | W | |
| M 57 | 2.26 | 1.90 | 0.089 | 0.01 Te | | | W | W | |
| N 36 | 2.36* | 1.20* | 0.125 | 0.02 Al; 0.01 Bi; 6 gr $H_2O/ft^3$ | | | W | W | |
| N 41 | 2.32* | 1.25* | 0.125 | 0.01 Te; 0.003 B; 8 gr $H_2O/ft^3$ | | | W | W | |
| N 42 | 2.40* | 1.23* | 0.125 | 0.01 Te; 0.003 B; 8 gr $H_2O/ft^3$ | | | W | W | |
| N 43 | 2.44* | 1.39* | 0.125* | 0.01 Te; 0.003 B; 8 gr $H_2O/ft^3$ | | | W | W | |
| N 44 | 2.52 | 1.45 | 0.113 | 0.01 Te; 0.003 B; 8 gr $H_2O/ft^3$ | | | G | G | |
| N 45 | 2.50* | 1.40* | 0.126* | 0.01 Te; 0.003 B; 8 gr $H_2O/ft^3$ | | | W | W | |
| N 46 | 2.50* | 1.39* | 0.113* | 0.01 Te; 0.003 B; 8 gr $H_2O/ft^3$ | | | W | W | |
| N 47 | 2.42* | 1.13* | 0.090* | 0.01 Te; 0.003 B; 8 gr $H_2O/ft^3$ | | | W | W | |
| N 48 | 2.46* | 1.35* | 0.090* | 0.01 Te; 0.003 B; 8 gr $H_2O/ft^3$ | | | W | W | |
| N 49 | 2.46* | 1.21* | 0.080* | 0.01 Te; 0.003 B; 8 gr $H_2O/ft^3$ | | | W | W | |
| N 50 | 2.42* | 1.39* | 0.073* | 0.01 Te; 0.003 B; 8 gr $H_2O/ft^3$ | | | W | W | |
| N 51 | 2.38* | 1.20* | 0.078* | 0.01 Te; 0.003 B; 8 gr $H_2O/ft^3$ | | | | | W |
| N 53 | 2.42* | 1.40* | 0.086* | 0.01 Te; 0.003 B; 8 gr $H_2O/ft^3$ | | | | | W |
| N 14 | 2.10 | 1.40 | 0.090 | 0.02 Al; 0.01 Zn | Run out | M | | | G |
| N 15 | 2.36* | 1.58* | 0.066 | 0.01 Al; 0.01 Bi | W | W | | | W |
| N 16 | 2.02 | 1.50 | 0.090 | 0.004 Te; 0.02 Al | W | W | | | LM |
| N 17 | 2.34* | 1.55* | 0.066 | 0.02 Al; 0.01 Bi | W | W | | | W |
| N 18 | 2.10 | 1.40 | 0.089 | 0.02 Al; 0.01 Zn | LM | LM | | | G |
| N 32 | 2.44* | 1.23* | 0.078 | 0.02 Al; 0.01 Te | | | W | W | |
| N 33 | 2.43* | 1.23* | 0.080 | 0.02 Al; 0.01 Bi | | | Thermal data | | |
| N 35 | 2.38* | 1.25* | 0.080 | 0.02 Al; 0.01 Te | | | W | W | |
| N 37 | 2.38* | 1.40* | 0.066 | 0.02 Al; 0.01 Te | | | W | W | |
| N 38 | 2.44* | 1.38* | 0.073 | 0.02 Al; 0.01 Bi | | | W | W | |
| N 39 | 2.44* | 1.29* | 0.119* | 0.02 Al; 0.01 Te; 0.003 B | | | W | W | |
| N 40 | 2.42* | 1.18* | 0.120* | 0.01 Te; 0.003 B | | | W | W | |
| N 60 | 2.19 | 1.60 | 0.089 | 0.01 Bi; 0.01 Al | | | | | W |
| N 61 | 2.19 | 1.60 | 0.089 | 0.01 Bi; 0.02 Al | | | | | W |
| N 62 | 2.19 | 1.60 | 0.089 | 0.01 Bi; 0.03 Al | | | | | W |
| M 61 | 2.54* | 1.52* | 0.090 | 0.01 Bi; 0.01 Te | | | LM | M | |
| M 62 | 2.26 | 1.90 | 0.088 | 0.01 Bi; 0.01 Te | | | M | HM | |
| M 63 | 2.36* | 1.50* | 0.093 | 0.25 Ce; Basic slag | | | W | W | |
| M 64 | 2.26 | 1.90 | 0.088 | 0.01 Te; 0.01 B; 0.01 Al | | | LM | M | |
| M 65 | 2.41 | 1.60 | 0.093 | 0.01 Te; 0.01 B; 0.01 Al | | | LM | LM | |

*(continued on next page)*

TABLE 1 — BASE COMPOSITIONS AND ADDITIONS — *continued*

| Heat No. | Base Composition, % C | Si | P, est | Addition, % | Casting Size — Fracture, in. 1⅜ | 1⅞ | 2 x 2 x 8 | 3 x 3 x 8 | 4 x 4 x 8 |
|---|---|---|---|---|---|---|---|---|---|
| M 59 | 2.52 | 2.25 | 0.096 | 0.13 Ce (Desulfurized 0.015S) | | | LM | LM | |
| M 60 | 2.52 | 2.25 | 0.096 | 0.26 Ce (Desulfurized 0.015S) | | | LM | LM | |
| M 63 | 2.36* | 1.50* | 0.093 | 0.25 Ce (Desulfurized 0.015S) | | | W | LM | |
| M 78 | 2.43 | 2.25 | 0.120 | 0.18 Ce | | G | | | |
| M 79 | 2.60* | 2.25 | 0.120 | 0.44 Ce | | W | | | |
| M 80 | 2.54* | 2.25 | 0.120 | 0.60 Ce | | W | | | |
| M 83 | 2.50 | 2.25 | 0.120 | 0.31 Ce | | G | | | |

* — Commercial chemical analysis.
W — White.
M — Mottled.
LM — Lightly mottled.
HM — Heavily mottled.
G — Gray.

produce a white fracture when the amount added exceeded 0.35 to 0.40 per cent in undesulfurized heats M79 and M80. The latter heats were made at a nominal analysis of 2.25 per cent Si and 2.50 per cent C. This carbon and silicon level is above the mottling curves for tellurium in Fig. 2.

A white fracture was produced when the added cerium exceeded about 0.25 to 0.30 per cent in desulfurized heats M60 and 63. Many additional heats not reported in Table 1 have been made using cerium additions. These heats all show that cerium in amounts of 0.35 to 0.40 per cent produces a white fracture in 1⅞-in. D bar, without prior desulfurization of the iron. This is accomplished over the composition range of 1.50 to 2.25 per cent Si and 2.50 per cent C. Cerium may then be regarded as an element having an effect on mottling similar to that of Bi and Te but requiring a larger addition.

TABLE 2 — FREEZING TIME OF WHITE IRON POURED IN GREEN SAND, 2700 F POURING TEMPERATURE

| Casting Size, in. | Freezing Time, min | Comments |
|---|---|---|
| 1 15/16 x 1 1/16 x 8 | 12.0 | Riser on each end of bar |
| 2 15/16 x 2 13/16 x 8 | 23.0 | Riser on each end of bar |
| 4 x 4 x 8 | 39.0 | Riser on each end of bar |

## FREEZING TIME OF TEST BAR CASTINGS

The freezing time of the castings in green sand molds was determined from cooling curves using platinum type thermocouples inserted into the mold cavity. The freezing time was chosen as the time from pouring the iron (at 2700 F) to the end of the eutectic freezing process. Freezing times are listed in Table 2. Freezing times could be plotted as a function of A/V ratio, or freezing time vs. section size. The latter type of plot, shown in Fig. 3 for these data, is the middle curve for bars. The present data were found to be in agreement with the data in the reference[4] on an A/V vs. freezing time basis. These data and reference[4] were then used to compute the freezing curves for plates and spheres, the upper and lower curves on Fig. 3.

Freezing time of the 3 x 3 x 8 in. bar is 23 min, while the 2 x 2 x 8 in. bar freezes is 12 min. One may ask then why the two bars show almost identical

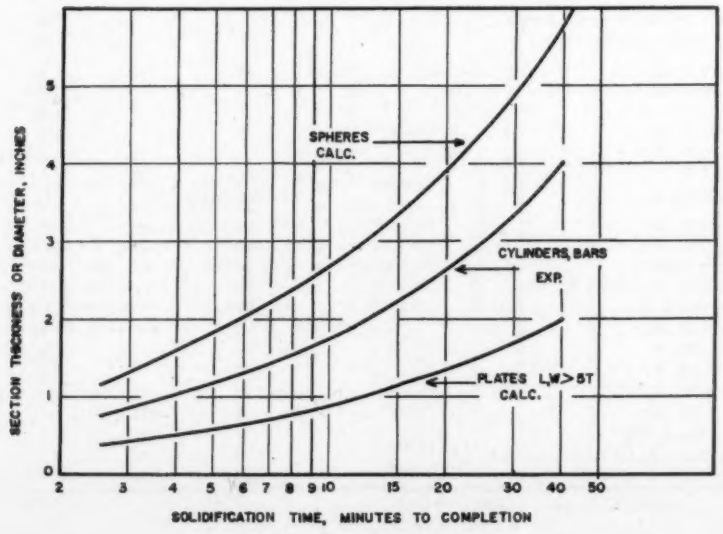



Fig. 3 — Relationship of freezing time of cast sections to section size of casting in a green sand mold. The metal is poured at 2700 F. Points on middle curve are from Table 2. Upper and lower curves are calculated.

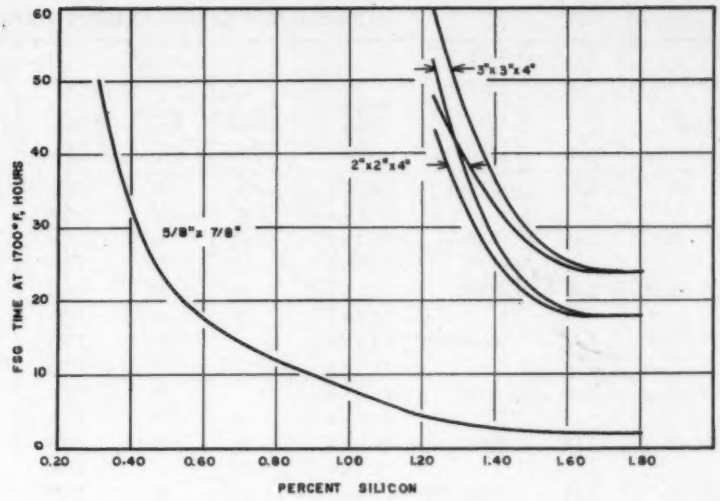



Fig. 4 — Time required for FSG at 1700 F as related to per cent silicon in the iron for 2 x 2 x 4 in. and 3 x 3 x 4 in. bars. The lower curve is the minimum time required for a 0.75 x 0.625 x 1.50 in. bar having over 3500 graphite nodules/cu mm (see text).

curves of C and Si limits for mottling in Fig. 2. In only one case heat M55, Table 1, was the 3 x 3 x 8 in. bar mottled when the smaller bar was not, with normal size Te additions. The 4 x 4 x 8 in. bar definitely has lower limits for carbon and silicon than the smaller bar. The answer to this involves the question of the temperature where mottling begins. Cooling curve and other data indicate that mottling begins at the upper temperatures in the eutectic range.

The carbon and silicon limits for mottling would be more closely related to the time to cool through the temperature range where mottling begins than it would to the time to cool through the end of the eutectic freezing temperature. The start of eutectic freezing is reached almost as soon in the 3 x 3 x 8 in. bar as in the 2 x 2 x 8 in. D bar, even though the end of eutectic freezing takes much longer in the bigger bar. This is believed to be the reason for the lack of difference in C and Si limits for mottling in the 2 x 2 x 8, 3 x 3 x 8 in. bars and 1⅞-in. D round bars.

## ANNEALING HEAVY SECTIONS

The time-temperature requirements for annealing light section castings of iron treated with bismuth, tellurium and other elements were described in a reference.[3] This reference reports the work done earlier in this project on the effect of a host of addition elements on annealability. This report extends the results to heavier sections represented by the 2 x 2 x 4 in. and 3 x 3 x 4 in. test bars. Annealability tests consist of determining the hours required for first stage graphitization (FSG) at 1700 F and the cooling rate in degrees F/hr necessary for second stage graphitization (SSG). The results of these tests are reported in Table 3. The FSG time at 1700 F is summarized graphically in Fig. 4. Figure 4 shows the time required to FSG the 2 x 2 x 4 in. and 3 x 3 x 4 in. test bars at 1700 F as a band for the range of less than or more than a certain number of hours at a specified per cent silicon. For comparison, the minimum FSG time required in a ⅝ x ⅞-in. sections is shown on Fig. 4 from a reference.[3] Some of the reasons for the differences in FSG time required in different section sizes will be discussed.

## NODULE NUMBER*

After studying graphitization during annealing in several heats, it soon became apparent that developing a sufficient number of graphite nodules was a major problem in heat treatment of heavy section malleable iron. For example in heats N29, N23, N40, N36 and N38, varying from 1.20 to 1.40 per cent Si, and varying in additions with 0.01 per cent Te, 0.01 per cent Bi, 0.02 per cent Al, 8 gr water vapor/cu ft and 0.003 B per cent, nodule numbers in the range of 650 to 2900 nodules/cu mm developed. This is too low for most rapid graphitization according to previous work.[3] Most rapid graphitization at SSG requires over about 3500 nodules/cu mm.

Inspection of Table 3 shows in all cases that the number was marginal or too low for FSG, and definitely too low for SSG at most rapid rates. Much effort was expended to raise nodule number. Many of the known methods of increasing nucleation were tried as:

1. *Boron.* Heats N39 to N53 had 0.003 per cent added.
2. *Aluminum.* Heats N32 to N39 had 0.02 per cent Al added.
3. *Water vapor or hydrogen.* Heats N35, N36 and N41 to N53 had 6 to 8 gr water vapor/cu ft furnace atmosphere gas.
4. *Pretreatment.* All heats were annealed with and without pretreatments of 2 hr at 600 F and 4 hr at 1200 F.
5. *Hot shakeout.* The 2 x 2 x 8 in. and 3 x 3 x 8 in. castings were shaken out 35 min after pouring, and the 4 x 4 x 8 in. castings were shaken out 45 min

*Nodule numbers are determined by the method described in reference 5.

TABLE 3 — ANNEALABILITY EXPERIMENTS

| Heat No. | Si, % | P, % | Section Size, in. | FSG Treatment Time, hr | | Temp., F | SSG Cooling Rate, F/hr | | Nodule Number | | Comments* |
|---|---|---|---|---|---|---|---|---|---|---|---|
| | | | | Greater Than | Less Than | | Greater Than | Less Than | For Shorter FSG Time | For Longer FSG Time | |
| N 41 | 1.25 | 0.125 | 2 x 2 x 4 | 30 | 36 | 1700 | | | 1960 | 2000 | 1 |
| | | | 3 x 3 x 4 | 40 | 46 | 1700 | | | 1660 | 1680 | 1 |
| N 42 | 1.23 | 0.125 | 2 x 2 x 4 | 42 | 48 | 1700 | | | 2220 | 1980 | 1 |
| | | | 3 x 3 x 4 | 52 | 60 | 1700 | | | 1720 | 1540 | 1 |
| N 43 | 1.39 | 0.125 | 2 x 2 x 4 | 30 | 36 | 1700 | | | 3220 | 3120 | 1 |
| | | | 3 x 3 x 4 | 40 | 48 | 1700 | | | 2260 | 2160 | 1 |
| N 45 | 1.40 | 0.126 | 2 x 2 x 4 | | 24 | 1700 | | | | 3020 | 1 |
| | | | 3 x 3 x 4 | | 34 | 1700 | | | | 2560 | 1 |
| N 46 | 1.39 | 0.113 | 2 x 2 x 4 | | 56 | 1700 | 10 | 20 | 2900 | 2700 | 1 |
| | | | 3 x 3 x 4 | | 56 | 1700 | 10 | 20 | 2500 | 2300 | 1 |
| N 47 | 1.13 | 0.090 | 2 x 2 x 4 | 30 | 36 | 1700 | | | 1100 | 1040 | 1 |
| | | | 3 x 3 x 4 | 40 | 46 | 1700 | | | 760 | 780 | 1 |
| N 48 | 1.35 | 0.090 | 2 x 2 x 4 | 24 | 30 | 1700 | | | 2600 | 2440 | 1 |
| | | | 3 x 3 x 4 | 34 | 40 | 1700 | | | 1780 | 1860 | 1 |
| N 49 | 1.21 | 0.080 | 2 x 2 x 4 | | 56 | 1700 | 10 | 20 | 1500 | 1600 | 1 |
| | | | 3 x 3 x 4 | | 56 | 1700 | 10 | | | 1480 | 1 |
| N 50 | 1.39 | 0.073 | 2 x 2 x 4 | | 56 | 1700 | 10 | 20 | | 3000 | 1 |
| | | | 3 x 3 x 4 | | 56 | 1700 | 10 | 20 | | 2200 | 1 |
| N 51 | 1.20 | 0.078 | 4 x 4 x 4 | 50 | 56 | 1700 | | | 800 | 900 | 1 |
| N 53 | 1.40 | 0.086 | 4 x 4 x 4 | 44 | 50 | 1700 | | | 1500 | 1680 | 1 |
| M 2 | 1.85 | 0.096 | 2 x 2 x 4 | 12 | 24 | 1700 | | | 662 | 1312 | 2 |
| | | | 3 x 3 x 4 | 12 | 24 | 1700 | | | 1750 | 2250 | 2 |
| M 7 | 1.85 | 0.116 | 2 x 2 x 4 | | 12 | 1700 | | | | 5000 | 2 |
| | | | 3 x 3 x 4 | 12 | 24 | 1700 | | | 2662 | 2920 | 2 |
| M 12 | 1.90 | 0.116 | 2 x 2 x 4 | 12 | 24 | 1700 | | | 1420 | 1780 | 2 |
| | | | 3 x 3 x 4 | 12 | 24 | 1700 | | | 787 | 1160 | 2 |
| M 16 | 1.60 | 0.090 | 2 x 2 x 4 | 12 | 24 | 1700 | | | 1370 | 1912 | 2 |
| | | | 3 x 3 x 4 | 12 | 24 | 1700 | | | 862 | 1280 | 2 |
| M 17 | 1.60 | 0.090 | 2 x 2 x 4 | 12 | 24 | 1700 | | | 1330 | 2650 | 2 |
| | | | 3 x 3 x 4 | 12 | 24 | 1700 | | | 630 | 1130 | 2 |
| M 50 | 1.40 | 0.096 | 2 x 2 x 4 | | 24 | 1760 | | | | 1300 | 3 |
| | | | 3 x 3 x 4 | | 24 | 1760 | | | | 500 | 3 |
| M 55 | 1.60 | 0.090 | 2 x 2 x 4 | | 24 | 1760 | | | | 1425 | 3 |
| | | | 3 x 3 x 4 | | 24 | 1760 | | | | 775 | 3 |
| M 56 | 1.60 | 0.091 | 2 x 2 x 4 | | 24 | 1760 | | | | 1540 | 3 |
| | | | 3 x 3 x 4 | | 24 | 1760 | | | | 1310 | 3 |
| M 57 | 1.90 | 0.089 | 2 x 2 x 4 | | 24 | 1760 | | | | 1630 | 3 |
| | | | 3 x 3 x 4 | | 24 | 1760 | | | | 1330 | 3 |
| M 50 | 1.40 | 0.096 | 2 x 2 x 4 | | 24 | 1815 | | | | 530 | 3 |
| | | | 3 x 3 x 4 | | 24 | 1815 | | | | 460 | 3 |
| M 55 | 1.60 | 0.090 | 2 x 2 x 4 | | 24 | 1815 | | | | 3030 | 3 |
| | | | 3 x 3 x 4 | | 24 | 1815 | | | | 630 | 3 |
| M 56 | 1.60 | 0.091 | 2 x 2 x 4 | | 24 | 1815 | | | | 1150 | 3 |
| | | | 3 x 3 x 4 | | 24 | 1815 | | | | 1360 | 3 |
| M 57 | 1.90 | 0.089 | 2 x 2 x 4 | | 24 | 1815 | | | | 1980 | 3 |
| | | | 3 x 3 x 4 | | 24 | 1815 | | | | 1030 | 3 |
| M 59 | 2.25 | 0.096 | 2 x 2 x 4 | | 24 | 1700 | 10 | 20 | | 990 | 2 |
| | | | 3 x 3 x 4 | | 24 | 1700 | 10 | 20 | | 610 | 2 |
| M 60 | 2.25 | 0.096 | 2 x 2 x 4 | | 24 | 1700 | 20 | | | 1350 | 2 |
| | | | 3 x 3 x 4 | | 24 | 1700 | 10 | 20 | | 870 | 2 |
| M 61 | 1.52 | 0.090 | 2 x 2 x 4 | 18 | | 1700 | 20 | | 1400 | | 4 |
| | | | 3 x 3 x 4 | 18 | | 1700 | | 20 | 1460 | | 4 |
| M 62 | 1.90 | 0.088 | 2 x 2 x 4 | 18 | | 1700 | 20 | | 1500 | | 4 |
| M 63 | 1.50 | 0.093 | 2 x 2 x 4 | 18 | | 1700 | | 20 | 1770 | | 4 |
| | | | 3 x 3 x 4 | 18 | | 1700 | | 20 | 1460 | | 4 |
| M 64 | 1.90 | 0.088 | 2 x 2 x 4 | 18 | | 1700 | 20 | | 1410 | | 4 |
| | | | 3 x 3 x 4 | 18 | | 1700 | 20 | | 1630 | | 4 |
| M 65 | 1.60 | 0.093 | 2 x 2 x 4 | 18 | | 1700 | | 20 | 1130 | | 4 |
| | | | 3 x 3 x 4 | 18 | | 1700 | | 20 | 1130 | | 4 |

* 1. Hot shakeout and standard addition of 0.01 Te; 0.003 B; 8 gr $H_2O/ft^3$ — Pretreatment 2 hr 600 F — 4 hr 1200 F.
2. Pretreatment 2 hr 600 F — 4 hr 1200 F.
3. No Pretreatment.
4. Pretreated 4 hr at 600 F; 4 hr 1200 F.

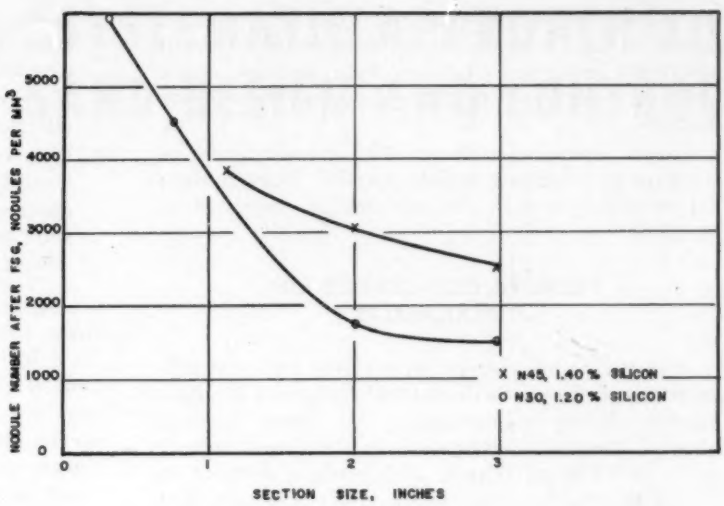



Fig. 5 — Effect of section size on number nodules after FSG in heats N45 and N30. The 2 in. and 3 in. sections represent 2 x 2 x 4 in. and 3 x 3 x 4 in. bar castings, while the 0.25 to 1.125 sections are thin slices cut from the bar castings.

after pouring. They were cooled under an air blast on the mechanical shake out.

6. All of the above combined.

All of these efforts at increasing nodule number failed. They did produce an increase compared with irons not receiving the treatments, but the highest nodule number did not rise above 3000/cu mm, and in most cases remained below 2500/cu mm. From this it appears that the nucleating factors reported in the literature operate most effectively in thin section castings. Certainly this was the experience previously.[3] Thus, it appears there is a mass effect acting against nucleation during annealing. To study this, thin sections were cut from thick as-cast sections and then annealed.

The effect on nodule number in heats N30 and N45 is shown in Table 4 and in Fig. 5. This confirms the mass effect against nucleation. Possible improvement in nodule number and annealability in heavy sections might come from increasing per cent silicon in the iron. This was studied, and the trend is illustrated in Fig. 6. Increasing silicon content results in higher nodule number, as shown in Fig. 6, within the composition range studied. It should be acknowledged that the nodule number, shown on Fig. 6, is a maximum obtained by treatments 1-5 above.

Figure 6 still shows a significantly lower number of nodules in the 3 x 3 x 4 in. casting compared with the 2 x 2 x 4 in. casting. However, even the maximum number of nodules, shown in Fig. 6, is too low for

TABLE 4 — NODULE NUMBER IN SECTIONS CUT FROM HEAVY SECTIONS

| Heat No. | Analysis, % C | Si | As-Cast Section, (in.) | Cut Section, (in.) | Addition, % | Nodule Number |
|---|---|---|---|---|---|---|
| N 30 | 2.28 | 1.25 | 2x2x8 | ¼ | 0.01 Te | 5000+ |
| N 30 | 2.28 | 1.25 | 2x2x8 | ¾ | 0.01 Te | 4500 |
| N 30 | 2.28 | 1.25 | 2x2x8 | 2x2x4 | 0.01 Te | 1700 |
| N 30 | 2.28 | 1.25 | 3x3x8 | 3x3x4 | 0.01 Te | 1500 |
| N 45 | 2.50 | 1.40 | 3x3x8 | 1⅛ | 0.01 Te; 0.003 B; 8 gr $H_2O$/ft³ | 3800 |
| N 45 | 2.50 | 1.40 | 3x3x4 | 3x3x4 | 0.01 Te; 0.003 B; 8 gr $H_2O$/ft³ | 2560 |
| N 45 | 2,50 | 1.40 | 2x2x8 | 2x2x4 | 0.01 Te; 0.003 B; 8 gr $H_2O$/ft³ | 3020 |

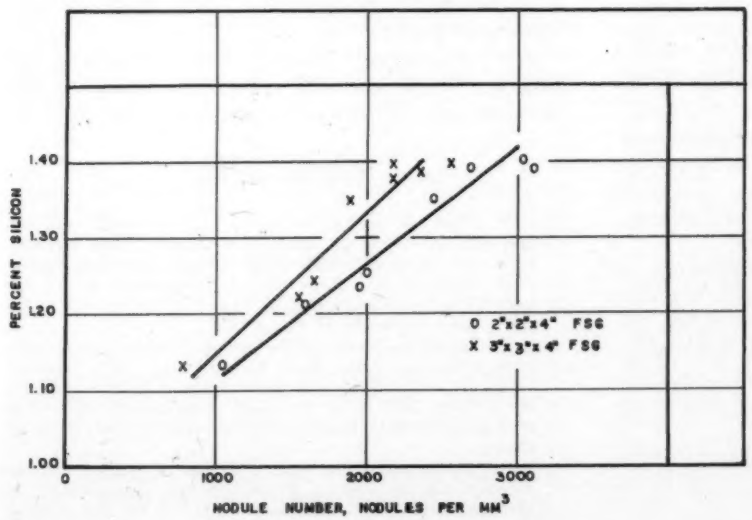



Fig. 6 — Effect of silicon content on nodule number in 2 x 2 x 4 in. bar and 3 x 3 x 4 in. bar castings. This nodule number is developed only with full use of nucleation, tested as 1-6 in the text.

maximum rate of graphitization. Thus the FSG time curves in Fig. 4 for the heavy sections will remain above the minimum curve for light sections, unless some means is found to increase nucleation in heavy sections.

Higher temperature during FSG was considered as a means of increasing nodule number. Temperatures up to 1800 F were of no assistance, as indicated in Table 3.

## PHOSPHORUS EFFECT ON ANNEALABILITY

A second major problem of studying FSG and SSG in heavy sections was caused by segregation of phosphorus during solidification. All heavy sections showed phosphide segregation in the range studied, 0.073 to 0.126 per cent P. This made it difficult to detect the end of FSG. Much effort was expended with etching reagents to determine the time required for less than one per cent carbides to remain after FSG, as distinct from residual phosphides that cannot, of course, be graphitized.

This uncertainty of end point is the reason for the rather wide bracketing of FSG time, as shown in Table 3 and in Fig. 4. Low phosphorus content iron would eliminate this problem. Generally the phosphides are fully dissolved after second stage annealing. This may be due to the high solubility of phosphorus in ferrite.

## SECOND STAGE ANNEALING

Based on nodule counts after FSG, it was predicted from the previous report[3] that SSG could be accomplished at a rate cooling slower than 20 F/hr and faster than 10 F/hr. This turned out to be true on the heats tested. Heats N46, N49, N50, M59 and M60 illustrate this point. However, in cases of exceptionally low nodule number a cooling rate of less than 10 F/hr may need to be used, although no case of this type was encountered in this work.

## CONCLUSIONS

In reviewing the results of annealability experiments with heavy section white iron castings certain facts emerge:

1. A drastic decrease in the number of graphite nodules developed during annealing occurs as the as-cast section size increased. The decrease in nodule number substantially increases the time requirements for FSG and SSG.
2. When Bi, Te or Ce are used for prevention of mottling there is no improvement in the number of nodules developed in heavy sections during annealing. All three are accompanied by low nodule numbers in spite of their different chemical nature. This lends credence to the idea that increasing mass is an anti-nucleating factor, per se.
3. The factors known to cause large increases in nodule number in lighter section castings seem to be much less effective in heavy sections.
4. When thin sections are cut from heavy as-cast sections, they show an increase in nodule number compared to the heavy sections after annealing.
5. Increasing FSG annealing temperature showed no effect of improving nodule number.
6. Only increasing the silicon percentage produces a significant increase in nodule number. This point applies to iron treated with the known methods of increasing nucleation in light sections and to the silicon range studied, 1.13 to 1.40 per cent.

Considering these points, it appears that the FSG times, shown in Fig. 4, are realistic for 2 x 2 in. and 3 x 3 in. bar sections. Further, the minimum FSG time at 1700 F, shown in Fig. 4, for light sections will not be reached in heavy sections until some means are found to increase the nodule number developed during annealing. Second stage graphitization rates hinge on the same problems of nodule number. The maximum rate is about 20 F/hr with the most favorable nodule numbers obtained in this work. Slower rates of 10 F/hr or less are required at the lowest nodule numbers.

Cerium has been revealed as another element capable of producing white fractures at C and Si percentages where mottle or gray fractures would normally occur. The effectiveness of cerium in raising C and Si limits for mottling is similar to that of tellurium, presented in Fig. 2, but the percentage addition required is larger. No chemical analyses are available from this work to determine recovery of cerium.

## ACKNOWLEDGMENT

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# SYSTEMATIC APPROACH TO SAND DESIGN AND CONTROL

## *Progress report 5 — interpretation of data*

by A. H. Zrimsek and G. J. Vingas


### ABSTRACT

**This is an analysis of the interaction of the various physical properties of sand mixtures, considering all data presented in previous progress reports. This interaction of physical properties shows that southern and western bentonite react differently with mulling efficiency and time and wood flour content of the sand mixture. The wood flour addition to southern bentonite bonded sands reduces the density differential substantially, as though less clay is present in the sand than is actually present. Western bentonite, on the other hand, remains unaffected by the wood flour addition.**


### INTRODUCTION

Considerable data have already been presented in the first four reports of this series, with an even greater amount to be given in the six to eight progress reports that will follow. To those readers possessing a thorough understanding of graphical representation, this is an ideal method for presentation of data gathered for this series of reports. In the previous progress reports all of the principles involved were not brought forth clearly because of the limited nature of the individual report.

The writers feel that at this stage of the series an analysis of the interaction of the various physical properties, considering all the data, is necessary to clarify and accentuate some of the principles brought to light.

### RESULTS

The data presented in this report were selected from the previous four of this series. Data are included which emphasize some of the basic principles and characteristics of sand formulation.

### DISCUSSION

Although the graphical representation of data in previous reports of the series is straightforward, it fails in some instances to dramatically demonstrate the principles involved. Except for the discussion of green compression to green shear ratios, little mention has been made about the interaction of the various physical properties recorded. The study of relationships of one property to another can often clarify a principle involved.

In the earlier progress reports, the interaction of the various physical properties with water content and ramming energy was presented. An example of this can be seen in Figs. 1a and 1b. These figures show that as water content is increased both green compression and shear first increase in magnitude to a peak, then decrease in magnitude. They also show that rammed density first decreases to a minimum and then rises. As seen in these figures, dry compression and shear rise in magnitude with increasing water.

The figures also show that all properties increase in magnitude with increased ramming energy. This is typical of the data presented. It is apparent that for the system depicted in Figs. 1a and 1b, three ram green compression strengths between 5.0 psi and 16.8 psi can be obtained. This strength range just about covers the range encountered in various foundry operations. Judging by green compression alone, this system could pass the control specification of almost every foundry in the country, but, of course, the sands in this system would not perform satisfactorily in most foundries. Green compression alone does not describe sand quality. No one single physical property does for that matter.

#### *Physical Properties Interaction*

The importance of considering the interaction of the various physical properties as a means of sand control and design cannot be overemphasized. Consider the relationship between green compression and dry compression. It is apparent from Figs. 1a and 1b that as water is increased, green compression strength drops and dry compression rises. However, the exact relationship between the two properties is not readily seen. By replotting the 3 ram data of Figs. 1a and 1b, in a form as shown in Fig. 2, the relationship becomes apparent.

By also plotting the 3 ram green and dry relation-

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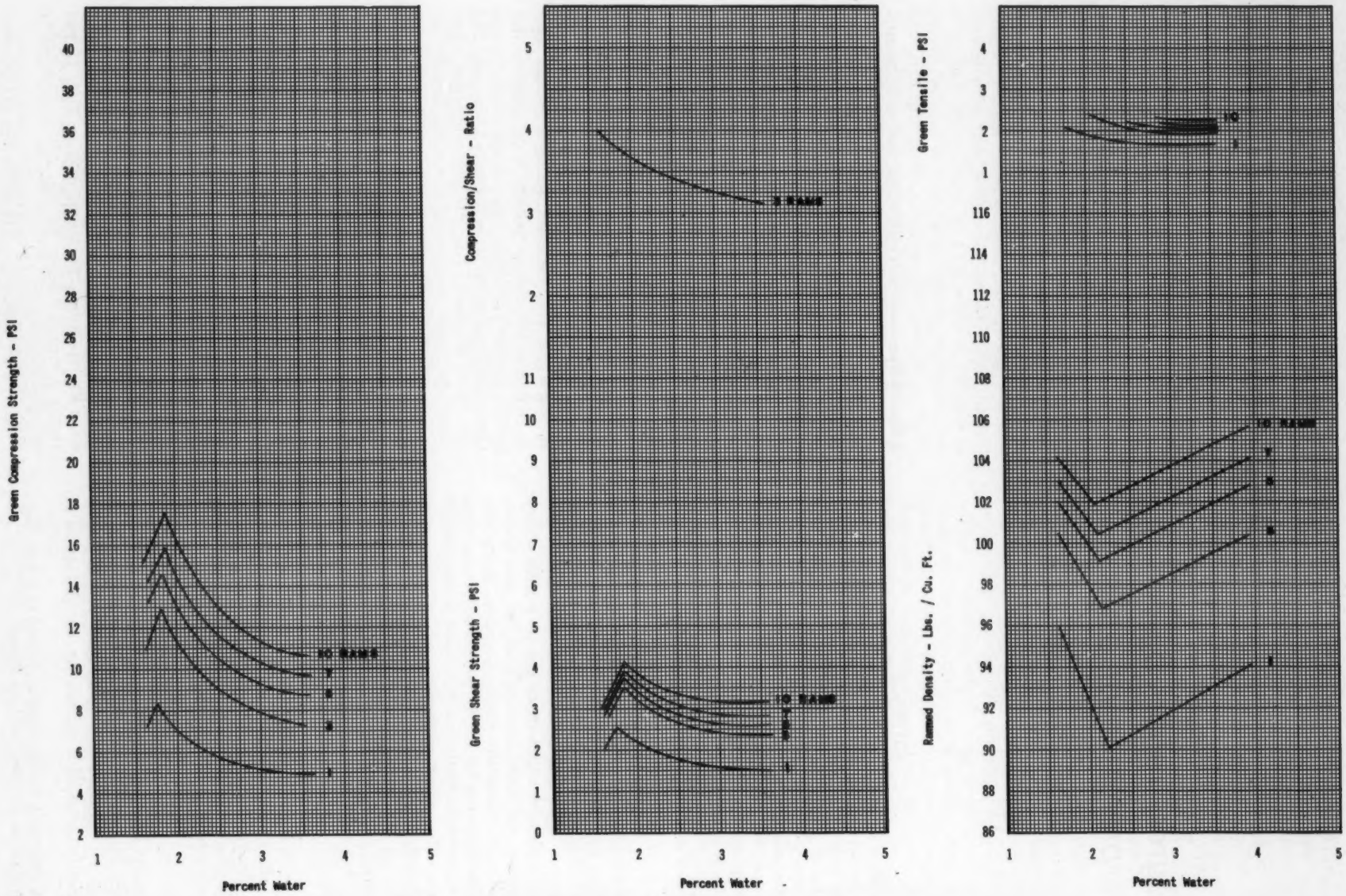



Fig. 1a — Water content and ramming energy effect on physical properties of 4.75 per cent western bentonite bonded Portage silica sand. Mulled 6 min.

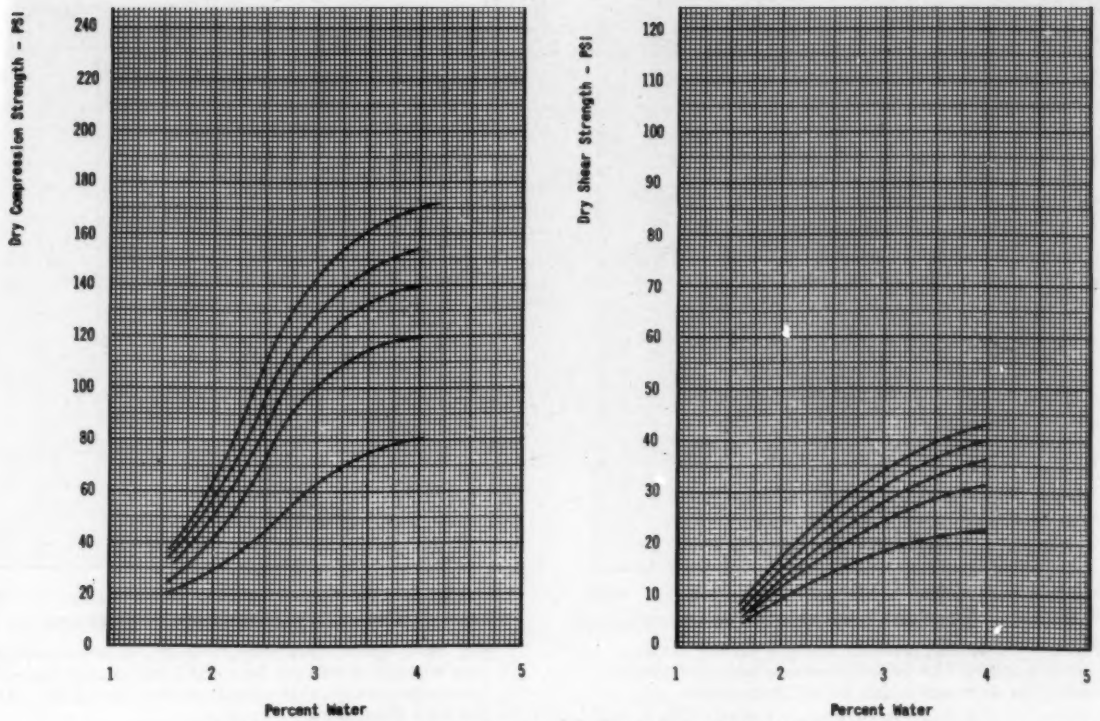



Fig. 1b — Water content and ramming energy effect on physical properties of 4.75 per cent western bentonite bonded Portage silica sand. Mulled 6 min.

ship for 7.45 and 10 per cent western bentonite bonded sands at 6 min mulling, it is possible through interpolation to pick the clay content required to provide a given combination of green compression and dry compression. To obtain a green compression of 14 psi and dry compression of 108 psi, a bentonite content of 7.45 per cent is needed. If the same 14 psi green is desired but only 80 psi dry, approximately 6.2 per cent western bentonite would be required. Almost any combination of green and dry compression strengths can be had by simply changing clay and water contents while holding mulling intensity constant.

This same type of presentation also gives a striking clarification of the importance of mulling control in any sand control program. Figure 3 gives a good picture of the effect of mulling time on changing the relationship of green to dry. Recalling that high dry strength occurs at high water contents, it can be seen that the relationship of green to dry is least affected at high water. It also emphasizes that the dry sands normally encountered in iron foundries are highly affected by changes in mulling efficiency.

The importance of mulling can be further indicated by comparing results obtained from mixtures bonded with 7.45 per cent western bentonite mulled in different sized mullers or with different methods of additions. Compare the relationships of Fig. 4 with those of Figs. 2 and 3, and it becomes evident that addition of water to sand followed by bentonite is a more efficient method of addition as compared with adding water to dry mixture of sand and bentonite.

It can also be seen that the efficiency of the small 12 in. diameter muller is substantially higher than the 18 in. diameter muller. The smaller muller imparts properties to 7.45 per cent western bentonite bonded sands which are comparable to the strengths of 10 per cent western bentonite bonded sands mulled in an 18 in. diameter muller.

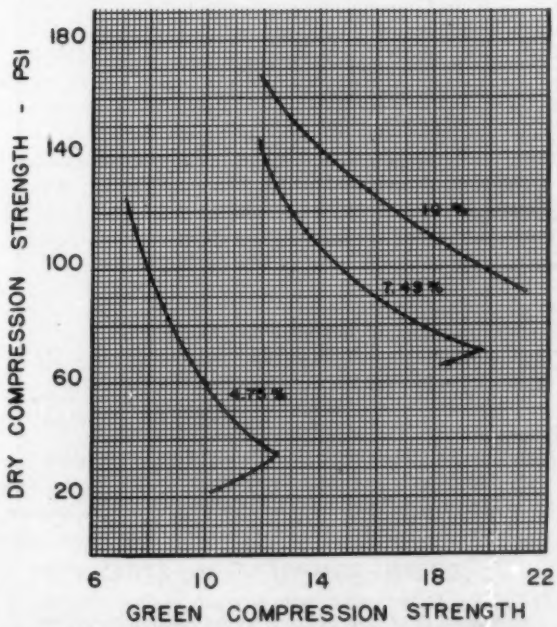



Fig. 2 — Green compression strength vs. green shear strength of 4.75, 7.45 and 10 per cent western bentonite systems mulled 6 min in an 18 in. laboratory muller.

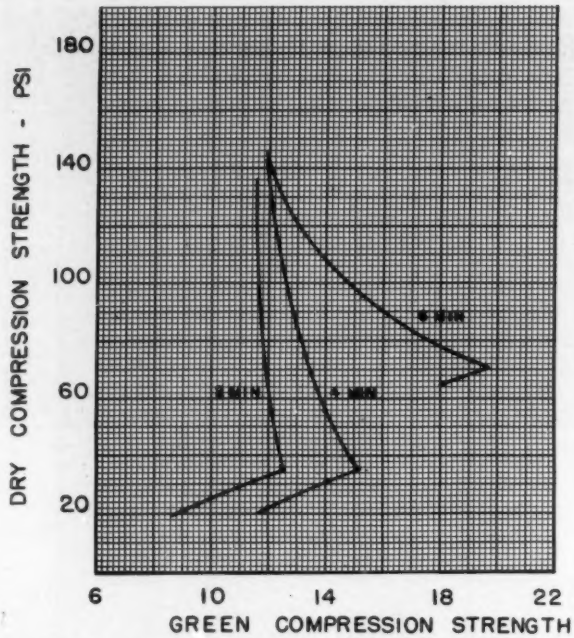



Fig. 3 — Green compression strength vs. dry compression strength of 7.45 per cent western bentonite system mulled for 2, 4 and 6 min in an 18 in. muller.

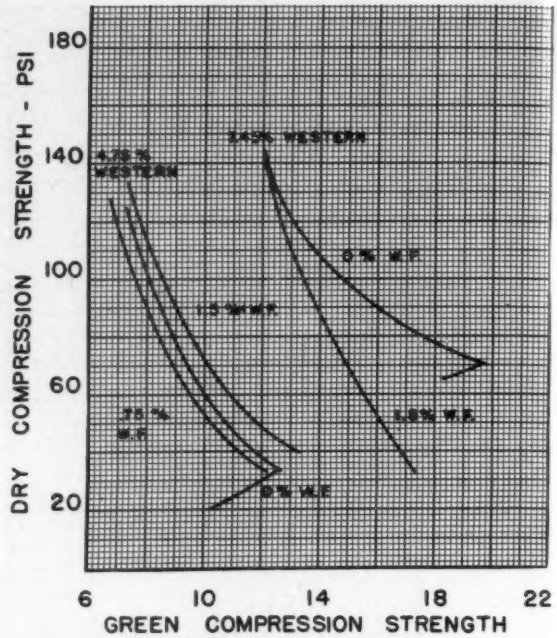



Fig. 5a — Green compression strength vs. dry compression strength of 4.75 per cent and 7.45 per cent western bentonite systems with additions of 0.75, 1.5 and 3.0 per cent wood flour.

*Wood Flour Data*

When data collected from mixtures containing wood flour are replotted in the form of green versus dry strength, as in Fig. 5, the difference in the effect of wood flour additions on western and southern bentonite can be seen. The relationship of green to dry for 4.75 per cent western is hardly affected by wood flour additions up to 3 per cent. Southern bentonite, on the other hand, is drastically affected by 0.75 per cent wood flour additions, with larger additions up to 3 per cent having little additional effect. The ef-

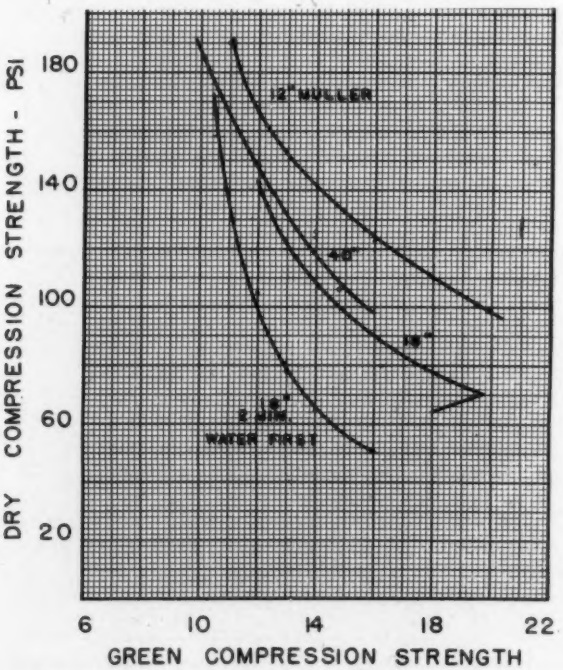



Fig. 4 — Green compression strength vs. dry compression strength of 7.45 per cent western bentonite system mulled 6 min in a 12, 18 and 40 in. muller. Data presented are also for the same system mulled 2 min in an 18 in. mulled with water added first.

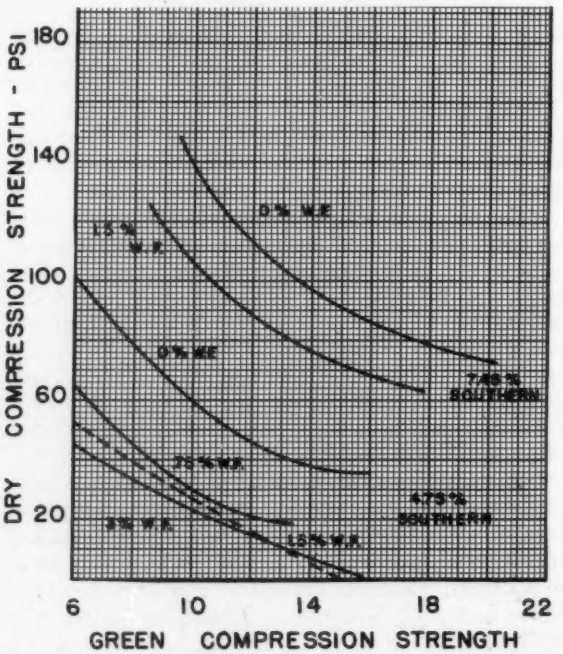



Fig. 5b — Green compression strength vs. dry compression strength of 4.75 per cent and 7.45 per cent southern bentonite systems with additions of 0.75, 1.5 and 3.0 per cent wood flour.

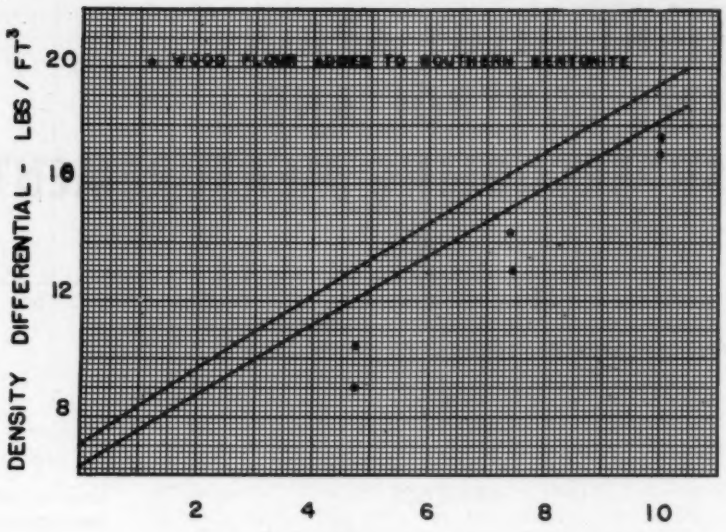



Fig. 6 — One to 10 rams density differential in lb/cu ft vs. clay content of sands bonded with southern and western bentonite and some mixtures in which wood flour was added.

fect of wood flour on higher western bentonite contents is one of retarding mulling. Comparing Fig. 5 with Fig. 3, it can be seen that a 7.45 per cent western-1.5 per cent wood flour combination mulled 6 min has green and dry characteristics similar to a simple 7.45 per cent western sand mulled only 4½ min. This agrees with observations made in progress report 4.

The relationship of green compression strength to green shear has been discussed at some length in previous reports of the series and will therefore not be covered here.

*Ramming Effect*

In the initial report of this series it was noted that the density differential between 1 and 10 rams remained constant at any given clay content regardless of mulling time or water content beyond that required for minimum density. Subsequent data collected from mixtures containing from 0.75 to 3 per cent wood flour show the 1-10 ram density differential at any given western bentonite content was not altered.

The 1-10 ram density difference for southern bentonite bonded sands was lowered significantly. If the data collected on density differential are summarized, a straight-line relationship between clay content and density differential becomes apparent. This relationship is given in Fig. 5. The great majority of data gathered fall within the band depicted. The addition of wood flour to western bentonite bonded sands did not have the effect of changing this relationship (Fig. 6).

Wood flour additions to southern bentonite bonded sands, however, had the effect of changing the ramming characteristics to resemble lower clay content then the actual one. Some of the density data for wood flour-southern bentonite mixtures are plotted in Fig. 5. If 1-10 ram density differential is an indication of clay content, the data imply that wood flour added to southern bentonite sands lowers the effective clay level.

The validity of this implication seems to be strengthened by the green versus dry presentation made earlier, and also by observation about the change in "feel" of the sands brought about by wood flour additions. Such additions to southern bentonite sands tended to make the sands feel lower in clay content than they actually were.

The interaction between green compression and green shear, green compression and dry compression, and 1-10 ram density differential at this point appear to hold the most significance.

## CONCLUSIONS

1. No one particular physical property of sand can adequately describe sand quality.
2. A study of the relationship of green compression strength and dry compression strength accentuates the importance of mulling, clay type and content and the influence wood flour has on clays.
3. The density differential from one to 10 rams is related to clay content in simple sand-clay-water formulations when moisture of the sand mixture is beyond that required for minimum density.
4. The interaction of the physical properties clearly shows that southern and western bentonite react differently with mulling efficiency, mulling time, and wood flour additions.
5. Wood flour addition to southern bentonite bonded sands reduces the density differential substantially as though less clay was present than actually existed in the mixture. Western bentonite, however, remains unaffected.

The above conclusions are based solely on the data and observations collected thus far. Subsequent reports will contain data which will further substantiate the above conclusions or repudiate them.

## ACKNOWLEDGMENT

The writers are indebted to Magnet Cove Barium Corp. for permission to conduct the necessary research work to investigate the principles developed in this series and publish the results of these investigations.

# LIGHT ALLOY GRAIN SIZE CONTROL AND SUPERCOOLING MEASUREMENTS

by V. B. Kurfman


## ABSTRACT

**A simple cooling curve determination is described which has shown supercooling effects up to 15 F in commercial purity magnesium and aluminum alloys. These high supercooling values are related to coarse grain size.**

**The effects are large enough to allow use of the method for quality control of grain refinement processes. In some cases, off composition melts can be analyzed by inspection of the resulting curve.**

**A complete measurement can be completed in the melt room within 1-2 min.**

**Data are presented for Mg-Al, Mg-Zn, Al-Si, Al-Cu and Al-Mg.**


## INTRODUCTION

Supercooling or undercooling is a well known phenomenon in pure metals and certain alloys. The best known modern studies of the subject have dealt with homogeneous nucleation in small isolated droplets of clean metals. Under these conditions supercooling of about 18 per cent of the absolute melting point has been observed.[1] Nucleation then occurs, the droplets solidify liberating the latent heat of freezing and temperature rises to approach the liquidus temperature, i.e., recalescence occurs.

Considerable interesting work has also been done on supercooling of larger melts in which large effects are observed. In Walker's work on Ni,[2] as in that on Fe by Barbenheur and Bleckmann,[3] the necessity for keeping the melt undisturbed and free of chance nucleants is stressed. Cibula reported some supercooling of coarse grained Al alloys cast in a sand mold.[4] He noted that the addition of various grain refining agents eliminated this effect.

The theory of homogeneous nucleation and supercooling has been well described by Turnbull.[1] Heterogeneous nucleation is of more importance in the freezing of castings. However, the theory is not so well defined. The work of Chalmers and his associates[5] treats the interaction of such nuclei and supercooling. The "constitutional supercooling" concepts have been given a rather detailed theoretical exposition in treating the interaction of thermal gradient, heterogeneous nuclei and diffusion barriers to freezing.

The interaction of supercooling with nucleation or grain refinement processes in commercial practice remains rather vague. The magnitude of the effects which may occur, their relation to the final grain diameter of the casting and the supercooling necessary or sufficient to allow the use of a given grain refinement agent is generally unspecified.

During the present study, it has been found possible to observe supercooling in commercial casting alloys of as much as 15 F (8 C), using simple techniques and few special precautions. It is possible to use this sort of measurement to estimate the grain diameter of various casting alloys, or at least to determine if a melt will be coarse grained.

## EXPERIMENTAL PROCEDURE

The melts described are magnesium- or aluminum-base alloys prepared according to commercial practice from standard commercial purity materials. The treatment of the melts was according to production procedure except for cases in which deliberate grain coarsening was attempted.

The cooling curve samples were taken in thin-wall armco iron crucibles of 20 $cm^3$ capacity, i.e., 30-35 gm of liquid Mg. A standard glass-coated iron-constantan thermocouple with a stripped and twisted bare metal junction was used with a strip chart recorder operated at 12 in./min chart speed. Relative temperatures were estimated to $\pm$ 1 F with this equipment. For absolute liquidus temperature determinations, the recorder was standardized against an indicating potentiometer, and in some cases calibrated against the melting point of sublimed Mg.

A selected coil of thermocouple wire was reserved for the measurements, and for any given series of cooling curves a sufficient length of wire was installed to allow its continued use by clipping off the inch or so of wire lost in the sample. In no case did the recorded melting point for sublimed Mg deviate more than $\pm$ 2 F from the literature value of 1202 F (650 C).

For the magnesium alloys the sample was dipped from the melt, dusted with an agent to prevent burning and allowed to air cool in a sheltered position.

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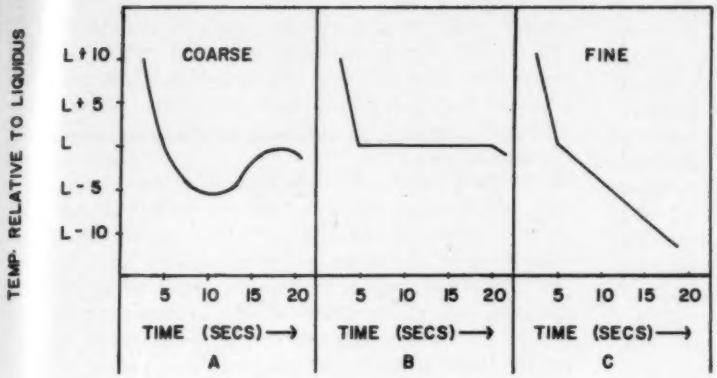



Fig. 1 — Types of liquidus arrest observed on rapid cooling.

The cooling rate prior to freezing was typically 240 ± 15 F/min.

For the aluminum alloy samples the crucible and the thermocouple junction were lightly coated with red mud to inhibit attack on the steel. The melt surface was not protected.

## RESULTS AND DISCUSSION

A representation of typical cooling curves is shown in Fig. 1. Three forms of liquidus arrest are shown — one in which distinct supercooling occurs with recalescence to the liquidus temperature, one in which the arrest does not detectably deviate from horizontal and one in which the curve extends smoothly downward in temperature at a substantially reduced slope from the pre-liquidus portion.

For this discussion, the melt supercooling is considered to be the height of the recalescence pip in Fig. 1a, the difference in temperature between the lowest point reached before the reheating occurs and the highest subsequent point. This is not strictly correct, for some supercooling is certainly concealed in the curve in Fig. 1b. If no supercooling occurred, one would expect a curve of the third type (Fig. 1c).

### *Supercooling as a Grain Size Function*

A few data points are shown in Fig. 2 to illustrate the relation of supercooling to grain size. For Mg-Al alloys which supercooled more than 1 F (0.6 C), the grain diameter of the cooling curve sample was consistently more than 0.015 in. Conversely, for samples having an average grain diameter more than 0.015 in., supercooling of at least 1 F (0.6 C) was nearly always observed. Also, melts showing a supercooling of 4-7 F (2-4 C) usually had an average grain diameter (A.G.D.) greater than 0.1 in. Grain diameters for melts showing no detectible supercooling (less than 1 F) ranged from 0.005 in. to 0.015 in.

Four different grain refinement treatments were used for magnesium-base alloys, and in all cases measurable supercooling was eliminated when sufficient refinement occurred. The processes used were the addition of Zr to Mg-Zn, Mg-Th or Mg-rare earth, the addition of C to Mg-Al, the treatment with $Cl_2$ of Mg-Al and the superheating of Mg-Al.[6] Data points indicated as grain refined in Fig. 2 resulted from such treatments on AZ91.

All grain coarsening treatments tried were effective in producing observable supercooling. Mg-Al alloys which have been grain refined by superheating or carbon treatment can be coarsened by the addition of trace amounts of Be or Zr. Either addition yielded several degrees of supercooling in alloys containing more than 3 per cent Al.

Aluminum-base alloys can be conveniently grain coarsened by holding at high temperatures, and all Al system alloys examined containing more than four per cent solute showed at least 5 F (3 C) supercooling on holding at 1500 F (816 C) after chlorina-

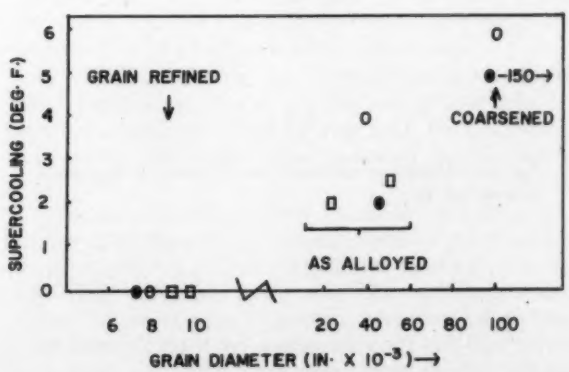



Fig. 2 — Supercooling vs. average grain diameter (AZ 91). Grain refinement — black circle, superheat; white circle — $Cl_2$; square — C.

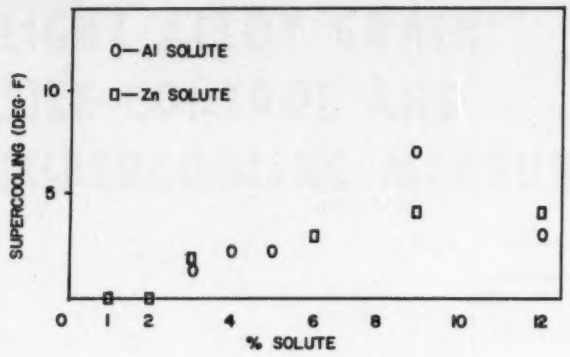



Fig. 3 — Maximum supercooling vs. binary composition Mg-base alloys.

tion at 1350 F (732 C). Examination of grain refinement treatments was not attempted for aluminum-base alloys, except to note that a previously overheated melt showed some refinement on holding at 1400 F (760 C). The refinement was easily seen from the cooling curves as well as metallographically.

### *Supercooling as a Solute Function*

Mention was made of the importance of solute concentration in determining how much supercooling might occur in a coarse grained alloy. To clarify the role of solute concentration, a series of alloys was prepared in the Mg-Al, Mg-Zn, Al-Si, Al-Cu and Al-Mg systems. An effort was made to grain coarsen these alloys as much as possible to determine the maximum observable supercooling as a function of solute concentration. The Mg-base alloys were coarsened by addition of Be and the Al-base alloys by chlorination and overheating. The data are shown in Figs. 3 and 4, where all data points refer to the observed supercooling for samples coarser than 0.1 in. A.G.D.

In no case was supercooling observed with pure Mg or Al, even though extremely coarse grain structures occurred. The maximum undercooling observed increases with composition up to some value where a leveling off or even a decrease may occur, as shown.

### *Cause of Supercooling*

The results presented can be explained in terms of the heat balance associated with the growth of suitable nuclei, and the various barriers to the growth of such nuclei.

In the pure metals and dilute alloys, crystallite growth rates can be extremely high. The growth of a few grains, perhaps initiated at the crucible wall, is sufficiently fast to liberate latent heat as fast as the sample is cooled. However, for higher concentration alloys, freezing tends to take place below the equilibrium temperature. Solute is rejected, according to phase diagram relationships, around the growing crystals.

This depresses the melting point, and if freezing is to proceed it must occur at a temperature corresponding to the liquidus temperature of this solute-rich region. This can allow the bulk liquid to undercool sufficiently for fresh nuclei to begin growing.[5,7]

If enough nuclei grow, they can liberate latent heat faster than the sample is cooled and the temperature may actually rise, approaching the equilibrium liquidus.

A temperature measurement taken just in the nucleation region would indicate the maximum supercooling resulting from these diffusion barriers to growth. It makes no difference then if nucleation occurs at the crucible wall since the solute-rich bound-

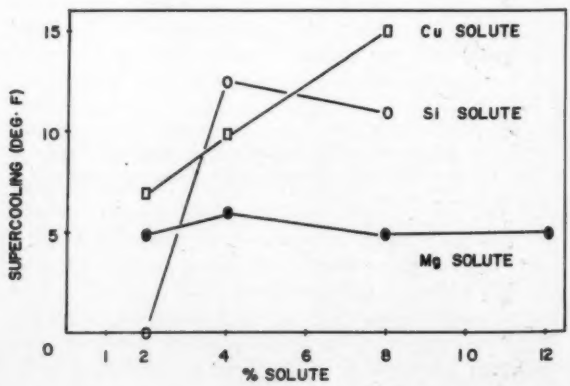



Fig. 4 — Maximum supercooling vs. binary composition Al-base alloys.

ary layer can isolate the crystals in that region. Hence, rather large supercooling effects can be observed even if the crucible material is a nucleation catalyst for the melt, and the occurrence of such effects in iron crucibles is not surprising, for high solute alloys.

Furthermore, it is of little consequence whether or not the thermocouple nucleates the melt. If the diffusion boundary is to be an effective barrier to crystal growth, the solidification of metal on the thermocouple wires must proceed at substantially the reduced temperature which characterizes the freezing of the balance, until recalescence begins.

The failure to observe supercooling in dilute alloys then is attributed to the failure to achieve stable diffusion boundaries. In more concentrated alloys, supercooling is observed unless the melt is well nucleated. If the melt is grain refined, enough growth sites are available with nominal supercooling to balance the rate of heat removal and permit freezing to proceed smoothly with no recalescence effect.

### *Validity of Measurements*

The isothermal liquidus arrests sometimes seen (Fig. 1b) require an evaluation of the maximum thermal gradient which may occur across the sample. Several attempts were made to measure this effect, and none was detected. The use of an unshielded thermocouple may permit shorting out, which would prevent measurement of the surface versus interior temperature so a trial calculation was made.

The maximum gradient may be calculated on a basis of uniform heat flow from a knowledge of the cooling rate. Since a sphere contains a maximum of volume per unit surface, the maximum gradient at the sample surface can be calculated based on the measured cooling rate assuming a spherical sample of equal volume. Using a cooling rate of 4 F/sec, Mg heat capacity of 0.32 cal/deg. C/gm, liquid Mg density of 1.6 gm/cm$^3$ and thermal conductivity of 0.2 cal/cm$^2$ deg. C/cm/sec, the surface gradient is calculated to be 3.1 C/cm. The radius for a sphere of 20 cm$^3$ volume is 1.7 cm, and, assuming a linear decrease in gradient to the center of the sphere, one calculates a maximum surface to center temperature difference of 5 F (2.7 C).

If recalescence should begin at the sample surface before that region is undercooled 5 F (2.7 C), then the reheating would prevent observation of any undercooling at the sample center. This effect would appear as an isothermal arrest on the cooling curve, as was often observed for moderately fine grain samples (Fig. 1b). It would also be possible that a sample showing 5 F (2.7 C) recalescence would actually have experienced an additional concealed supercooling of 5 F for a total of 10 F (5.4 C).

This concealed supercooling may then account for the failure to get a better correlation between grain size and cooling curve for samples finer than 0.015 in. A.G.D. There may be a wide range of grain sizes from 0.005-0.015 in. A.G.D. corresponding to such a concealed effect in the range 0-5 F (0-2.7 C). Undercooling greater than this has been associated in all cases with coarse grain samples.

### *Use of the Method*

The possible utility of this method as a control for grain diameter lies in its speed. If a melt is being held on the melt floor at say liquidus plus 200 F, then a supercooling measurement may be completed within one min of the time the sample is removed. This is even faster than the speed with which a fracture bar can be cast, broken and examined. Furthermore, this test can be quantitative, and, thus, can provide an objective measure for quality control under favorable circumstances.

The chief disadvantage of the method, apart from the equipment necessary, is that one may not have free choice of the control point. For example, Mg-Al alloys can only be categorized as coarse or not coarse from the direct measurement of recalescence. This is a result of the concealment of a certain degree of undercooling in relatively flat liquidus arrests. Work is now in progress to clarify this point and permit extension of the method, and current results show some promise.

Nevertheless, for a foundry or laboratory which has moderately high speed temperature recording equipment available, the cooling curve will distinguish the grain size effects mentioned. In the processing of any particular melt which must have grain size controlled by some transient process (as for example superheat grain refined Mg-Al melts), knowing whether or not the grain refinement treatment is present when the melt is ready to pour may save scrapped castings.

### *Binary Metal Analysis*

A further possibility exists for those who may have such equipment available. For many binaries it is possible to obtain a useful analysis from an estimate of the liquidus temperature of the heat. This may also be true of complex alloys if the control of one element is the major problem. Such an analysis may sometimes be used only as a temporary expedient to allow processing of a heat to proceed without waiting for a chemical or spectroscopic analyses.

For example, in Mg-Al alloys through most of the commercial range, one per cent Al will depress the liquidus about 12 F. The measurement of this temperature by the technique described will allow the estimation of Al to $\pm$ ½ per cent, if the temperature measurement is accurate to $\pm$ 5 F. With good, calibrated equipment it is certainly possible to do much better than this. However, such refinements are often not necessary in production foundry operations.

A foundry running gas control specimens might be in the best position to use this sort of tool. The recording equipment could be kept with the gas testing equipment and both units operated by the same technician.

This type of measurement has been used primarily in our laboratory in the investigation of grain refinement treatments. It has been found most valuable in following transient grain size effects wherein the timing of some melt treatment may depend on the melt response to a prior treatment. It has also saved a number of off-composition experimental melts.

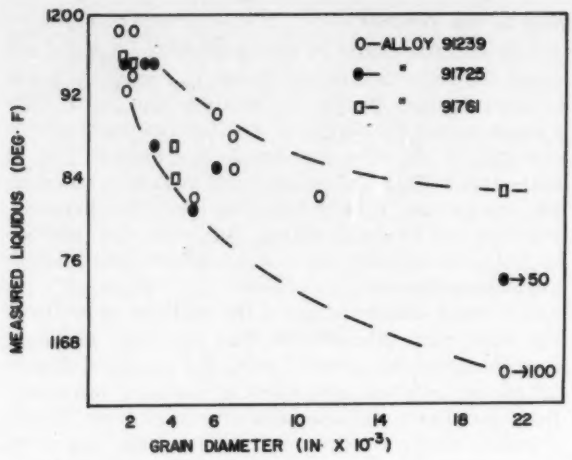



**Fig. 5 — Grain diameter of EZ33 vs. liquidus temperature.**

The estimation of the liquidus temperature has enabled the alloying errors to be found and corrected. It happens that this laboratory facility is not set up for routine analysis of melts before processing, as is done in the production facility. To do this type analysis in the laboratory requires holding a technician and his equipment available on a stand-by basis, which becomes excessively expensive for most routine experimental work.

### *MgZr Grain Size Control*

A final example of cooling curve utility in predicting grain size effects deals with MgZr alloys. Certain of these alloys, notable E33, are sensitive to grain size control for the achievement of optimum properties. Zirconium is added to control the grain size, and an addition of about ½-1 per cent Zr is sufficient to reduce the A.G.D. in this test specimen to ± 0.010 in. and eliminate the recalescence. Further additions of Zr allow reduction of grain size to 0.0015-0.003 in. in the particular specimen studied, with some further change in the shape of the cooling curve. More important, Zr in Mg is a peritectic grain refiner, and its addition substantially raises the liquidus temperature.

Figure 5 shows the liquidus temperature as a function of A.G.D. for EZ33. It is clear that a good correlation exists between the observed liquidus temperature and the metallographically determined grain size. The result is of interest primarily because of the good correlation between grain size and liquidus temperature for a quaternary alloy. Most foundrymen would probably not care to base their grain size control on the estimation of temperature to ± 2 F.

The first addition of Zr (one-half per cent in this series) is sufficient to eliminate supercooling as measured by recalescence. However, the steady progression in liquidus temperature with increasing Zr content allows the development of a much better correlation.

## CONCLUSION

A rapid open-air cooling curve test is sufficiently sensitive to detect supercooling effects of the order of 1-2 per cent (½-1 C) or more.

This amount of supercooling characterizes coarse grained alloys in the MgAl, MgZn, AlSi, AlCu and AlMg systems of commercial interest, and its measurement can be used as a tool for grain size control. Supercooling disappears in grain refined melts, and reappears when the metal is grain coarsened. Supercooling increases with increasing alloy content to a maximum which appears to be characteristic of the alloy system for coarse grained alloys.

There is no apparent reason why the method could not be extended to other alloys, providing the solute content is sufficient. Where this method is used to follow grain size effects, a crude liquidus temperature estimate is also provided, enabling melt analysis to be read from the phase diagram for simple alloys.

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# PERMANENT MOLD DESIGN

by John Wall


## ABSTRACT

**The advantages of the permanent mold process are detailed by the author. He covers such items as gating and risering design, mold coatings use and shrink factors, as well as some automotive uses of this process.**


## INTRODUCTION

The principal of permanent mold casting has changed little since one of our prehistoric ancestors first cast a crude axe head by filling a cavity between two stones with molten metal. The technical advances have, however, been vast.

Permanent molding, or gravity die casting, as it is more properly called in the European countries, is used in the production of both nonferrous and ferrous castings. We are concerned only with the nonferrous metals, and the general term permanent molding includes all the various phases and techniques in the process.

When the permanent mold is used with sand forming part of the casting, it is called semi-permanent molding. When the mold is spinning as the molten metal is introduced, it is called centrifugal casting. When the metal is forced into the mold by pressure on the surface of the molten bath, it is called low pressure die casting. When the molten metal is poured into the hinge part of a book type mold, and the mold is closed upon the molten metal, it is called compression molding.

While trying not to indicate the superiority of one method over the other, there are distinct advantages for using each process in certain definite applications.

## PERMANENT MOLD ADVANTAGES

The permanent mold process is used with advantages for a number of reasons, but usually to cast large numbers of parts that require closer tolerance and better surface finishes than can be obtained by sand casting. As the chilling rates are much higher in the permanent mold, it produces a casting with better metallurgical characteristics and higher mechanical properties than can be produced with sand molds. This same rapid chilling rate also increases the production rate of the permanent mold.

Most of the casting alloys will show their maximum mechanical properties when gravity poured into a metal mold. The ideal gating of the permanent mold will fill the cavity with a minimum of turbulence to reduce the absorption of gases, dross formation and mold erosion. It should regulate the rate of metal entry into the cavity, to establish the best possible temperature gradients to promote progressive solidification and produce a casting with minimum excess metal in the gates and risers.

The design of the gating and risering of a permanent mold casting is becoming increasingly important, not only to produce a sound casting, but for ease in gating removal by automatic trimming devices. This becomes important in the high production of automotive parts.

The gating must also be designed to be adaptable to automatic ladling, which is becoming increasingly popular as the size of the castings increase, making manual handling of large amounts of molten metal troublesome.

This can be accomplished in various ways, usually by designing the gating with a runner system allowing the molten metal to be poured into a common funnel, and distributing the metal to the various ingates. In some cases, especially in long thin castings, particularly those requiring sand cores, it is beneficial to allow the mold to be tilted during the casting process to reduce turbulence and provide a more suitable position to install the sand cores. In these instances, a refractory lined and heated ladle can be installed integral with the mold. A measured amount of molten metal can be automatically poured into the ladle when in the horizontal position. As the mold is tilted from the horizontal position at a controlled rate, the molten metal is automatically poured into the cavity until the casting cycle is completed and the mold is in the vertical position. This type of mold is easily automated and adaptable to a single pouring station, rotary table type of casting device.

### *Automated Permanent Molds*

Since the use of large permanent mold castings in the automotive industry has been greatly increased, the design of the permanent mold has changed from the simple hand operated, low productive device, to a highly automated self contained casting machine, competitive, especially in the larger castings, with the production rates of pressure die casting. The inconsistencies of the human element have been removed by automatic ladling, mechanical timers and automatic mechanical mold operation. It is possible for the modern permanent mold device to contain automatic insert heaters, dispensers and core extractors. Automatic ejection and casting unloading is common.

The use of mold coatings, which is so necessary in the permanent mold process, still requires some artistry in application. The proper use of the proper materials can easily determine the success or failure of the process. There are many commercial mold coatings now available. At one time each foundry made their own secret formulas, and helped retard the growth of the permanent mold by intimating there was a secret to the operation of a successful permanent mold. There is no secret formula for the operation of any casting operation other than just

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plain common sense. There are many successful permanent mold foundries in operation today, that prior to their entering this field had no practical foundry experience.

There are several reasons why the use of mold coatings are necessary. Basically, they are to prevent erosion of the mold, provide a release agent, insulate the molten metal against chilling too quickly and provide chilled areas to promote progressive solidification.

Mold coatings that are of the insulating type usually contain whiting, kaolin or china clay in a solution of sodium silicate and water. The chill coatings contain graphite. Each permanent mold may require mold coatings with specific properties so the formula is varied with the application. The mold is usually heated to approximately 400 F (204 C), the mold coating sprayed on in thin layers to the required depth, rarely greater than 0.010 in. In localized areas requiring chilling, the mold coating is removed entirely with a wire brush, steel wool or scraper or the thickness of the coating is reduced to a minimum to prevent mold erosion.

The gating and risering are usually heavily coated for maximum insulation. This can be sprayed on or painted on by hand with an ordinary small paint brush.

In a normal permanent mold, no further heating or chilling is required. However, in some instances copper, brass or air chills are needed in localized areas to promote solidification. Because the molds are usually made of heat resistant cast iron, water chilling is seldom used.

In some cases, local areas require additional heat to increase fluidity and prevent misruns and cold shuts. This can be accomplished by the use of gas burners or electrical heaters mounted to the outside of the mold sections in the area requiring the heat.

### *Automotive Piston Molds*

In special cases, as in the highly developed permanent molds for high production such as the automotive piston, various unique devices are used to insure continuous trouble free operation. Many of the devices although, they are peculiar to piston casting, are worth mentioning as they can be applied to other methods of casting.

With few exceptions the automotive piston is cast with a collapsible core made of 5 sections. Depending on the design of the piston and the clearances allowed, the collapsing sequence of the core is a cleverly designed and accurately machined tool, allowing the center section to collapse and be easily extracted from the casting. One side section is then moved into the void created, in many cases with little clearance, and then removed from the casting, allowing the other side sector enough clearance to be removed readily.

The core segments are made of H-13 type hot work die steel with the wear surfaces and keys nitrided to prevent galling. The center part of the 3 piece center core is water cooled, the other parts of the core being cooled by convection.

The outer surface of the piston is formed by a rather simple 2 piece parallel actioned mold. The upper part of this mold forming the ring belt area is usually made of a mild steel allowing this part to be water cooled. The lower part is usually made of heat resistant type of cast iron. This part may contain some complex inserts, and provides locations for the wrist pin cores. Because the outer molds contain the gating and risers and is water cooled, some provision must be made to insulate the gating area from the chilled area. This is accomplished by a simple saw cut divorcing the area adjacent to the gates and risers from the chilled area.

### *Shrink Factors*

Because the operating temperature of the permanent mold segments may vary considerably, greater attention must be paid to the design of the mold in regard to the clearances, shrinkages and mold register than in most other types of casting. Often it is necessary to use several shrink factors for different areas in the same casting. It is normal for adjacent mold sectors to operate at a difference in temperature as much as 400-500 F (244-260 C), making operating clearances a design problem.

Because of this same temperature difference, great attention in the design of the register of the mold components must be made to promote accuracy and continued trouble free operation.

Although emphasis may be placed on the selection of materials to reduce warpage of the mold sectors during operation, it is common for the various components to warp, twist and grow, especially when new. Usually the mold will stabilize after several operations and little further warpage or growth will be experienced. Normally, the mold is adjusted or reworked at this point in case the casting has been affected dimensionally by the warpage of the segments.

Constant vigilance must be maintained as it is a common human trait to spray the mold with mold coating often to correct small defects that may appear in local areas. This leads to a gradual build up of mold coating in certain areas, affecting the dimensional accuracy of the casting.

The ultimate goal of many high production castings is, of course, pressure die casting. However, as mentioned previously, most of the casting alloys will show their maximum qualities when gravity poured. This and a number of other reasons places in doubt the advantage of pressure die casting such castings as the V8 engine block. The block casting can be designed to eliminate all the sand cores and collapsible cores to make it possible to pressure cast. However, the production rate of the pressure cast would not be greatly improved over the permanent mold, especially where inserts must be heated and cast in place, such as bearings and cylinder sleeves.

Although recently there is a lot of interest shown in the pressure casting of the aluminum cylinder blocks for automotive engines, the permanent mold process seems secure in its position. This is mainly because of its lower initial cost, superior properties of the casting, ability to cast heat treatable alloys and a greater selection of castable alloys, shorter lead time, necessary to produce the mold and the improved cycle time of the modern automated mold.

# INOCULATION INFLUENCE ON GRAY IRON STRUCTURE AT VARIOUS CARBON EQUIVALENCES

Progress Report
Gray Iron Division
Sponsored by
AFS Training & Research Institute

by H. D. Merchant, L. I. Toriello and J. F. Wallace

## ABSTRACT

This project was conducted to investigate the influence of an effective inoculant on the structure of gray iron, particularly the graphitic structure and eutectic cell size. The materials and interest in this work were obtained from a research project currently sponsored at Case Institute of Technology by the AFS Training and Research Institute and performed under the direction of the Research Committee of the Gray Iron Division. The research reported in this paper is a sequel to the work reported recently[1] on the effect of inoculation on the risering of gray iron which was also sponsored by the AFS Training and Research Institute. This prior work indicated that more mold wall movement occurred in green sand molds with inoculated compared to uninoculated gray iron. The cause of this greater wall movement was hypothesized to be the more mushy type of solidification and finer eutectic cell size of the inoculated iron. This study was undertaken to explore this hypothesis.

## INTRODUCTION

A ladle addition made as an inoculant may affect the mechanical properties of gray iron by influencing eutectic cell size, graphitic pattern and metallic matrix. It may also have some alloying influence whereby it may change composition and theoretical solid-liquid and solid-solid transformation temperatures. The effectiveness of an inoculant is measured by considering all these influences on a solidified casting. The purpose of this study was to examine the effects of additions of a known effective inoculant of 0.5 per cent Si as FeSi containing appreciable aluminum and calcium on eutectic cell size, graphitic pattern and metallic matrix in various grades of unalloyed hypoeutectic and hypereutectic gray irons.

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To permit a direct comparison of the inoculated and uninoculated irons without an alloying influence from the silicon addition, it was compared with gray iron of the same composition which was alloyed with 0.5 per cent silicon metal. The latter has been shown to be a poor inoculant.[2] The subject of inoculation of gray iron has been widely investigated, and is the subject of many excellent publications.[3-5]

The results obtained in this work, however, only apply to the particular procedure employed, such as melting practice, super-heating and pouring temperature, inoculation procedure, etc. Variations in these factors will influence the inoculation effect at various carbon equivalences to a marked extent. The use of the high frequency induction furnace undoubtedly influenced these results. An effort was made to standardize these conditions in this work so that the influence of only composition could be determined. Accordingly, one should not expect to find similar results at the same carbon equivalences unless these other variables are similar in the iron poured in the foundry in question.

## PROCEDURE

The casting employed in this investigation was a cylindrical bar 2 in. diameter and 9 in. long. It was cast horizontally utilizing a blind riser located at one end of the casting. Riser and riser neck were dimensioned according to formulas developed in a recent investigation, and were designed to be adequate for the size of casting and type of iron poured.[6] The casting was single gated tangentially into the riser, utilizing optimum pouring times and gating systems recommended by another recent report.[7]

Forty-six heats were melted within a carbon equivalent range of 3 to 4.7 per cent. All iron was melted in a 30 lb induction furnace. The metal charge consisted of various amounts of pig iron, steel and gray iron scrap. The final chemical composition of the heats is presented in the table. The heats have been arranged in order of increasing carbon equivalence in this table. The furnace and pouring temperatures were approximately 2700 and 2550 F (1482 and 1399 C), respectively, as checked by Pt – Pt-10 per cent Rh thermocouple. Care was taken to control tapping, inoculating and pouring temperatures as closely as possible. Such control is essential because of their

observed influence on the mode of solidification.[8,9]

The 46 heats were melted in pairs with about the same composition for each pair. Sufficient silicon metal to add 0.5 per cent silicon was added to the first heat in each pair at least 15 min before pouring; 0.5 per cent silicon, as 85 per cent grade ferrosilicon containing significant quantities of aluminum and calcium, was added to the furnace in the second heat of each pair just before pouring. After the additions were plunged into the melt, the molten iron was stirred vigorously with a steel rod for a few seconds, and allowed to drop to the desired temperature be-

**DATA ON COMPOSITION, BRINELL HARDNESS AND EUTECTIC CELL COUNT FOR INOCULATED AND UNINOCULATED IRONS OF VARIOUS CARBON EQUIVALENCES**

| Pair No. | Treatment | Composition,[1] % | | | | Bhn | Avg. No. of Cells/sq in. | No. of Cells/sq in. at Various Distances from the Casting Surface, in. | | | | | |
|---|---|---|---|---|---|---|---|---|---|---|---|---|---|
| | | C | Si | C.E.[2] | E.G.[3] | | | 0.08 | 0.34 | 0.41 | 0.58 | 0.73 | 0.88 |
| 1 | Si[A,B] | 2.61 | 1.49 | 3.11 | 0.76 | 401 | — | — | — | — | — | — | — |
| | FeSi | 2.58 | 1.46 | 3.07 | 0.73 | 248 | 1210 | 1306 | 1004 | 977 | 904 | 684 | 450 |
| 2 | Si | 2.79 | 1.52 | 3.30 | 0.94 | 229 | 285 | 316 | 247 | 80 | 207 | 127 | 170 |
| | FeSi | 2.73 | 1.72 | 3.30 | 0.90 | 217 | 975 | 1138 | 880 | 650 | 493 | 445 | 400 |
| 3 | Si | 2.89 | 1.55 | 3.41 | 1.05 | 229 | 570 | — | — | — | — | — | — |
| | FeSi | 2.80 | 1.85 | 3.42 | 0.99 | 229 | 907 | 958 | 804 | 835 | 560 | 356 | 435 |
| 4 | Si | 2.95 | 1.54 | 3.46 | 1.10 | 197 | 292 | 375 | 268 | 203 | 193 | 190 | 160 |
| | FeSi | 2.95 | 1.69 | 3.51 | 1.12 | 204 | 880 | 920 | 750 | 620 | 525 | 468 | 423 |
| 5 | Si | 2.97 | 1.75 | 3.55 | 1.15 | 187 | 384 | 510 | 208 | 240 | 235 | 215 | 183 |
| | FeSi | 3.03 | 1.76 | 3.62 | 1.21 | 192 | 844 | 950 | 783 | 560 | 413 | 338 | 342 |
| 6 | Si | 2.92 | 2.08 | 3.61 | 1.13 | 197 | 445 | 498 | 418 | 255 | 266 | 156 | 263 |
| | FeSi | 3.04 | 1.99 | 3.70 | 1.24 | 201 | 875 | 910 | 793 | 625 | 525 | 408 | 380 |
| 7 | Si | 3.05 | 2.19 | 3.78 | 1.27 | 179 | 480 | 590 | 418 | 279 | 249 | 224 | 149 |
| | FeSi | 2.95 | 2.17 | 3.67 | 1.09 | 197 | 1070 | 1115 | 960 | 665 | 635 | 608 | 411 |
| 8 | Si | 2.84 | 2.60 | 3.71 | 1.10 | 217 | 422 | 474 | 348 | 300 | 331 | 224 | 193 |
| | FeSi | 2.84 | 2.45 | 3.66 | 1.09 | 232 | 1031 | 1196 | 995 | 772 | 636 | 565 | 395 |
| 9 | Si | 3.01 | 2.38 | 3.80 | 1.25 | 137 | 646 | — | — | — | — | — | — |
| | FeSi | 3.14 | 2.37 | 3.93 | 1.38 | 167 | 935 | 1013 | 790 | 438 | 515 | 540 | 343 |
| 10 | Si | 3.12 | 2.37 | 3.91 | 1.36 | 146 | 525 | 673 | 455 | 352 | 194 | 182 | 182 |
| | FeSi | 3.16 | 2.32 | 3.93 | 1.40 | 170 | 985 | 1063 | 825 | 630 | 590 | 510 | 343 |
| 11 | Si | 3.16 | 2.31 | 3.93 | 1.39 | 157 | 570 | 611 | 553 | 361 | 405 | 340 | 264 |
| | FeSi | 3.17 | 2.27 | 3.93 | 1.40 | 163 | 770 | 812 | 745 | 745 | 360 | 319 | 360 |
| 12 | Si | 3.62 | 1.21 | 4.02 | 1.74 | 170 | 236 | — | — | — | — | — | — |
| | FeSi | 3.64 | 1.29 | 4.07 | 1.77 | 159 | 316 | — | — | — | — | — | — |
| 13 | Si | 3.24 | 2.55 | 4.09 | 1.50 | 157 | 515 | 578 | 465 | 288 | 303 | 244 | 229 |
| | FeSi | 3.15 | 2.29 | 3.91 | 1.38 | 164 | 735 | 918 | 685 | 345 | 322 | 179 | 385 |
| 14 | Si | 3.47 | 1.96 | 4.12 | 1.67 | 126 | 358 | 495 | 357 | 201 | 129 | 153 | 160 |
| | FeSi | 3.39 | 1.92 | 4.03 | 1.58 | 207 | 405 | — | — | — | — | — | — |
| 15 | Si | 3.51 | 2.06 | 4.20 | 1.71 | 140 | 310 | 388 | 330 | 182 | 129 | 122 | 137 |
| | FeSi | 3.53 | 2.18 | 4.26 | 1.75 | 107 | 210 | — | — | — | — | — | — |
| 16 | Si | 3.35 | 2.60 | 4.30 | 1.61 | 106 | 465 | 655 | 216 | 260 | 250 | 231 | 215 |
| | FeSi | 3.33 | 2.91 | 4.67 | 1.62 | 108 | 560 | 615 | 535 | 426 | 382 | 295 | 160 |
| 17 | Si | 3.45 | 2.16 | 4.17 | 1.67 | 130 | — | — | — | — | — | — | — |
| | FeSi | 3.55 | 2.52 | 4.39 | 1.72 | 106 | 675 | 820 | 483 | 508 | 540 | 583 | 445 |
| 18 | Si | 3.53 | 2.54 | 4.38 | 1.72 | 103 | 264 | 328 | 177 | 189 | 184 | 229 | 103 |
| | FeSi | 3.45 | 2.93 | 4.43 | 1.63 | 93 | 595 | 645 | 543 | 319 | 395 | 300 | 194 |
| 19 | Si | 3.09 | 3.97 | 4.41 | 1.39 | 123 | 263 | 355 | 212 | 156 | 165 | 145 | 114 |
| | FeSi | 3.08 | 4.17 | 4.47 | 1.34 | 116 | 510 | 622 | 397 | 345 | 350 | 190 | 320 |
| 20 | Si | 3.47 | 3.35 | 4.59 | 1.53 | 126 | 282 | — | — | — | — | — | — |
| | FeSi | 3.37 | 3.27 | 4.46 | 1.56 | 109 | 426 | 550 | 413 | 293 | 194 | 137 | 91 |
| 21 | Si[B] | 3.68 | 2.96 | 4.67 | 1.62 | 99 | — | — | — | — | — | — | — |
| | FeSi[B] | 3.71 | 2.70 | 4.61 | 1.68 | 103 | — | — | — | — | — | — | — |
| 22 | Si[B] | 3.69 | 2.55 | 4.54 | 1.71 | 98 | — | — | — | — | — | — | — |
| | FeSi | 3.60 | 2.95 | 4.58 | 1.62 | 96 | 582 | 626 | 565 | 464 | 435 | 374 | 422 |
| 23 | Si[B] | 4.25 | 1.29 | 4.68 | 2.02 | 116 | — | — | — | — | — | — | — |
| | FeSi[B] | 4.23 | 1.30 | 4.66 | 2.00 | 131 | — | — | — | — | — | — | — |

NOTES:

Si — 0.5% Si added as silicon metal.
FeSi— 0.5% Si added as FeSi just before pouring.
1 — All heats contained approximately 0.05% P, 0.085% S and 0.80% Mn.
2 — % C.E. = % Carbon equivalent = %C + ⅓ Si.
3 — % E.G. = % Eutectic graphite = %C + 0.1% Si − 2.0 (hypoeutectic).
= 2.3 − 0.23% Si (hypereutectic).
A — Solidified white.
B — Eutectic cells could not be revealed.

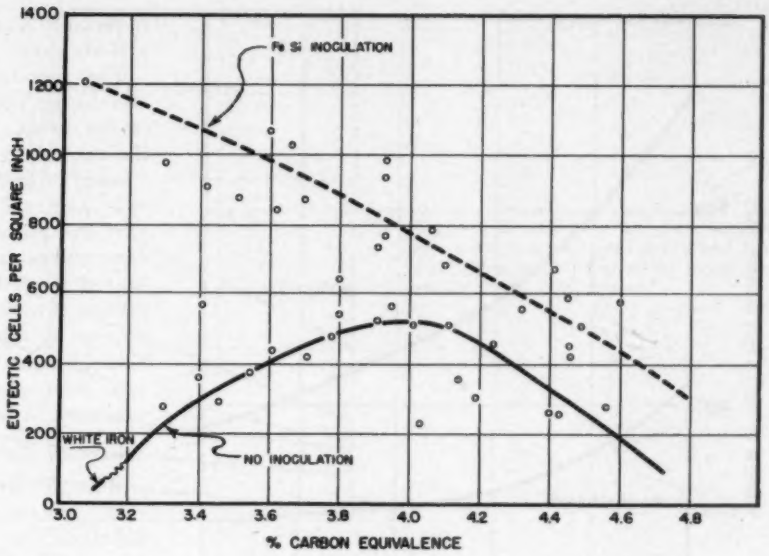



Fig. 1 — Relationship between eutectic cell count and carbon equivalence for uninoculated and inoculated cast irons.

fore pouring directly into the mold. The chemical analysis of silicon metal and 85 per cent grade ferrosilicon is:

| Inoculant | Si | Al | Ca | Fe |
|---|---|---|---|---|
| Silicon Metal, % | Rem. | 0.05 | 0.001 | 0.5 |
| 85% Ferrosilicon, % | 82.54 | 1.96 | 0.92 | Rem. |

All molds were made in green sand on a jolt-squeeze molding machine under equivalent conditions. The molds were compacted in slip flasks, stripped and jacketed for pouring. At least a 2 in. layer of sand surrounded all parts of the casting to prevent rapid loss of heat at any location. The castings were allowed to solidify without moving, and were cooled at least one hr in the mold before shakeout. The green sand employed was a synthetic type utilizing Michigan Lake sand (AFS 50 fineness) and containing about 4 per cent moisture and 5 per cent Western bentonite bond.

### *Composition, Hardness, Structure Data*

After cooling to room temperature, all castings were sand blasted and processed to obtain the desired data on composition, hardness and structure. Two ½-in. thick transverse sections were cut from the longitudinal center of each casting. One piece was employed to determine the eutectic cell size after proper polishing and etching by special techniques. The same piece was later used to ascertain the Brinell hardness at a location half-way between the surface and center of the casting.

A ⅝-in. wide, 1 in. long and ⅜-in. thick radial piece was cut from the second piece to examine the microstructure from the center to the surface of the casting. A representative microphotograph was taken approximately half-way between surface and center of the casting. The graphitic pattern and matrix structures were examined in each case.

## RESULTS AND DISCUSSION

The eutectic cell count (number of cells/sq in.) for the uninoculated and inoculated irons is plotted as a function of per cent carbon equivalence (C.E.) and per cent eutectic graphite (E.G.) in Figs. 1 and 2, respectively. The method of computing carbon equivalence and eutectic graphite is shown in the table. For uninoculated iron, the cell count increases with carbon equivalence or eutectic graphite, attains a peak at about 4.0 per cent carbon equivalence or at 1.35 per cent eutectic graphite, and then gradually decreases.

For inoculated iron, on the other hand, the cell count is initially high for low carbon equivalent irons, but the cell size gradually becomes larger or the cell count becomes smaller with increasing carbon equivalence or eutectic graphite. The entire casting section is white below about 3.2 per cent carbon equivalence in uninoculated iron, but it remains completely gray for as low as 3 per cent carbon equivalence when effective inoculation is employed.

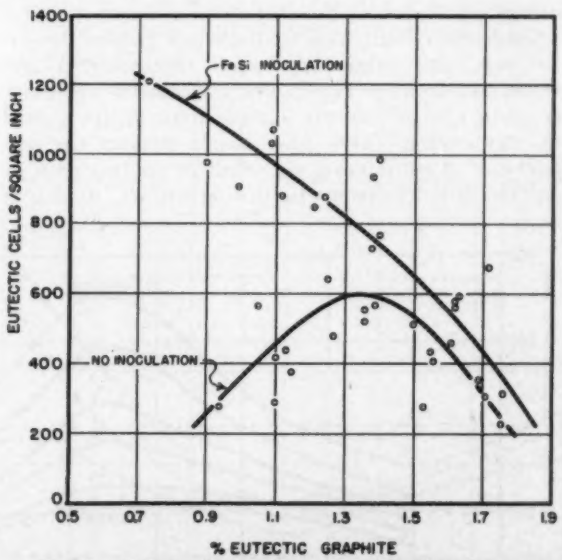



Fig. 2 — Relationship between eutectic cell count and eutectic graphite in uninoculated and inoculated cast irons.

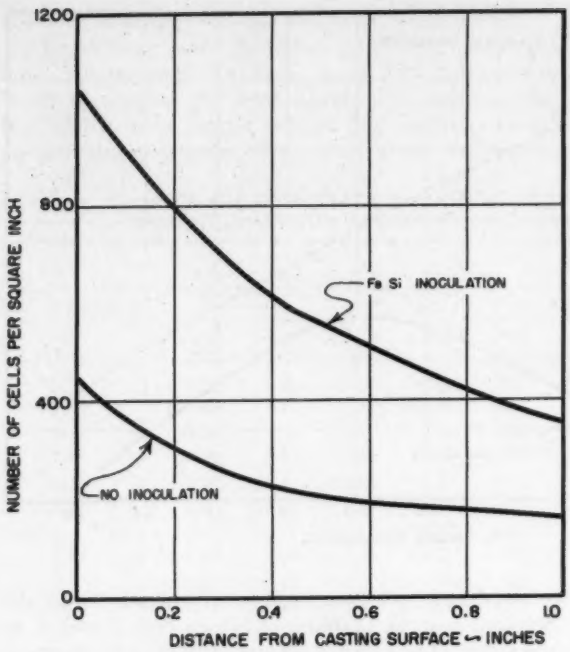



**Fig. 3 — Relationship between eutectic cell count and distance from casting surface for a typical set of uninoculated and inoculated cast irons with carbon equivalence of 3.6 per cent.**

The addition of 0.5 per cent silicon, as 85 per cent ferrosilicon containing appreciable amounts of aluminum and calcium, as an inoculant is highly effective in reducing the eutectic cell size of cast iron for low carbon equivalent irons. Effectiveness of the inoculant, as shown in Figs. 1 and 2, diminishes gradually as the carbon equivalence or eutectic graphite increases.

A macroscopic survey of a typical casting section of the 2 in. diameter cylinder for both inoculated and uninoculated irons shows a fine eutectic cell or a high cell count at the surface that becomes coarse towards the center. This is illustrated graphically for 3.6 per cent carbon equivalent uninoculated and inoculated irons in Fig. 3. The inoculant is apparently more effective at the surface than at the center of the casting. This observation stresses the importance of considering the effect of an inoculant at specific distances from the casting surface, so that it is possible to estimate its effectiveness in castings of various sizes.

The eutectic ceil count at various distances from the casting surface is tabulated in the table for most of the heats. For the whole range of compositions, the eutectic cell count at various distances from the casting surface is plotted in Fig. 4 for uninoculated iron, and in Fig. 5 for inoculated iron. It is observed that the pattern of variation in eutectic cell count with composition is similar in all parts of the casting; the cell count is highest at about 4.0 per cent carbon equivalence for the uninoculated iron and decreases for lower and higher values. The extent of this difference in the cell count is greater at the surface than at the center of the casting for all compositions in both inoculated and uninoculated irons.

### *Inoculant Effectiveness*

The effectiveness of the inoculant, expressed as per cent increase in cell count due to inoculation, can be calculated for the whole range of compositions from Figs. 4 and 5. The results of these computations are plotted in Fig. 6 for the surface and for the center of the castings. It is observed that the inoculant is most effective in low carbon equivalent irons and least effective in about 4.0 per cent carbon equivalent irons. As the carbon equivalence increases, the effectiveness of the inoculant decreases, attains a minimum at about 4.0 per cent and then increases slightly with higher carbon equivalence in the hypereutectic range.

Figure 6 also demonstrates that the effectiveness of the inoculant is greater at the surface than at the center of the casting. The difference in effectiveness of the inoculant is more in low carbon equivalent irons than in high carbon equivalent irons. For the 2 in. diameter circular section, the effectiveness of the inoculant at the surface is about twice as great as at the center in all irons with less than 4.0 per cent carbon equivalence, and is about 1½-times as great in the irons with higher than 4.1 per cent carbon equivalence. The average effectiveness of the inoculant projected over the entire casting section, is shown as a dotted curve between the surface and center curves. It is apparent that the average effectiveness curve will decrease as section thickness increases.

Because of the tendency of primary graphite to float in hypereutectic irons, discrete variations in composition sometimes occur. This results in uneven distribution of eutectic cell sizes across the casting sec-

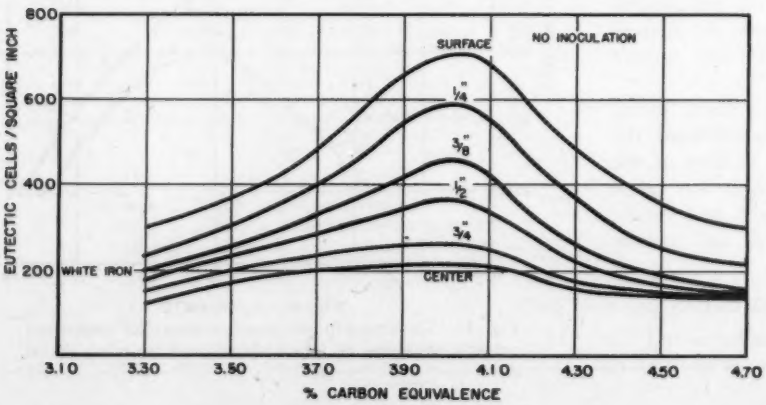



**Fig. 4 — Relationship between eutectic cell count, carbon equivalence and distance from casting surface for uninoculated cast irons in 2 in. rounds.**

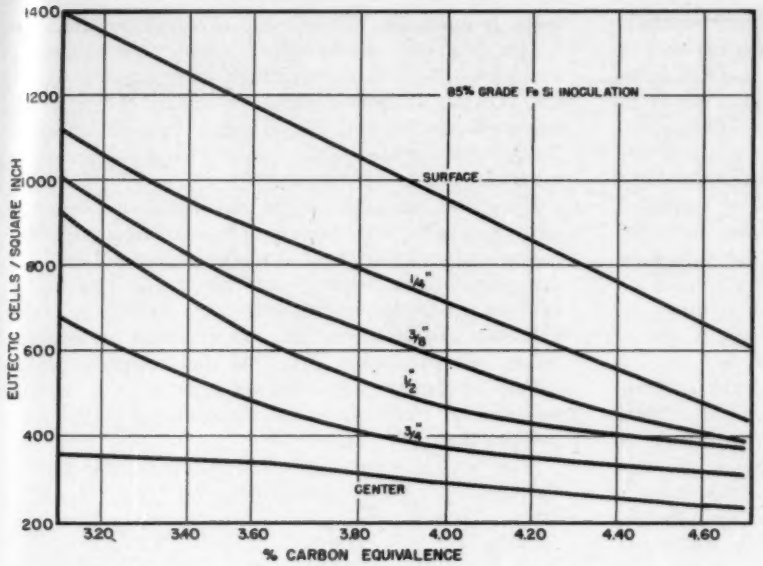



Fig. 5 — Relationship between eutectic cell count, carbon equivalence and distance from casting surface for inoculated cast irons in 2 in. rounds.

tion. Figure 7 shows three such anomalies. Figure 7A illustrates fine cells at the surface and near the top of the horizontal casting. The cell size gradually becomes coarse towards the bottom portion of the casting. This anomaly suggests flotation of nuclei or segregation towards the top of the casting during solidification.

Figure 7B contains fine eutectic cells at the center, and the cells gradually become coarse towards the surface of the casting. This suggests the migration of nuclei or nucleating conditions towards the center of the casting during solidification. A third type of anomaly, found more often in hypoeutectic irons, is shown in Fig. 7C. Eutectic cells are fine at the surface, but they assume maximum coarseness half-way between surface and center of the casting. The center contains both fine and coarse cells. The size, shape and distribution of graphite also change in uninoculated and inoculated irons as the carbon equivalence is varied.

These changes are illustrated by the various microphotographs in Fig. 8. The microstructures examined were located at midlength of the 8 in. long, 2 in. diameter bar half-way between the surface and center. Low carbon equivalent, hypoeutectic, uninoculated iron (3.30 per cent C.E.) shows highly interdendritic, fine, Type E graphite (Fig. 8A); inoculation of the same iron also results in an interdendritic location but the flakes are much thicker and in some respects like Type A graphite (Fig. 8B).

Medium carbon equivalent, uninoculated, hypoeutectic iron (3.75 per cent C.E.) contains a slightly interdendritic, short to medium length randomly oriented, Type A graphite with a few isolated areas of fine Type D graphite (Fig. 8C); inoculation eliminates Type D graphite, and produces a slightly longer and thicker, less branched Type A graphite (Fig. 8D). Higher carbon equivalent, uninoculated, hypoeutectic iron (4.10 per cent C.E.) has long, thin, branched Type A graphite (Fig. 8E); inoculation gives slightly shorter and thicker, unbranched, Type A graphite (Fig. 8F).

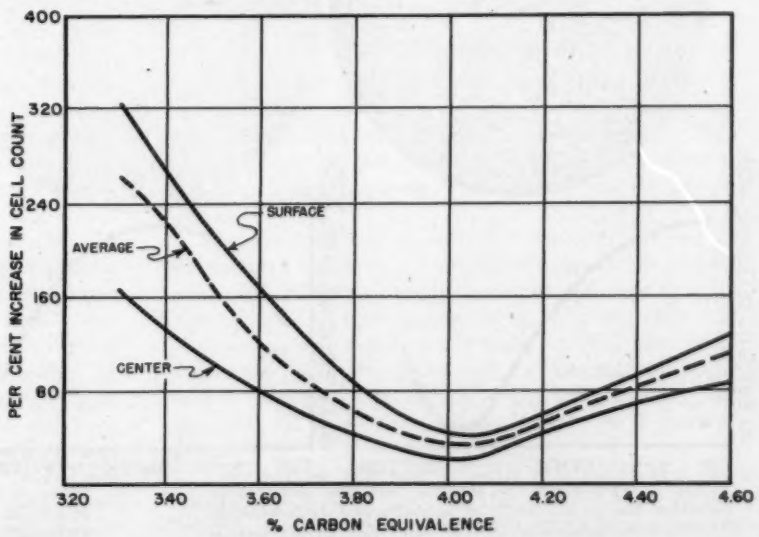



Fig. 6 — Increase in eutectic cell count from inoculation for various carbon equivalences at the surface, center and for the average in a 2 in. round.

From the foregoing observations, one may conclude that, for the hypoeutectic range of cast irons, as the carbon equivalence is increased in uninoculated iron, the graphitic pattern changes from Type E or D to short Type A to long Type A graphite. Inoculation makes the appearance of the dendritic pattern less pronounced and produces randomly distributed, Type A graphite. As the carbon equivalence is increased in these inoculated irons, the graphitic pattern varies from short and thick Type A graphite flakes to long and comparatively thin Type A graphite flakes.

The effect of inoculation and carbon content on the eutectic graphite in hypereutectic iron is not as clear-cut as in the hypoeutectic case but does show a trend. The hypereutectic irons close to the eutectic have a similar eutectic structure to the high carbon, hypoeutectic irons, i.e., long, thin, branched Type A graphite in the uninoculated iron, similar to that shown in Fig. 8E. When this hypereutectic iron is inoculated, a slightly shorter and thicker unbranched Type A graphite similar to that in Fig. 8F is obtained.

As the carbon content of the hypereutectic irons is increased, the eutectic structures appear similar to the lower carbon hypoeutectic irons. Thus, when the carbon content of hypereutectic irons is high, many areas of Type D graphite appear, particularly in the uninoculated irons. It appears as though the kish graphite solidification was removing some of the nucleation sites for eutectic graphite solidification, and the kish does not appear to act as nucleation sites for eutectic graphite solidification.

It can be concluded from these results that as the carbon equivalence of uninoculated hypoeutectic irons is increased, the cell size becomes smaller. The larger cell size of the low carbon equivalent uninoculated irons is accompanied by a Type D graphite. As the carbon content increases, the cell size becomes smaller and the graphite changes to fairly long, thicker, unbranched flakes. This is in agreement with the results of Adams.[10]

However, the cell size of inoculated hypoeutectic irons increases with higher carbon equivalence. The lower carbon equivalent irons have a small cell size and short, thick Type A graphite flakes. The higher carbon equivalent inoculated irons exhibit a larger cell size with thinner and longer Type A graphite flakes. The resulting flakes in the inoculated, high carbon, hypoeutectic irons are only slightly shorter and thicker than the uninoculated irons of the same composition.

The per cent eutectic graphite in gray iron may be calculated by these equations:

$$\% \text{ E.G.} = \% \text{ C} + 0.1\% \text{ Si} - 2.0 \quad \text{(hypoeutectic iron)}$$
$$= 2.3 - 0.23\% \text{ Si} \quad \text{(hypereutectic iron)}$$

The per cent eutectic graphite depends upon carbon and silicon contents in hypoeutectic irons, but only upon silicon content in hypereutectic irons. An improved correlation can be established between per cent eutectic graphite and graphitic pattern than between per cent carbon equivalence and graphitic pattern. For uninoculated irons containing less than about 1.35 per cent eutectic graphite, increasing eutectic graphite has an effect of changing the graphitic pattern from Type D to Type A; for all irons containing more than 1.35 per cent eutectic graphite, increasing eutectic graphite has an effect

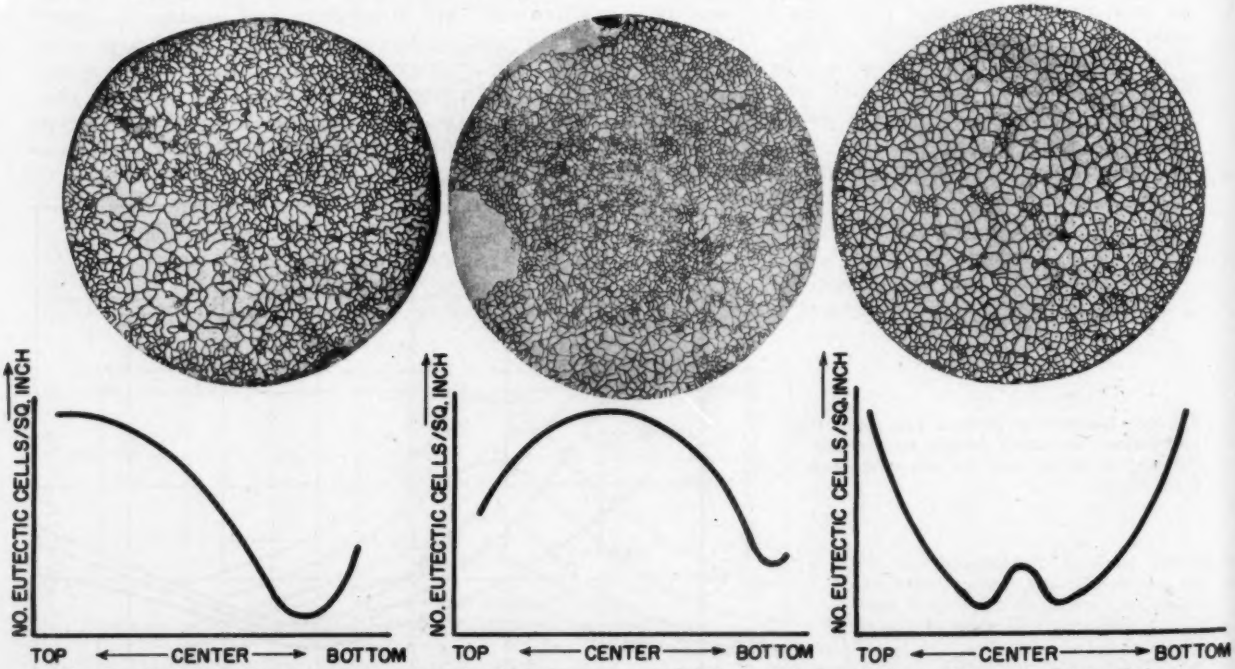



Fig. 7 — Variations in eutectic cell size distribution across the 2 in. round in uninoculated hypereutectic gray iron. Schematic plot of cell count across section below, in each case. *Left to right* — A — Fine cell size at top, B — fine cell size at longitudinal center, C — Fine cell size at top and bottom, discontinuous at center.

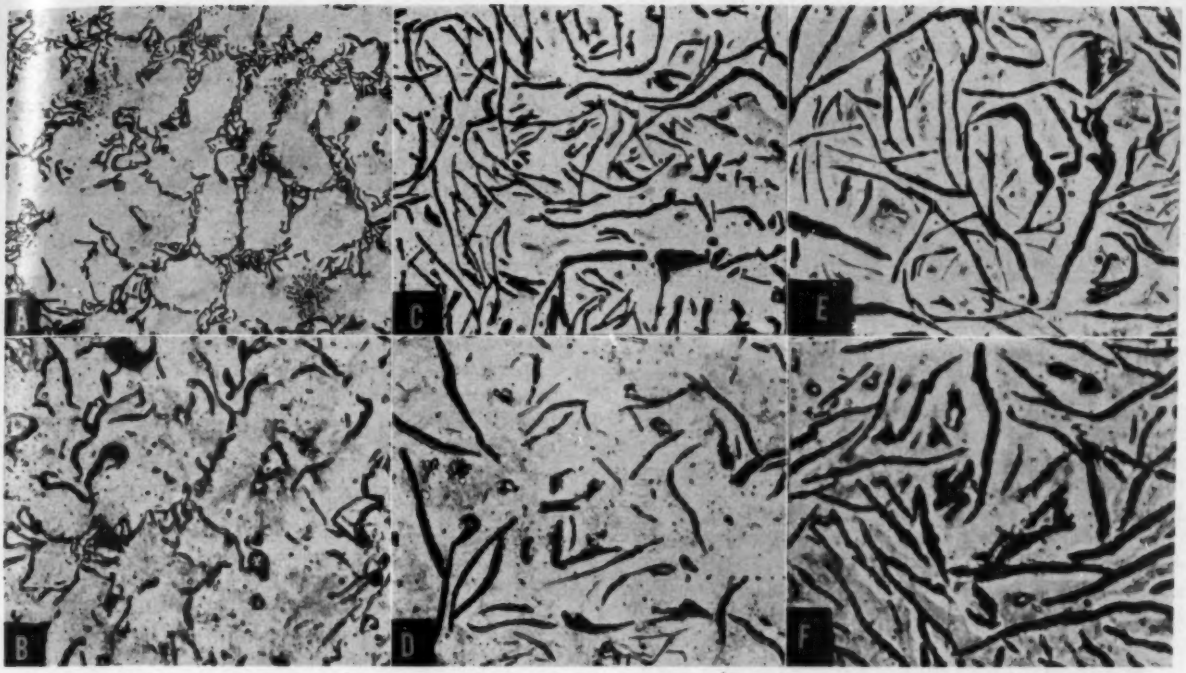



Fig. 8 — Shape and distribution of graphite structure of uninoculated and inoculated irons at various carbon equivalences. *Top row* — uninoculated (A, C, E), *bottom row* — inoculated (B, D, F). *Left to right* — 3.30 (A, B), 3.75 (C, D) and 4.10 (E, F) per cent carbon equivalence. Etched. 100 ×.

of changing the graphitic pattern from Type A to Type D.

The amounts of ferrite and pearlite in the matrix follow generally established patterns. Some ferrite is obtained in the low carbon equivalent iron when Type D graphite flakes are present, as illustrated in Fig. 9A. Ferrite also occurs in the higher carbon equivalent irons surrounding the larger graphite flakes, as shown in Fig. 9C. Inoculation alters the graphitic structure as previously discussed, and increases the amount of pearlite, as demonstrated in Figs. 9B and 9D. While the matrix is predominately pearlite in inoculated irons, the lamellar spacing appears coarser than for the pearlite grains in the uninoculated irons.

The larger amount of pearlite in the inoculated irons in the 3.0 to 4.30 per cent carbon equivalent range results in an increase in hardness, as indicated from the Brinell hardness number of each iron in the table. However, above 4.30 per cent carbon equivalence, the graphitic structure and the matrix of the uninoculated and inoculated irons are essentially the same, and, consequently, the effect of inoculation on hardness in hypereutectic irons is both somewhat erratic and slight in most cases. Pairs No. 15 through 23 in the table show the erratic hardness results. The uninoculated irons at or below 3.30 per cent C.E. are mottled or white in pairs 1 and 2. Accordingly, the uninoculated iron is harder in these cases. Inoculation of these irons tends to reduce or eliminate the extent of mottling and lower hardness.

The results obtained in this work are only applicable to the particular conditions of melting and composition employed. The influence of inoculants on the structure of any iron depends on the existence of prior nuclei in the iron. If these nuclei are present in large amounts, inoculation is not effective; however, if these nuclei are lacking, inoculation is effective. Foundrymen are well aware that the chilling properties of an iron will vary for the same carbon equivalence.

The chilling behavior of an iron is a measure of the existence of nuclei for eutectic graphite solidification. In this work, the authors attempted to standardize the melting conditions, so that the effect of carbon equivalence on the influence of inoculation could be measured. In practice, however, carbon equivalence or composition is only one of the significant variables, and the desirability and influence of inoculation at a given carbon equivalence will depend on other factors as well.

## CONCLUSIONS

The eutectic cell size of uninoculated irons attains a minimum at about 4.0 per cent carbon equivalence or 1.35 per cent eutectic graphite. The cell size becomes larger as the carbon equivalence or eutectic graphite either increases or decreases from the above values.

Effective inoculation with 0.5 per cent silicon, as 85 per cent ferrosilicon containing appreciable aluminum and calcium, greatly decreases the eutectic cell size of low carbon equivalent gray irons. The reduction in cell size from inoculation decreases with increasing carbon equivalence or eutectic graphite and inoculation has only a slight influence on the structure of irons over 4.0 per cent carbon equivalence.

The eutectic cell size of both inoculated and uninoculated irons is finer at the surface than at the center of a 2 in. diameter round for all compositions.

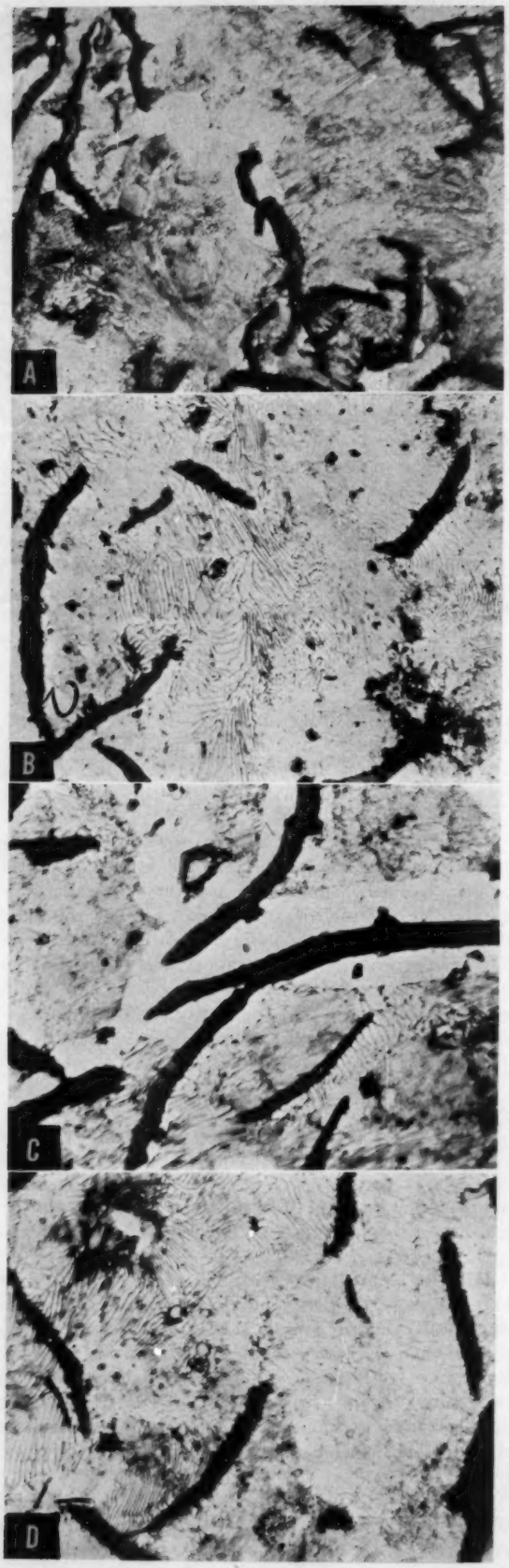



**Fig. 9 — Matrix structure of various uninoculated and inoculated irons. *Top to bottom* — A and C, uninoculated, B and D, inoculated. A and B, low carbon equivalence, C and D, high carbon equivalence. Etched. 500 ×.**

The influence of FeSi inoculation in decreasing the cell size is approximately twice as large at the surface as the center below 4.0 per cent carbon equivalence, and about one and one-half times as large over 4.0 per cent carbon equivalence.

The graphitic pattern of uninoculated irons is altered from Type D to Type A as the carbon equivalence of the iron is increased from 3.30 to 4.30 per cent; raising the carbon equivalence over 4.30 per cent causes the graphite to change from Type A back towards Type D. Fine Type A graphite occurs with large eutectic cells and coarse Type A graphite ensues with fine eutectic cells.

Effective inoculation with FeSi always produces Type A graphite, and apparently, because of its effect on the eutectic solidification temperature, reduces the dendritic pattern of the primary austenite to some extent. The fine eutectic cells of inoculated irons produce a thick, short Type A graphite flake; the larger cells result in a longer, thinner graphitic flake.

Some ferrite occurs in these unalloyed, 2 in. round cast sections around the Type D graphite of the low carbon equivalent irons and Type A graphite of the irons near the eutectic composition. Inoculation of these hypoeutectic irons eliminates all or nearly all of the ferrite, and produces a fairly coarse pearlite and a higher hardness. The structures of the inoculated and uninoculated hypereutectic irons are similar and the effect of inoculation on hardness is small.

The results obtained and conclusions based on these results apply to the melting and inoculation conditions utilized. Changes in melting stock, melting technique and inoculating conditions will undoubtedly alter the extent of the structural variation and compositions at which the finer cell sizes occur.

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# SOLIDIFICATION OF GRAY IRON

Progress Report
Gray Iron Division
Sponsored by
AFS Training & Research Institute

by H. D. Merchant and J. F. Wallace

## ABSTRACT

The work on gray iron solidification described herein was undertaken as a portion of the project on gray iron risering sponsored at Case Institute of Technology by the AFS Training and Research Institute and performed under the direction of the Research Committee of the AFS Gray Iron Division. A previous progress report[1] submitted on this project indicated that more mold wall movement was obtained in green sand molds with inoculated than with uninoculated gray iron. The investigation described here and other work on the influence of inoculation and the eutectic cell size was conducted in an attempt to determine the reasons for the variation in mold wall movement behavior resulting from inoculation. This has been written, along with another report describing the effect of inoculation on eutectic cell size, to present some reasons for the behavior of inoculated and uninoculated iron.

## INTRODUCTION

From a commercial standpoint, gray irons within a range of carbon equivalence of 3.8 to 4.2 per cent are the most important. Accordingly, an iron with a carbon equivalence of about 4 per cent was selected as a typical iron for this investigation. Since the acid-lined cupola is the most commonly employed melting facility, iron from a hot-blast, acid-lined cupola was utilized. The selection of this source of molten iron is significant because of the known influence of melting procedure on the solidification behavior of gray iron.

Iron with a carbon equivalence of 4 per cent starts to solidify as primary austenitic dendrites and completes the solidification as an austenite-graphite eutectic. The addition of effective inoculants exerts a marked influence on the solidification behavior of the austenite-graphite eutectic, as discussed in a number of recent articles on this subject.[2-18] Effective inoculation can be obtained by a number of additions. An 85 per cent silicon grade of ferrosilicon containing appreciable aluminum and calcium was selected for use in this work because it is known to be an effective inoculant in this type of iron.

The investigation involves both thermal and quenching studies during solidification of inoculated and uninoculated gray irons. The thermal studies serve to show thermal gradients and periods of solidification in the castings. The quenching of castings during solidification is more reliable than thermal studies, however, in following the progress of solidification throughout the casting.

## PROCEDURE

The castings utilized in this investigation were a cylindrical bar 2 in. in diameter and 8 in. long, and a short square cross-sectioned bar 4 by 4 by 6 in. long. The longitudinal dimensions of the castings were so chosen that the temperature measurements at mid-length cross-section were not significantly influenced by heat abstraction from the ends. The cylindrical castings were small enough to be cooled rapidly by quenching in water. The cylindrical casting was horizontally cast with a blind riser located at one end and connected to its main body with a riser neck.

The square casting was vertically cast with an open riser positioned directly on the top. Riser and riser neck were dimensioned according to formulas listed in a recent progress report of this gray iron risering work, and were designed to be adequate for the size of casting and type of iron poured.[19] The cylindrical casting was single gated tangentially into the riser, while the square casting was single gated from the bottom of the casting. The gating systems were designed utilizing optimum pouring times, and were dimensioned according to the recommendations incorporated in another recent report.[20]

Four heats of castings were poured for which the iron was melted in a commercial, hot-blast, 54 in. diameter cupola and transferred to a holding ladle. The iron was tapped from the holding ladle into preheated hand operated ladles suspended from a crane and transferred to small 50 lb pouring ladles before pouring into the molds. The cupola charge consisted of various amounts of pig iron, steel and gray iron with small amounts of silicon carbide and ferrosilicon.

The approximate cupola tapping, holding ladle tapping and pouring temperatures were 2800, 2650

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and 2550 F (1538, 1460 and 1404 C), respectively, as determined by an optical pyrometer. Care was taken to control tapping, inoculating and pouring temperatures as closely as possible. Such control of these temperatures is essential because of their observed influence upon the mode of solidification.[4,21] The final chemical composition of the heats is presented in Table 1.

TABLE 1 — CHEMICAL COMPOSITION AND TYPE AND AMOUNT OF INOCULANT

| Heat No. | Casting Size, in. | Composition, %* C | Si | C.E.** | Inoculant |
|---|---|---|---|---|---|
| 1 | 4 x 4 x 6 long | 3.19 | 2.16 | 3.91 | None |
| 2 | 4 x 4 x 6 long | 3.02 | 2.66 | 3.91 | 0.5% Si as FeSi |
| 3 | 2 Dia. x 8 long | 3.21 | 2.22 | 3.95 | None |
| 4 | 2 Dia. x 8 long | 3.12 | 2.69 | 4.02 | 0.5% Si as FeSi |

* All heats contained about 0.085% sulfur, 0.55% phosphorus and 0.80% manganese.
** % Carbon Equivalence = % C + ⅓ %Si.

### *Uninoculated Heats*

Heats 1 and 3 were not inoculated; 0.50 per cent silicon, as 85 per cent grade ferrosilicon containing known quantities of aluminum and calcium, was added to the stream while transferring the melt of heat 2 and 4 into small 50 lb pouring ladles. After pouring into the small ladles, the melt was allowed to drop to the desired temperature before pouring into the molds. The chemical analysis of the ferrosilicon is:

| Si, % | Al, % | Ca, % | Fe, % |
|---|---|---|---|
| 82.54 | 1.96 | 0.92 | Rem. |

Fourteen cylindrical bar molds were made in green sand on a jolt-squeeze molding machine under equivalent conditions. The molds were compacted in flasks and jacketed for pouring. At least a 2 in. layer of sand surrounded all parts of the casting to prevent rapid loss of heat at any location. Similar precautions were taken in making two square bar molds. However, these molds were hand rammed and compacted in slip flasks.

Before jacketing, six Pt-Pt-10 per cent Rh thermocouples were inserted in each square bar mold, along a mid-length cross-section plane of the casting at predetermined distances from the mold surface. The sand utilized for the molds was the regular system sand employed in the production foundry having approximately 4 per cent total moisture, 5 per cent clay bond and 8.5 per cent total combustibles. Four of the six thermocouples inserted in each of the two 4 in. square mold cavities were silica sheathed and enclosed in ⅛-in. OD and 1/16-in. ID quartz protection tubes. These were positioned ½, 1, 1½ and 2 in. into the mold cavity from the vertical mold wall used as a reference plane.

In addition, two unprotected but silica sheathed thermocouples were used to measure interface temperatures on both metal and mold sides. The bead of the thermocuple on the mold side of the interface was ground flat to conform to a hemispherical shape. It was then silica-washed and positioned flush with the mold wall. Two junctions of the thermocouple on the metal side of the interface were not welded into a junction bead, but were left unattached and positioned ¼-in. from the mold interface so that the molten and solid iron would complete the circuit.

Such positioning of thermocouples has been shown to measure interface temperatures on mold and metal sides.[22,23] All six thermocouples were positioned to be perpendicular to the same mold interface. The temperatures in each casting were recorded by connecting the thermocouples to a single point recorder via a potentiometer and a ten-way hand operated switch. This arrangement had the scale range of 1500-3000 F (816-1649 C), and it permitted sequential temperature readings from six thermocouple stations within about 30 sec.

### *Castings Poured*

Seven 2 in. diameter castings were poured from each of the first two heats. After pouring, six castings from each heat were knocked from the molds at predetermined times into a tank of flowing cold water, while the last casting of each heat was allowed to cool in the mold. The shakeout times after pouring for all castings are tabulated in Table 2. Two 4 in. square bar castings were poured from heat no. 3 and 4, allowed to solidify without moving and temperatures recorded for at least 45 min after pouring.

TABLE 2 — SHAKEOUT TIMES AFTER POURING (IN MIN)

| Casting No.* | Heat 1 Uninoculated | Heat 2 Inoculated |
|---|---|---|
| 1 | 3 | 2½ |
| 2 | 4 | 3¼ |
| 3 | 5½ | 5½ |
| 4 | 7 | 7 |
| 5 | 8½ | 8½ |
| 6 | 10 | 10 |
| 7 | 60 | 60 |

* All except casting number 7 in both heats were quenched in water immediately after shakeout.

After cooling to room temperature, all 2-in. diameter cylindrical castings were sand blasted and processed to obtain the desired information on structure and composition. A ½-in. thick transverse section was cut from the longitudinal center of each casting to provide for micro and macro examinations, to assess the progress of solidification at the time of quenching. The quenching procedure does not significantly affect parts of the casting which are solid at the time of quenching, but causes the remaining liquid to solidify in an easily distinguishable fine dispersion.

The photomicrographs in Fig. 1 indicate how the solid and liquid portions at the time of quench can be distinguished. The fine white dispersion in Fig. 1a is white iron that occurred in the portion of the casting that was liquid when quenched. The solidified eutectic cells at the time of quenching appear as black circles. In the castings that were quenched after solidification had progressed substan-

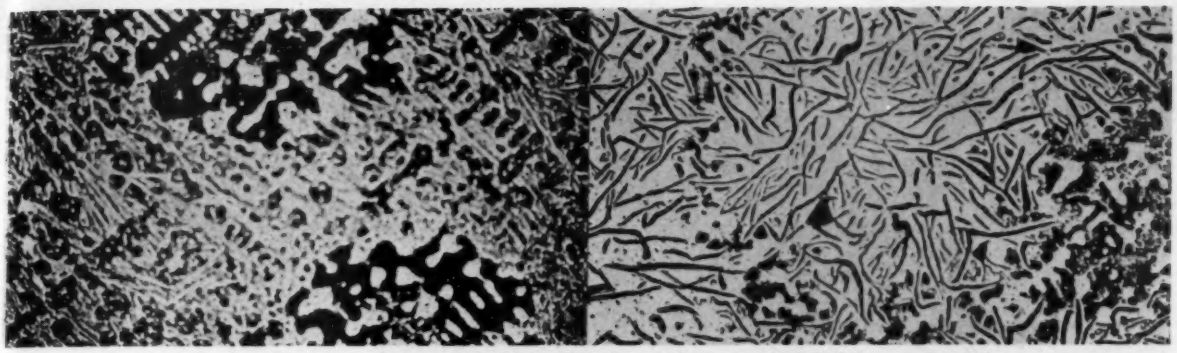


Fig. 1 — Typical microstructures of 2 in. diameter castings quenched during eutectic solidification. *Left* — (a) near outside of casting quenched soon after pouring. *Right* — (b) near center of casting quenched a considerable period after pouring. Picral etch. 100 ×.

tially inside the surface, as shown in Fig. 1b, the effect of the quench had to penetrate through the solidified skin.

Consequently, the remaining liquid in the center of the casting between the eutectic cells was cooled more slowly giving Type D graphite as distinguishable fine dispersions. The solidified portions of this bar at the time of quench appear as Type A graphite. Each speciman was examined after polishing and etching with 6 per cent picral, to determine the per cent solid eutectic.

A sampling plan was devised, to simplify the measurements of amount of solidification, in which the estimates of per cent solidified were made at a given radius on nine equidistant fields. The mean of these nine readings was regarded as an estimate of per cent solidified at the particular radius in question. The radii were so selected that they were 0.2 in. from each other in all solid and all liquid zones, and 0.08 in. from each other in mushy zones. The macrostructures of complete cross-sections of the bars were also studied, to provide further evidence of the progress of solidification in each casting. Typical macrostructures of a pair of uninoculated and inoculated castings quenched during solidification are shown in Fig. 2.

Mid-length sections from the cylindrical bars which solidified completely in the mold were employed to determine eutectic cell size after polishing and etching. From the same piece, a 5/8-in. wide, 1-in. long and 3/8-in. thick radial piece was sectioned, polished and employed to examine the microstructure from the center to the surface of the casting.

## RESULTS AND DISCUSSION

### *Thermal Measurements*

The thermal measurements of 4 x 4 in. bars are presented as cooling curves in Fig. 3 for uninoculated iron, and in Fig. 4 for inoculated iron. The cooling curves exhibit two thermal arrests in cooling rate. The first arrest occurs during the solidification of primary austenite, and the second arrest at the eutectic solidification. The thermal arrests also provide an approximate indication of the temperatures of start of primary and eutectic solidifications. However, these curves do not indicate the end of eutectic arrest as sharply.

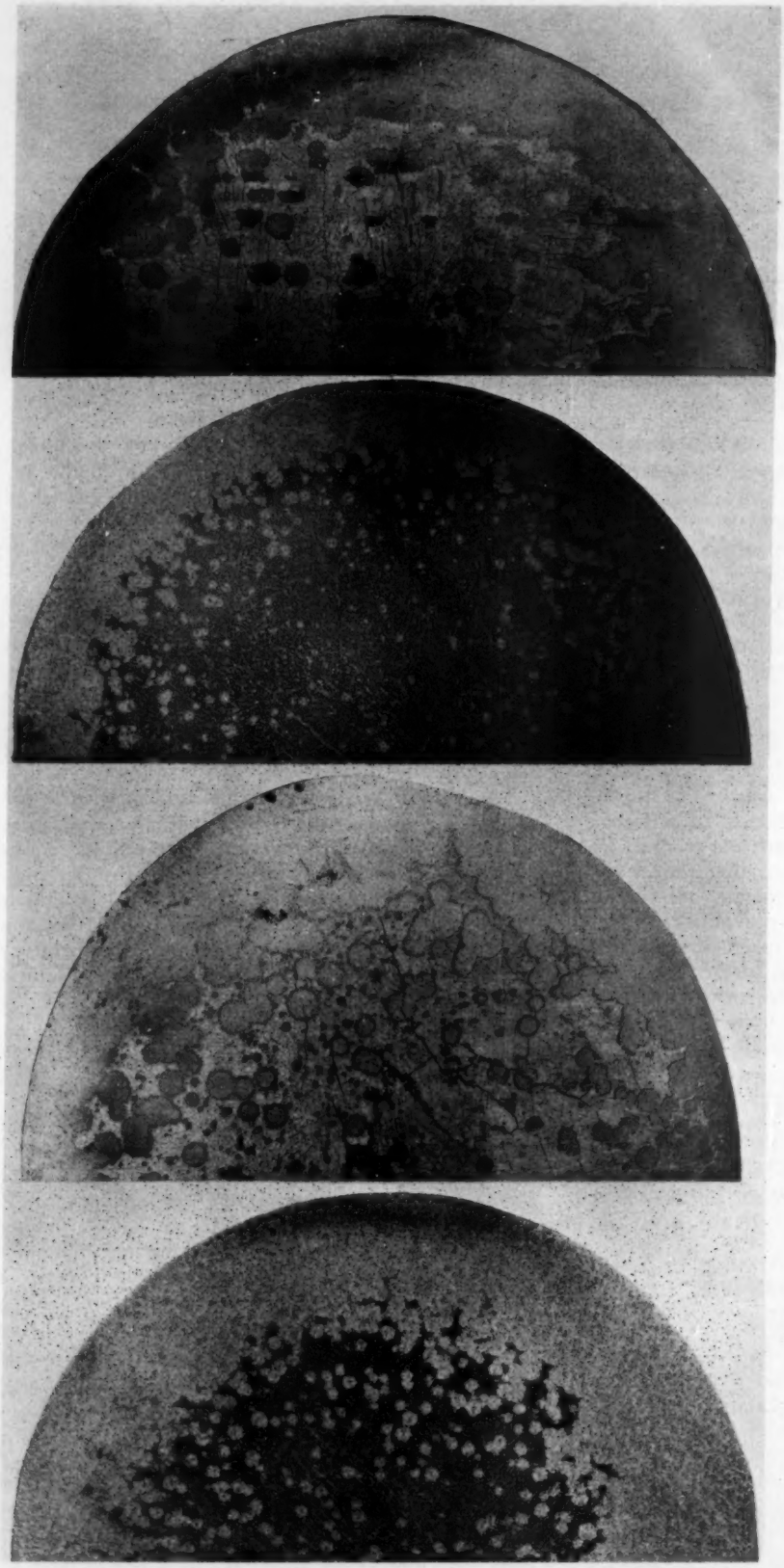
The cooling curves do not show any undercooling or recalescence usually observed, probably because of complex thermal exchanges taking place within the comparatively huge mass of the casting. The temperature of the primary austenite arrest is slightly lowered in inoculated iron due to the alloying effect of the inoculant and small change in the carbon equivalence. The eutectic arrest temperature of the inoculated casting, however, is raised by about 12 F indicating a reduction in undercooling. It is noted that this shifting of primary and eutectic arrest temperatures in inoculated gray iron reduces the temperature range over which the primary austenite solidfies.

The results of these thermal measurements are shown in Figs. 5 and 6, replotted as temperature versus the distance from casting surface at various times for uninoculated and inoculated irons, respectively. The figures provide the thermal gradients within the casting at various times during solidification. A tendency towards shallower temperature gradients is observed during the eutectic solidification in inoculated iron. This influence of inoculation on thermal gradients was expected because inoculation reduces undercooling, and has also been shown to increase the time required for eutectic solidification.[6]

Both of these factors tend to reduce the thermal gradients across the section of the solidifying casting. In addition, it has been observed[24] that the reduction in undercooling due to inoculation depends upon the solidfication rate, so that the higher the freezing rate the greater the reduction in undercooling. These factors also indicate that reduced thermal gradients occur across the section of the casting as a result of inoculation, at least at the start of eutectic solidification, since the freezing rate decreases toward the center of the casting.

### *Cooling Curves*

The cooling curves for the interfaces at the mold and metal casting surface are plotted in Figs. 7 and 8 for uninoculated and inoculated irons, respectively. It is apparent that the difference between mold and metal interface temperatures at various instants is consistently larger for uninoculated iron. This difference between interface temperatures



a) — Uninoculated, quenched 4 min after pouring.

b) — Inoculated, quenched 3¼-min after pouring.

c) — Uninoculated, quenched 5½-min after pouring.

d) — Inoculated, quenched 5½-min after pouring.

Fig. 2 — Typical macrostructures of 2 in. bar casting quenched during solidification. Picral etch. 3.5 ×. Reduced slightly in reproduction.

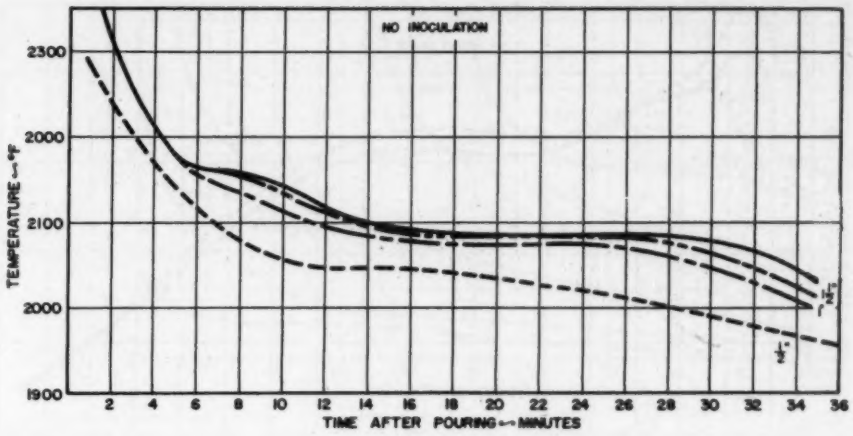



Fig. 3 — Cooling curves of uninoculated 4 in. square casting at various distances from casting surface.

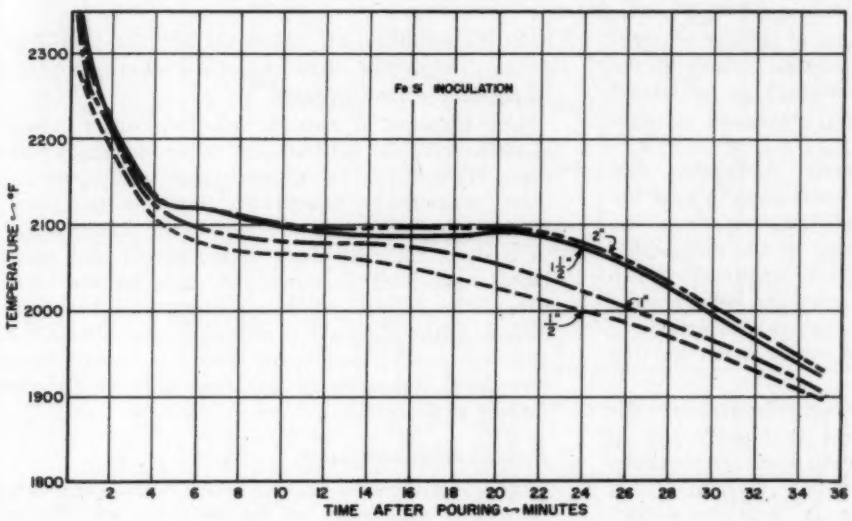



Fig. 4 — Cooling curves of inoculated 4 in. square casting at various distances from casting surface.

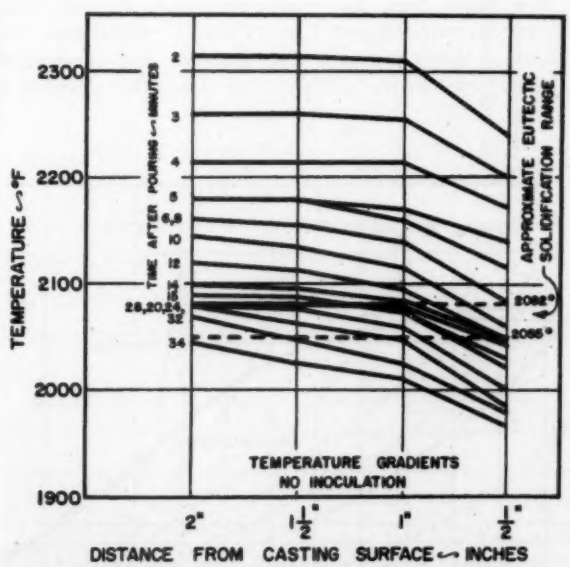



Fig. 5 — Temperature gradients across uninoculated 4 in. square casting during various stages of solidification.

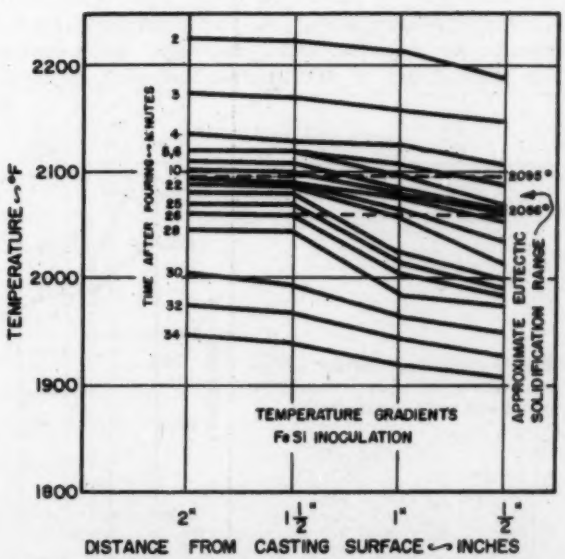



Fig. 6 — Temperature gradients across inoculated 4 in. square casting during various stages of solidification.

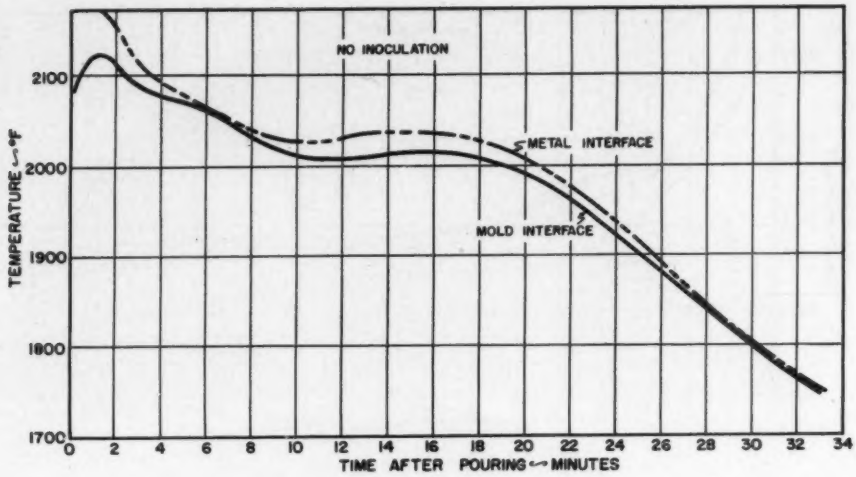



Fig. 7 — Cooling curves of uninoculated 4 in. square casting at metal and mold interfaces.

should indicate the information of an air gap between mold and metal and, therefore, it may be deduced that the air gap is not formed in inoculated iron until the solidification has progressed substantially.

In uninoculated iron, however, it appears that the air gap at the mold-metal interface is formed immediately after the start of solidification. Since the transfer of heat from the casting to the mold takes place primarily by radiation, the existence of an air gap may significantly influence the solidification rate and total solidification time, particularly in a casting where the heat transfer would occur by conducting without such a gap.

The results of the microscopic analysis of the quenched specimens are plotted in Figs. 9 and 10 for uninoculated and inoculated irons, respectively. Each curve represents the percentage of eutectic that has solidified at various distances from the surface of the casting at the time of quenching. These figures indicate the progress of solidification in both uninoculated and inoculated irons. It is evident that a completely solid skin is separated from the all-liquid interior by a fairly wide semi-liquid zone during the early stages of solidification for both types of iron. As solidification progresses, the solid skin becomes thicker and liquid zone is gradually replaced by semi-liquid, mushy zone.

The progress of eutectic solidification is plotted as isoeutectics or solidification sequence diagrams in Figs. 11 and 12, for uninoculated and inoculated irons, respectively. Isoeutectics illustrate that eutectic solidification starts throughout the whole section in a short time for both uninoculated and inoculated irons. This is particularly true for inoculated iron, since eutectic solidfication starts at the center only 1½-min after it has started at the surface.[8] By the time a solid skin ¼-in. thick has formed in the inoculated iron over 25 per cent solid eutectic has formed at the center.

### *Solidification Sequence*

The solidification sequence diagrams reveal some important differences in the mode of solidification between uninoculated and inoculated irons. The eutectic solidification starts about one min earlier in inoculated iron, but ends essentially at the same time as uninoculated iron. Thus, no significant difference exists between uninoculated and inoculated irons as far as the solidification time after pouring is

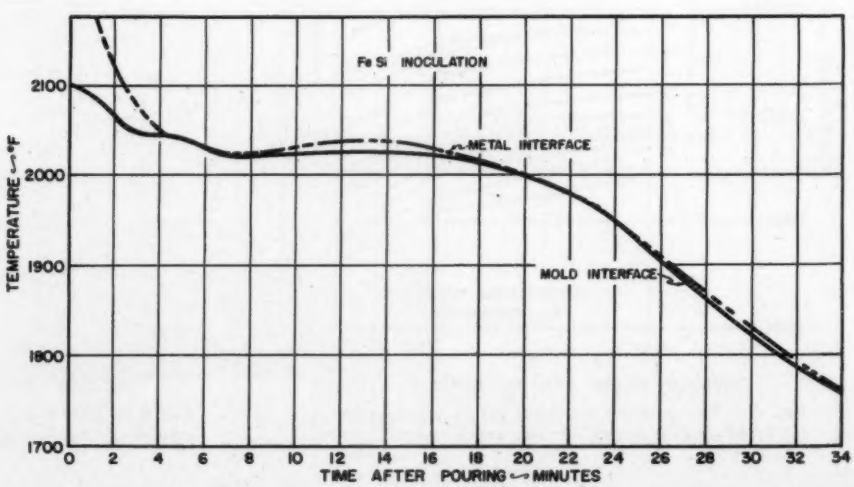



Fig. 8 — Cooling curves for inoculated 4 in. square casting at metal and mold interfaces.

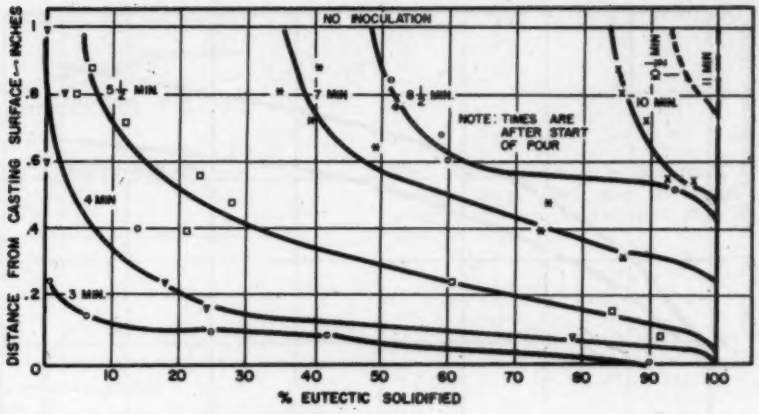



Fig. 9 — Amount of eutectic solidified vs. distance from casting surface at various times after start of pour for uninoculated gray iron.

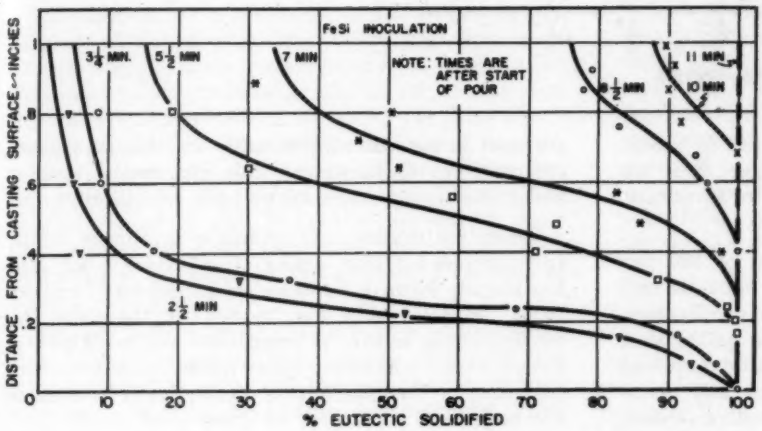



Fig. 10 — Amount of eutectic solidified vs. distance from casting surface at various times after start of pour for inoculated gray iron.

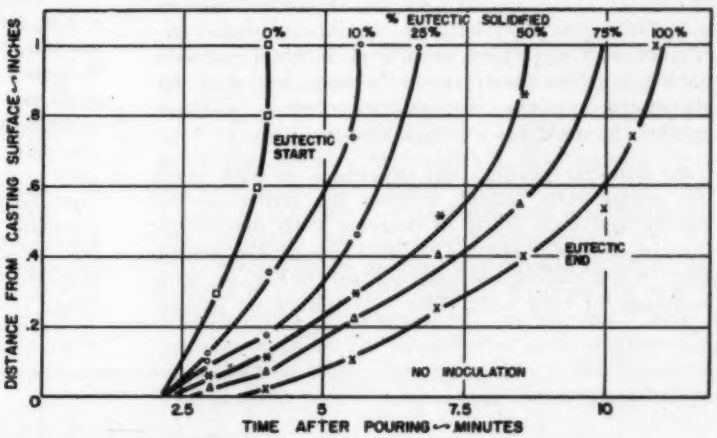



Fig. 11 — I.S.O. — eutectic diagram for uninoculated castings.

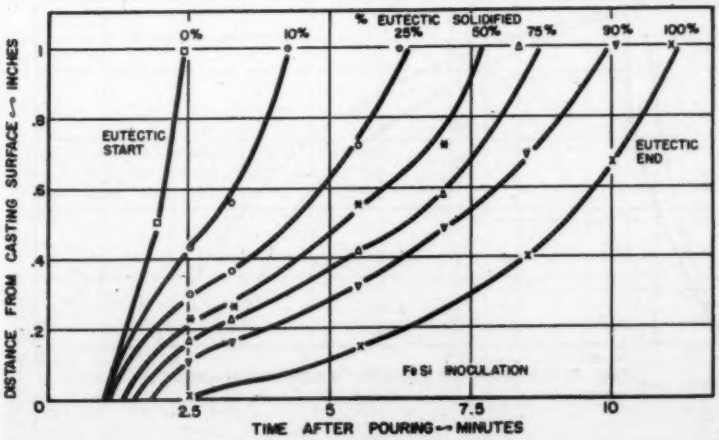



Fig. 12 — I.S.O. — eutectic diagram for inoculated castings.

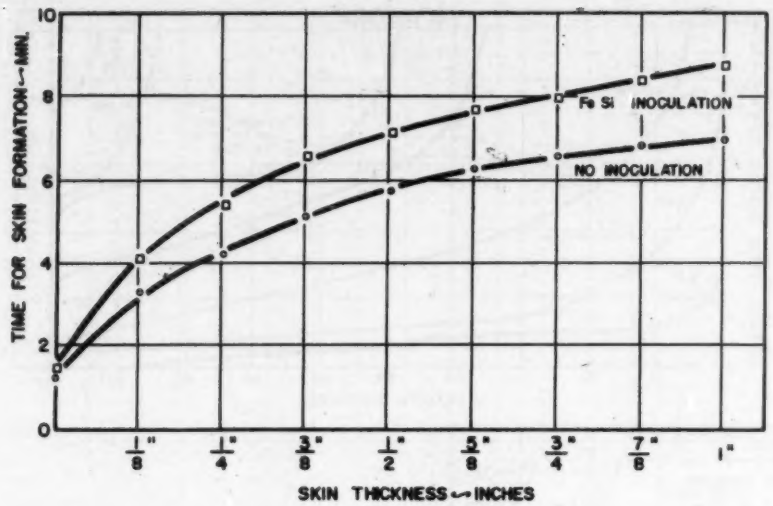



Fig. 13 — Time for formation of various skin thicknesses for uninoculated and inoculated gray iron castings.

concerned. Time for eutectic solidification in inoculated irons, however, is one min longer than for uninoculated irons. These results are in agreement with previous observations.[6]

The time required for the formation of 100 per cent solid skin or the time from the start to end of eutectic solidification at the surface of the casting is slightly higher in inoculated iron. As the skin becomes thicker, this difference in the time required for skin formation between inoculated and uninoculated irons becomes more pronounced, as shown in Fig. 13. This longer time for skin formation in inoculated gray iron has two significant effects. Resistance of the casting surface to internal pressure during graphite precipitation becomes less, and the inoculated casting continues to exert a pressure against the mold for a longer time.

As eutectic solidification progresses in both irons the mushy zone widens, reaches the center of the casting and then starts to decrease with the formation of the solid skin. The thickness of the mushy zone in the casting has been plotted in Fig. 14, with a change in time scale. The zero time scale is now the start or eutectic solidification. In inoculated iron, compared to uninoculated iron, the mushy zone is wider during all stages of eutectic solidification.

When the mushy zone attains a maximum width in uninoculated iron, about $\frac{1}{16}$-in. thick solid skin has already formed; in inoculated iron, on the other hand, no solid skin has formed by the time the whole casting section is completely mushy. This occurrence of a somewhat wider mushy zone and thinner skin in inoculated iron undoubtedly influences the mold wall movement in green sand molds.

No attempt was made to plot solidification sequence diagrams from the cooling curves, because diagrams so derived will merely be a poor approximation of the progress of solidification if it is assumed that a given temperature always represents the same stage of solidification across the whole section of the casting. Such as assumption is clearly untrue, because the solidification starts at a lower temperature at the edge than at the center because of the slower freezing rates at the latter location.[24] On the other hand, the solidification is completed at a lower temperature at the center than at the edge.[15]

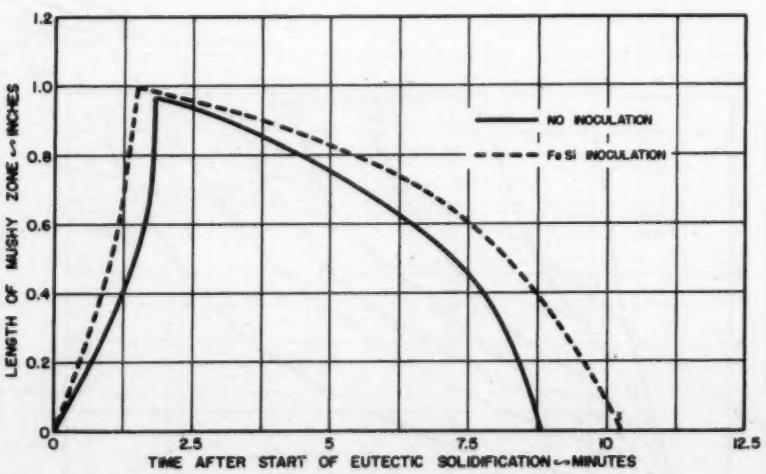



Fig. 14 — Length of mushy zone vs. time after start of eutectic solidification for uninoculated and inoculated irons.

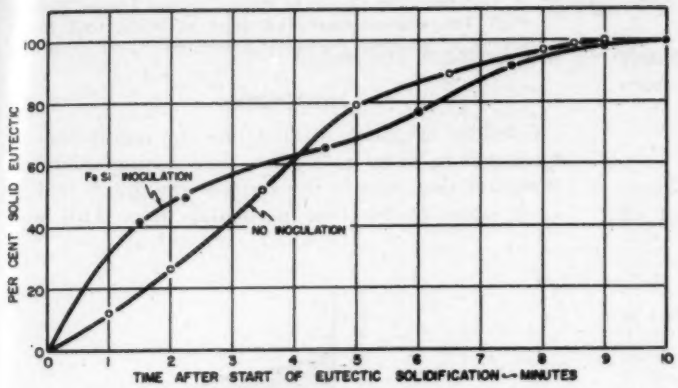



Fig. 15 — Per cent of solid eutectic at various times after start of eutectic solidification for uninoculated and inoculated irons.

In addition to a quantitative indication of the progress of solidification, this quenching experiment also provides a fairly precise estimation of the solidification rate at different depths in the casting during eutectic solidification. By using the formula in the Appendix, the area under each curve in Figs. 9 and 10, indicates the fraction of the casting that was solidified at the time of quenching. These data can be plotted as the per cent of the casting solidified versus the time after start of eutectic solidification for uninoculated and inoculated castings, as illustrated in Fig. 15.

The per cent solid eutectic is higher in inoculated castings than in uninoculated castings during the first stages of solidification, but the situation is reversed after about half of the eutectic has solidified. From this plot, the solidification rates of the casting during eutectic solidification can be computed for both uninoculated and inoculated irons, as shown in Fig. 16. It is apparent that the solidification rate of inoculated iron is initially higher than uninoculated iron, and that it decreases gradually as solidification progresses.

The solidification rate of uninoculated iron, on the other hand, is slow at the outset, gradually increases, attains a maximum and then decreases. These variations in the solidification rates of uninoculated and inoculated irons indicate changing rates of heat abstraction by the mold, and provide circumstantial evidence of the formation of air gap during the first stages of eutectic solidification of uninoculated castings.

The earlier formation of the solid skin, and less mushy type of solidification, that occurs in uninoculated irons assist in explaining the reduced mold wall movement in these irons compared to inoculated irons. The solid skin is better able to resist the metallostatic liquid force and graphitic expansion pressure during eutectic solidification. The less mushy type of solidification causes less entrapment of solidifying liquid, and permits better metal flow within the solidifying casting.

## CONCLUSIONS

The eutectic solidification temperature range occurs at higher temperatures in inoculated compared to uninoculated 4.0 per cent carbon equivalent gray iron. The thermal gradients across the casting section are slightly shallower during the eutectic solidification of an inoculated compared to uninoculated gray iron.

Inoculation of gray iron delays the formation of an air gap at the mold-metal interface during solidification. Associated with this delay in gap formation is the small temperature difference between mold and metal at the interface, high eutectic freezing rate across the casting section during the early stages of solidification and delay in the formation of a complete solid skin in the casting.

The inoculation of gray iron results in an earlier start of eutectic solidification, a faster initial rate of solidification, a subsequently slower rate and a greater depth of the liquid-solid zone compared to uninoculated iron. The time during which eutectic solidification occurs is somewhat longer than for un-

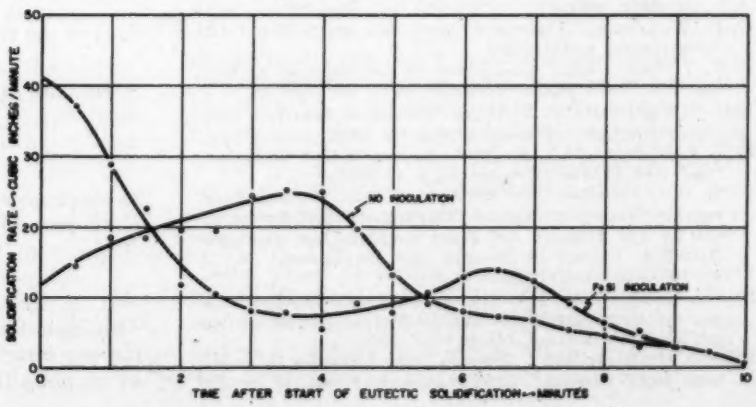



Fig. 16 — Solidification rates of uninoculated and inoculated irons at various times after start of eutectic solidification.

inoculated irons, but the total solidification times are similar for both irons.

These factors assist in explaining the greater mold wall movement that occurs in inoculated irons poured in green sand molds.

## ACKNOWLEDGMENT

The assistance of Superior Foundry, Inc., Cleveland, in conducting the melting and pouring for all castings is gratefully acknowledged.

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## APPENDIX

Consider a casting with a one in. radius that is quenched in water at a given time during solidification, and that data in the form of the graph in Fig. 17 is obtained for that particular time. This plot

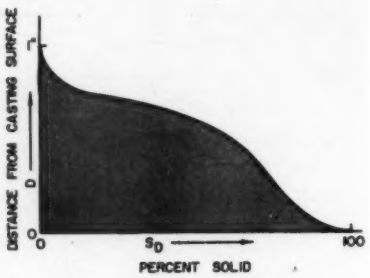



**Fig. 17 — Solidification from center to surface of casting at a given time (schematic).**

indicates the progress of solidification from the surface to center of the casting, and shows the per cent of the casting solidified at various distances from the surface of the casting. It is desired to estimate per cent of the whole casting solidified from this plot.

Assume a small circular strip dD in. thick at a distance $D$ in. from the casting surface, as shown in Fig. 18:

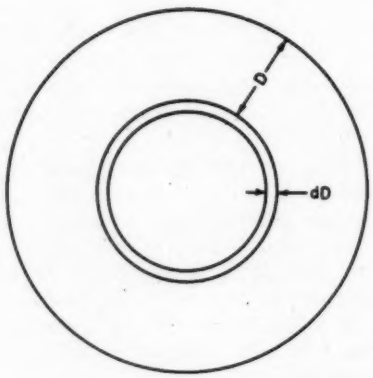



**Fig. 18 — Transverse section of a cylindrical casting with one in. radius (schematic).**

then,

$$\text{Area of the strip} = \pi(1-D)^2 - \pi[1-(D+\Delta D)]^2$$
$$= 2\pi\Delta D(1-D)$$

$\therefore$ Amount of solid in the strip
$$= (\%\ \text{solid in the strip}) \times (\text{area of the strip})$$
$$= S_D \cdot 2\pi \cdot \Delta D(1-D)$$

$\therefore$ Amount solid in the whole casting at the instant of quenching is given by:

$$\%\ \text{Solid} = 2\pi \int_0^1 (1-D)\, S_D \cdot dD$$

This equation may be used with Fig. 17 to calculate the per cent of the whole casting that has solidified by graphical methods.

1961

**Charles Edgar Hoyt Memorial Lecture**

# CAST METALS AND CAST SHAPES

by J. B. Caine


## ABSTRACT

**Evidence is presented that slight changes in shape and size by design can significantly increase load carrying ability by decreasing applied stress. The possible decrease in applied stress by design overshadows variations in strength of metals to such a degree as to make the classic mechanical properties of the metal of minor significance. Casting is the most efficient method of producing such shapes.**


## CASTING DESIGN

There are many aspects to casting design. Perhaps the one least appreciated, and perhaps the most important, has to do with the competitive position of the foundry industry. The industry must face one unpleasant fact. Many engineers, designers and users of castings are afraid of castings. The designer and engineer have been taught that castings make second grade parts. They have been taught that metal must be wrought for maximum physical properties. This teaching is not incorrect if based solely on classical tensile properties. The error involves improper evaluation of the significance of these properties. Improper evaluation has been compounded by the apathy of the foundry industry plus the human fear of the unknown. This combination has led to even ridiculous situations.

In one instance the engineers of one of the largest oil companies were so afraid of castings that they laid a pipe several hundred feet long, and sand bagged a cast steel valve they were testing to 600 psi water pressure. They were horrified when the writer stood on the valve during the test. Even if the valve had leaked, at worst the writer would have gotten his shoes wet. In no case would the casting have exploded.

Diametrically opposed to the experience of the previous paragraph, we have the belief of some designers that, in many instances, the strongest parts that can be made of a given metal at a given hardness are castings. This paper is an effort to answer these diametrically opposed viewpoints.

The proponents of castings in design have no supercast metal. In fact, they are willing to accept slightly lower classic mechanical properties in castings. The advantage of castings is derived from the ability of liquid metal to assume any shape, shapes that cannot be formed efficiently by any other forming process. This flexibility enables the design of shapes as castings that will uniformly distribute the load so that every ounce of metal is carrying its share. No part of the casting is overloaded, none underloaded. In short, stress concentration is at a minimum.

## DESIGN FUNCTION

True, a study of applied stress is a function of design and seems alien to the foundryman. It behooves the foundryman to become familiar with at least one phase of design, that of junctions of sections and connecting members. Proper design of junctions and connecting members offers many advantages. Unfortunately, it seems as if this phase of design must be sold to the designer. Someone must do the selling. If streamlined cast design is accepted by the designers it opens up a vast field for castings. The opportunity to cast parts that have never been cast before. What is needed is effort on the part of the foundryman, for he alone has the incentive—more business. To be effective this effort must be backed up with knowledge. The following discussion may be considered an introduction to the basic concepts of casting design.

One phase of casting design can be summarized by the practical example of Fig. 1. The lever of Fig. 1a has been made as an alloy steel forging, specified to a yield strength of 100,000 psi yield strength. The inefficiency of the oval section of Fig. 1a as a

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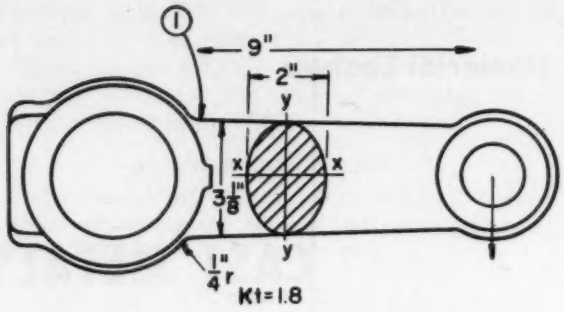



Fig. 1a — Oval section, area — 4.95 sq in. $I_x$ — 2.9 in.$^4$, $I_y$ — 1.2 in.$^4$ $S_x$ — 1.9 in.$^3$, $S_y$ — 1.2 in.$^3$ (1) A load of 21,000 lb about the *X* axis, and 13,400 lb about the *Y* axis, results in modulii of rupture of 100,000 psi in this area.

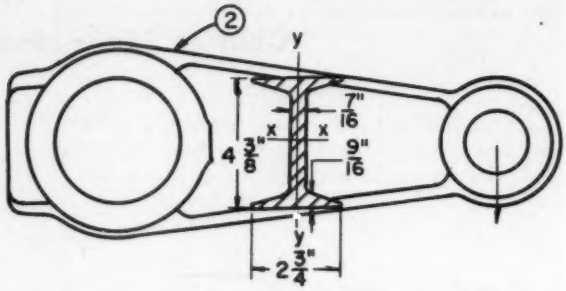



Fig. 1b — I-beam section, area — 4.55 sq in. $I_x$ — 12.1 in.$^4$, $I_y$ — 1.9 in.$^4$ $S_x$ — 5.5 in.$^3$, $S_y$ — 1.4 in.$^3$ (2) A load of 21,000 lb about the *X* axis results in a modulus of rupture of 34,000 psi. A load of 13,400 lb about the *Y* axis results in a modulus of rupture of 87,000 psi in this area.

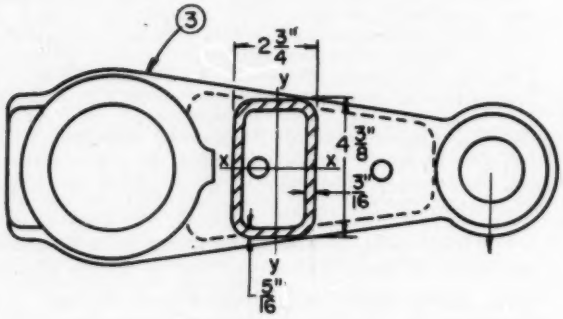



Fig. 1c — Cast tubular section, area — 4.6 sq in. $I_x$ — 11.2 in.$^4$, $I_y$ — 4.6 in.$^4$ $S_x$ — 5.1 in.$^3$, $S_y$ — 3.4 in.$^3$ (3) A load of 21,000 lb about the *X* axis results in a modulus of rupture of 37,000 psi. A load of 13,400 lb about the *Y* axis results in a modulus of rupture of 45,000 psi in this area.

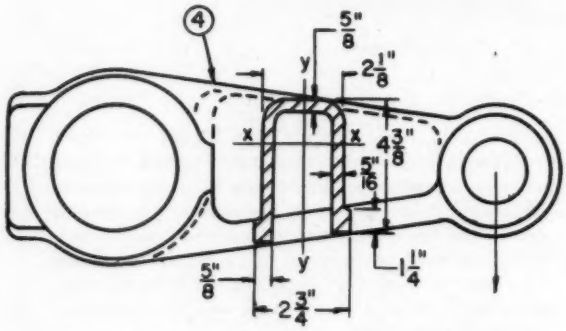



Fig. 1d — Cast *U* section, area — 4.6 sq in. $I_x$ — 9.1 in.$^4$, $I_y$ — 6.2 in.$^4$ $S_x$ — 4.2 in.$^3$, $S_y$ 5.0 in.$^3$ (4) A load of 21,000 lb about the *X* axis results in a modulus of rupture of 44,000 psi. A load of 13,400 lb about the *Y* axis results in a modulus of rupture of 27,000 psi in this area.

**Fig. 1 — Relationship between shape and imposed stress. *t* equals section thickness.**

load carrying member is self evident. Most of the metal is near the centerline, and is not carrying nearly its share of the load.

The connecting member of the lever of Fig. 1 can be either cast, or forged as an I-beam section, like Fig. 1b. The better metal distribution has decreased the imposed stress, as measured by modulus of rupture, for a given load in the direction of the arrow to a third of that of the oval section. The example shows that the modulus of rupture is decreased from 100,000 psi to 34,000 psi. This decrease in imposed load is exactly the same, as far as load carrying ability is concerned, as increasing the strength of the metal three times. If the lever is cast, this increase in load carrying ability can be obtained at practically no increase in cost. The more complex dies required to form the I-beam section can make the shape of Fig. 1b quite costly as a forging, especially if the quantities are small.

The imposed stresses just discussed, as well as those to be discussed later, are determined by the moment of inertia (I) and section modulus (S) about the *X* and *Y* axes ($I_x$, $I_y$, $S_x$, $S_y$) as given in Fig. 1.[1,2]

In many instances a part is subjected to loads in more than one direction and to twisting. The I-beam section of Fig. 1b, as well as most wrought sections, do not have the ability to resist loads in all directions. In the example of Figs. 1a and 1b the imposed stress caused by a load in the direction of the arrow (about the *X* axis) is decreased from 100,000 to 34,000 psi. The stress caused by a transverse load at right angles to the arrow and plane of the paper (about the *Y* axis) is only decreased from 100,000 to 87,000 psi. Therefore, the I-beam section is satisfactory if the load is only in one direction. It is unsatisfactory if there is major complex loading.

## CASTING FLEXIBILITY

The flexibility of casting allows other shapes that decrease imposed stress when loaded in all directions. The tubular section of Fig. 1c is one such shape. All stresses as measured by modulus of rupture are now less than 45 per cent those of the oval section

of Fig. 1a. Two outstanding advantages of cast design are self evident.

1. The load carrying ability of the part when made of the same metal, at the same hardness and strength is more than doubled. A cheaper, more machinable, more foolproof metal can be used. In the example of Fig. 1, carbon steel, or perhaps pearlitic malleable, or ductile iron can be substituted for the liquid quenched and tempered alloy steel now used. The oval section of Fig. 1a requires a metal of 100,000 psi yield strength. The tubular section requires a metal with 37,000 to 45,000 psi yield strength under static loading depending on just how the load is applied.
2. The possible savings in alloy, heat treatment and especially machining is in excess of 20 per cent of the cost of the rough part.

The cost of the core required to make the tubular section must be charged against the above savings. If this core cost is excessive, or the core is objectionable for any other reason, the flexibility of casting allows many compromises. One compromise is the U-section of Fig. 1d. No core is required. The inspection problems of hollow sections are overcome. It is never possible to obtain quite as uniform resistance to complex loading with an open sided section as can be obtained with a tubular section, but casting can come close.

The imposed stresses with the same loads of 21,000 and 13,400 lb are now 44,000 and 27,000 psi. The load carrying ability of the U-section of Fig. 1d is still over double that of the oval section. The substitution of the cheaper, more machinable metals, as discussed previously, is still possible.

It cannot be emphasized too strongly that the variations in imposed stress just discussed have the same influence on load carrying ability in service as equal variations in the strength of the metal. Variations due to shape in Fig. 1 of over 100 per cent grossly over shadow commonly accepted variations in strengths of metals to the point of making discussions about a few thousand psi in yield and tensile strength mere quibbling.

## DUCTILITY

A similar situation exists regarding so called ductility. If the 8 in. tension face of the connecting members of Fig. 1 elongates uniformly due to overloading even 4 per cent the lever as a whole will bend over 2 inches. Even a small fraction of such deformation will render any mechanism, or structure, inoperative due to misalignment. Under these conditions any discussion of tensile elongation over just one, or 2 per cent has no more meaning than a discussion of how many angels can dance on a head of a pin.

Tensile elongation only has meaning under conditions of stress concentration. If all deformation is concentrated in a small localized area, as is possible in the ¼-in. radius fillet of Fig. 1a, appreciable local deformation is necessary. But, the cast designs of Figs. 1c and 1d have no radius of any size. The connecting member blends into the hub at the point of maximum stress. It can be said without exaggeration that design by the elimination of points of stress concentration, has made discussion of differences in tensile ductility above just a few per cent as meaningless as quibbling over a few thousand psi tensile and yield strength.

For that matter the whole preceeding discussion has little meaning in any type of operating mechanism. The lever of Fig. 1 moves and is made of a ductile metal. Under these conditions, failure for most any reason, will be in fatigue. Now, even the old cry for maximum tensile properties for "insurance" has no meaning. Design and evaluation based on classic tensile properties is now dangerous. It has been and still is the cause of serious service failures.

The values of modulus of rupture in Fig. 1 are, in the metallurgist's language, maximum fiber stress. If a part is statically loaded transversely so that the modulus of rupture equals the tensile strength, it will not break. For that matter, it will show little permanent deformation. Stress decreases rapidly beneath the surface, as a straight line function, to zero at the neutral axis. Therefore, the highly stressed metal at the surface is supported by that beneath the surface. The average stress in parts similar to the lever of Fig. 1 is only half the modulus of rupture, or maximum fiber stress.

It cannot be emphasized too strongly that the principles of the preceeding paragraph do not apply in fatigue loading. The metal beneath the surface no longer supports the surface. If the fatigue strength of the surface metal is exceeded, a crack will start. This sharp crack causes stress concentration to such a degree that it overshadows the lower stress level beneath the surface and the crack progresses to failure.

## FATIGUE FAILURE

The lever of Fig. 1a will surely withstand the loads given in Fig. 1, if they are applied statically. If the same loads are applied intermittently so that fatigue becomes operative, the lever will surely fail. Failure is a matter of load applications and time. Fatigue failure is accentuated by stress concentration imposed by the ¼-in. radius fillet of Fig. 1a. This fillet at the highly stressed junction of sections of Fig. 1a imposes a stress concentration factor of about 1.8.[3,4] This means that the maximum fiber stress is increased from 100,000 to 180,000 psi at this point. This maximum fiber stress is much above the fatigue strength of even the alloy steel.

The streamlined junctions of Figs. 1c and 1d impose little, or no, stress concentration as the connecting members are blended into the hub, or boss. Stress concentration approaches 1.0. Therefore, the 37,000 to 44,000 psi maximum fiber stresses of Figs. 1c and 1d are not increased. These imposed stresses are above the fatigue strength for infinite life of carbon steel castings with cast surfaces, but to a much lesser degree than for the alloy steel forging of

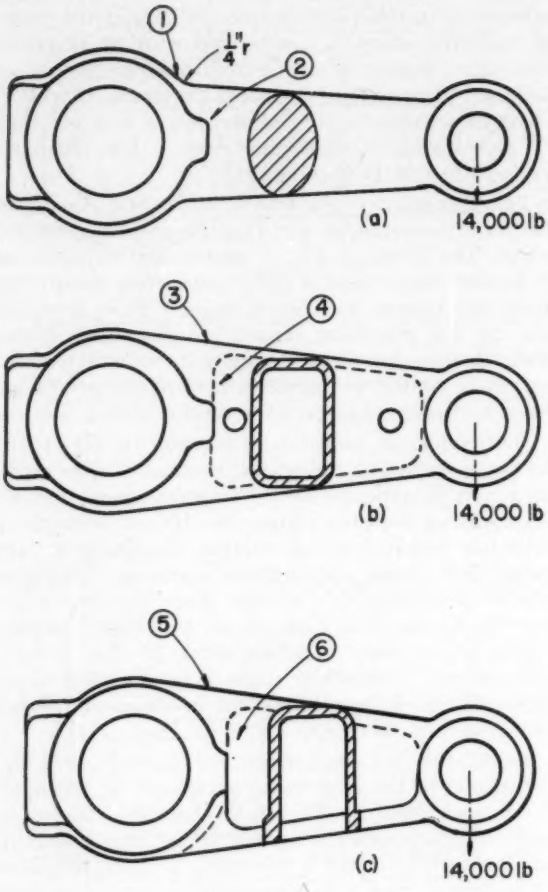



Fig. 2 — Relationship between shape, imposed stress, yield and fatigue strength. *(a)* — (1) Maximum fiber stress (modulus of rupture) due to 14,000 lb load — 66,000 psi. Stress concentration factor 1.8 = 118,000 psi. Fatigue strength of metal — 38,000 psi. (2) Average stress due to 14,000 lb load — 33,000 psi. Yield strength of metal 100,000 psi. *(b)* — (3) Maximum fiber stress ((modulus of rupture) due to 14,000 lb load — 25,000 psi. No stress concentration. Fatigue strength of metal — 26,000 psi. (4) Average stress due to 14,000 lb load — 15,000 psi. Yield strength of metal 45,000 psi. *(c)* — (5) Maximum fiber stress (modulus of rupture) due to 14,000 lb load — 29,000 psi. No stress concentration. Fatigue strength of metal — 26,000 psi. (6) Average stress due to 14,000 lb load — 14,500 psi. Yield strength of metal — 45,000 psi.

Fig. 1a. Instead of a difference in imposed stress in the order of 2, as discussed previously for static loading, the difference due to design is in the order of 4 to 5 times for fatigue loading. Again, this difference in imposed stress is the same as an increase in the strength of the metal.

It is interesting to compare the actual loads and stresses that would be applied by design using a conservative factor of safety of 3 based on yield strength. The maximum stress allowable for the 100,000 psi yield strength of steel of Fig. 2a would be 33,000 psi. A load in the direction of the arrow in Fig. 2 of 14,000 lb would impose an average stress of 33,000 psi in the junction of the connecting member and large boss. This would be considered conservative design.

Consider now what happens under fatigue loading. The maximum fiber stress at the surface is now the controlling stress, not the average stress. The 14,000 lb load imposes a maximum fiber stress of 66,000 psi. Stress concentration at the ¼-in. radius fillet may increase this stress by a factor of 1.8 to 118,000 psi. The fatigue strength of forged steel with a forged, unmachined surface is about 30 per cent of the tensile strength of 125,000 psi, or only 38,000 psi. The fatigue strength is exceeded by a factor of 3. The conservative static design of the preceeding paragraph is now grossly underdesigned for maximum service. The lever will fail after a few hundred thousand load applications of 14,000 lb.

## MINIMUM STRESS CONCENTRATION

Consider now the streamlined cast designs of Figs. 2b and 2c when stress concentration is at a minimum. The same static load of 14,000 lb will impose an average stress of 15,000 psi in the tubular section of Fig. 2b. This is well below a yield strength of 45,000 psi for carbon steel, pearlitic malleable and ductile iron and the factor of safety of 3 still holds. The maximum fiber stress in fatigue is 25,000 psi, and this value need not be increased because of stress concentration. A normalized carbon steel can be easily produced meeting 80,000 psi tensile strength, 45,000 psi yield strength.

The fatigue strength of cast steel with a cast surface is about 33 per cent of the tensile strength,[5] or 26,000 psi. The maximum fiber stress of the streamlined cast design of Fig. 2b is below the fatigue strength. The same stress of the lever of Fig. 2a is 3 times the fatigue strength with the same load. The cast lever of Fig. 2b will perform indefinitely under fatigue loading, whereas the forged lever will eventually fail. Failure will depend on how often the maximum load is applied during its service life. The preceding is true even though the cast lever is made of a cheaper and theoretically weaker metal.

The same difference in applied stress, the same increase in true service strength, or service life, is shown by the cast lever of Fig. 2c. The levers of Figs. 2b and 2c are almost identical as far as load carrying

ability is concerned. Their chief differences have to do with possible economy, ease of production and inspection.

A good question at this time is if the standard design of Fig. 2a is so grossly underdesigned in fatigue, why are there not more fatigue failures. There are a number of good answers to this question. Unfortunately, all are hard to pin down to exact data. The first answer must be that there are a lot of fatigue failures, more than most people realize. Most occur after months, or years of service, and no one knows about them except those immediately concerned.

Complaints are registered with a second group. The complaints go through channels until it all finally boils down to whose fault it is. Almost invariably a tension test bar is taken from the offending member. If the tensile properties do not dot every *i* and cross every *t* of the specification the metal is at fault. Every one is satisfied—except the metallurgist. It seems to make no difference if the failure exhibits every indication of failure except failure in tension. Everyone is still satisfied—except the metallurgist.

## STRESS VS. LOAD

However, the vast majority of parts do not fail. Although not proved, it is the writer's opinion that only a few operating mechanisms are operated at their full output for more than a small fraction of their useful life. The relation between stress and number of load applications for fatigue failure is logarithmic. Although the lever of Fig. 2a will withstand a load of 14,000 lb for only several hundred thousand applications, it will withstand a load of 10,000 lb indefinitely. The chances are it will be loaded most of its useful life at 10,000 lb, or less. The 14,000 lb load will only be applied occasionally.

This state of affairs does not mean that decreasing imposed stress by cast design is not significant. It is significant in either allowing higher loads with no increase in cost, or weight, or to decrease cost with the same load.

Admitted, Figs. 1 and 2 may be extreme examples. It is of interest to go to the other extreme and see what just a small change of ¼-in. in one dimension of a load carrying member will do. Figure 3 shows three cast sections, differing only by an increase of ¼-in. in the *H* dimension. Such a minor change can usually be made in any part without exceeding the overall dimensions, or influencing its performance. Such a change is simple with castings. Perhaps the reason such changes are not considered by the designer is that standardization mandatory with other forming processes makes such small changes impractical, or impossible.

The ¼-in. increase of the *H* dimension of the three cast sections of Fig. 3 decreases the major imposed transverse stress in bending 15 to 39 per cent. This causes a decrease in imposed stress at a 45,000 psi yield strength of carbon steel, ductile and malleable iron of 6800 to 17,600 psi.

Consider now the relation of this reduction in applied stress to its twin, tensile and yield strengths.

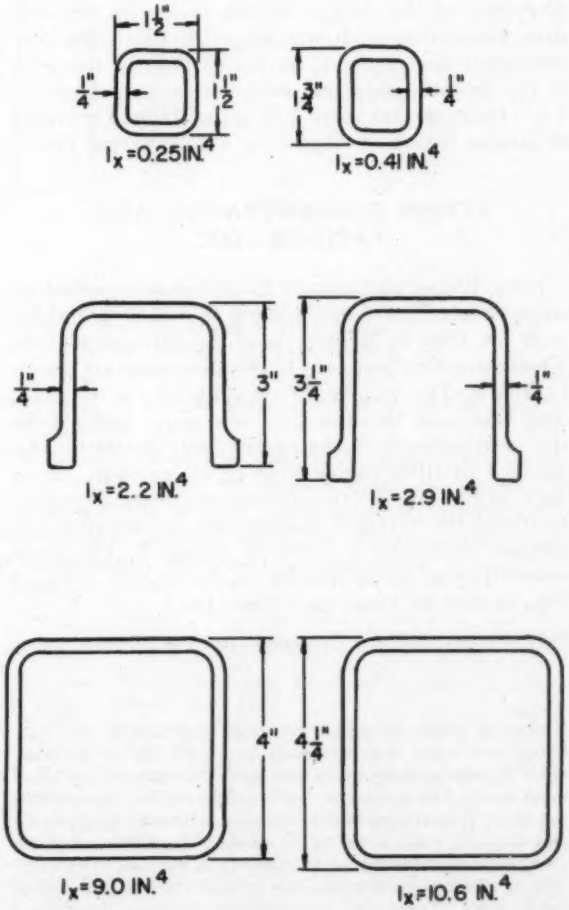



Fig. 3 — Relationship between size and imposed stress. *Top* — Imposed transverse stress of any magnitude decreased 39 per cent. *Left* — r = 0.48 in. *Right* — r = 0.60 in. Imposed compressive stress (buckling) of any magnitude decreased 20 per cent. *Center* — Imposed stress of any magnitude decreased 24 per cent. *Left* — r = 0.92 in. *Right* — r = 1.02 in. Imposed compressive stress (buckling) of any magnitude decreased 10 per cent. *Bottom* — Imposed transverse stress of any magnitude decreased 15 per cent. *Left* — r = 1.6 in. *Right* — r = 1.7 in. Imposed compressive stress (buckling) of any magnitude decreased 6 per cent.

A variation of 5000 psi in tensile and yield strengths is considered a major change—major to the point of warranting a separate grade in many metal specifications. A design change of less than ¼-in. is therefore equal to a separate grade in metal specification.

## COMPRESSIVE LOADING

Compression strength of the members of Fig. 3 is increased from 6 to 20 per cent by the increase in resistance to buckling under compressive loading. This increase is in the same order of magnitude as the increase in transverse strength in bending just discussed.

Torsional strength (resistance to torsional loads) is increased, but to a lesser degree than for transverse and compressive loading. A decrease in torsional applied stress of 3 to 17 per cent is due to the ¼-in. increase of Fig. 3. However, this is still a major increase in service strength when translated into lb/sq in.

Perhaps the difference of 2 to 5 times in load carrying ability of the levers of Figs. 1 and 2 may seem extreme. Recent information indicates that difference in the order of 5 in resistance to fatigue loading due to slight changes in shape may be too conservative. The crux of the fatigue discussion has to do with stress concentration. It was estimated that stress concentration due to the ¼-in. radius fillet of the lever of Fig. 2a multiplied the overall stress by a factor of 1.8. The important point is if actual fatigue strength, or fatigue life, is decreased by a comparable factor.

## STRESS CONCENTRATION AND FATIGUE LIFE

Nara, Wright and Briggs[6] have recently reported on stress concentration and fatigue life of *L* junctions, with the fillet in tension, as is the fillet of Fig. 2a. Those junctions pertinent to this discussion are shown in Fig. 4. The results are startling to say the least. The load used in these tests was quite high, above the yield strength of the steels, about 50,000 psi. An increase in fillet radius from ¼-in. to ⅝-in., as in Figs. 4a and 4b (all other factors remaining constant), increased the average* fatigue life, the resistance to fatigue failure, by a factor of almost 7. The stress concentration factor of the ¼-in. radius fillet of Fig. 2a may be closer to 7 than 1.8.

If a slight change in design is incorporated along with the larger fillet the increase in average fatigue life is almost unbelieveable, in the order of 22 times. The reason for this second and spectacular increase seems to be due to a slight change in stress pattern. The single point of high stress of Figs. 4a and 4b changes to two separate points when the corner is reinforced as in Fig. 4c.

However, simply changing the number of points of high stress does not automatically increase fatigue life, as shown by the design of Fig. 4d. The difference in fatigue life between Figs. 4c and 4d is not due to the forming processes, welding and casting. Nara, Wright and Briggs machined the fillets of the corners of Figs. 4c and 4d to the same radius. Fatigue lives were then essentially the same. The great difference in fatigue life must be due to slight differences in shape. The points of high stress of Fig. 4d coincide with rather abrupt changes in section that are themselves stress raisers.

The data of Fig. 4 and the reference[5] must be compared with accepted handbook data, that cast steel shows 10 to 20 per cent lower fatigue strength than wrought steel when the wrought steel is tested in the direction of hot work. The handbook data are based on small, polished specimens whose shape and surface are never approached in actual working parts. Which set of data is the most significant? The reader must judge for himself.

## LOAD CARRYING ABILITY

Figure 4 again emphasizes the theme of this discussion. What are now considered minor changes in shape and size have great, even spectacular influences on actual load carrying ability. These seemingly minor changes in shape and size are influential to the extent of making classic mechanical properties of little significance. The advantage of castings is their ability to form the proper shapes efficiently. For example, how can the corner of Fig. 4c be formed efficiently other than by casting.

The importance of Nara, Wright and Briggs's work cannot be overemphasized. The information is basic. *L* junctions are incorporated in practically all designs. The data of Fig. 4 should therefore be applicable to all design. The only question is if the fillet radius is in tension. Even more basic is the concept of reinforcing points of high stress to disperse stress concentration. This concept is applicable not only to *L* junctions, but to many other shapes.

Just a little stress analysis work will determine just what increase in section modulus is required to more uniformly distribute the stress in areas of high stress. Stress analysis also can and should be used to apply the fundamentals discussed in this paper to the actual part. These fundamentals indicate the possible best shapes to be used to build the actual part. When these shapes are joined to make any but the simplest working part the stress pattern becomes so complicated that theoretical analysis becomes either impractical, or impossible.

Perhaps the first design will be satisfactory. Perhaps unexpected areas of high stress are present.

*Average values for fatigue life have been used in Fig. 4 for comparison rather than minimum values because of the possibility of defects being the major factor determining the minimum values. The specimens were tested as received, no selection was made. Defects were found in some junctions by radiographic and magnetic particle testing. It makes little difference in the comparison if minimum, or average values are used. The variation in minimum values was even greater than for the average values given. There was no correlation between defects and fatigue life. Junction 4a with the lowest fatigue life showed the fewest defects.

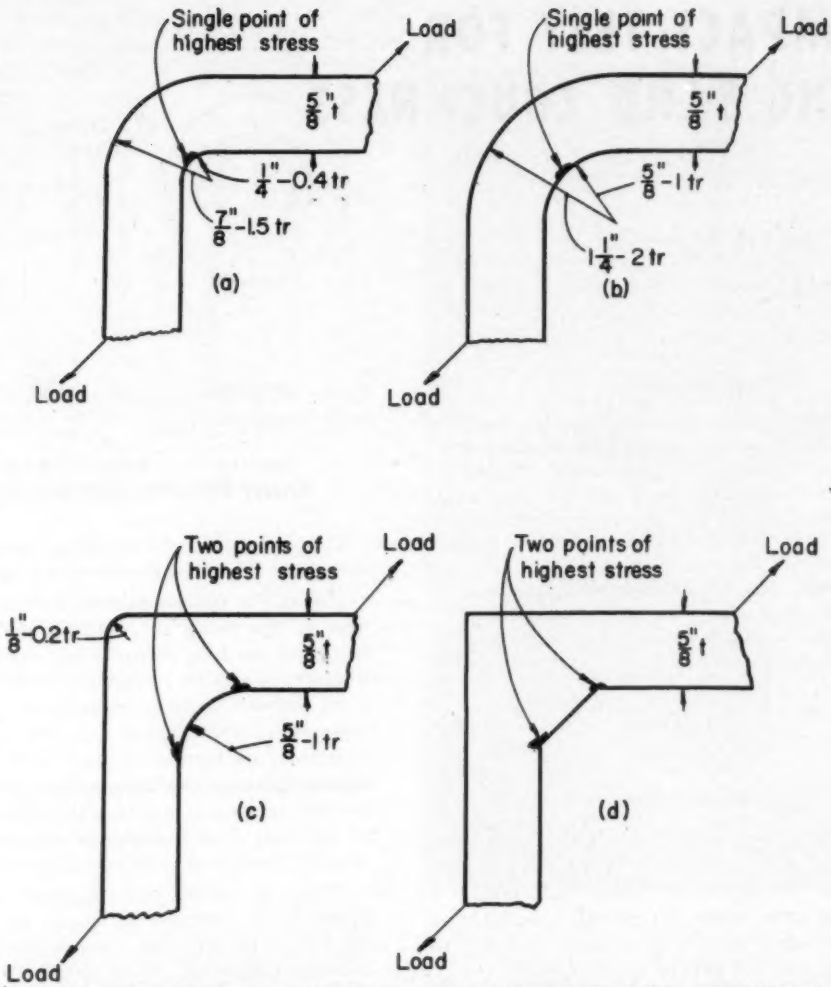



**Fig. 4 — Relationship between shape and fatigue life. (a) Average fatigue life — 270,000 cycles of 1170 lb load. Cast junction. (b) Average fatigue life of 1,850,000 cycles of 1170 lb load. Cast junction. (c) Average fatigue life of 5,900,000 cycles of 1170 lb load. Cast junction. (d) Average fatigue life 601,000 cycles of 1170 lb. load. Machine welded junction stress relieved.**

Stress analysis using relatively simple brittle lacquer techniques will go far as the final check. Brittle lacquer will give the overall picture of stress distribution. The more complicated strain gage techniques are necessary only if accurate quantitative data are needed. There are a number of foundries with stress analysis departments qualified to do this work.

## CONCLUSION

The foundry industry as a whole must be familiar with this work, not just a few foundries. Casting is about the only way the slight changes in shape and size discussed throughout this paper can be efficiently incorporated into a metal part. W. J. Grede in his Hoyt Memorial Lecture last year stressed some worrisome data. He showed the foundry industry to be moribund. True, it is not dying, but it is not growing. The use of ferrous castings, the largest segment of the foundry industry, is not keeping pace with population growth. Casting design offers the hope of reversing this trend—to make castings lead the parade rather than take a back seat to newer, better promoted forming processes.

The promotion of just one simple point, the use of fillet radii equal, or greater than the section thickness will do more to increase service strength, will do more to promote the use of castings, than any half dozen highly advertised "revolutionary" foundry methods.

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# JOLT IMPACT TEST FOR MOLDING SAND TOUGHNESS

by H. W. Dietert and T. E. Barlow


## ABSTRACT

A study of the toughness of molding sand was made as measured by jolt impact tester (AFS nonstandard test), termed green core impact test, pages 167, 168, FOUNDRY SAND HANDBOOK, sixth edition (1952). The length of the overhang for molding sand was increased from 7/16-in. to 9/16-in. to reduce the number of jolts to conservative numbers.

The relationship between various molding sand properties and jolt toughness of sand are shown. These relationships can be used to start a control program on sand toughness which will lead to better rammed molds.

A study of mold quality as affected by sand toughness will prove to be productive research.


## INTRODUCTION

Molding sand as used in the foundry must be both a mobile and a static material. When the sand is rammed around the pattern the sand must possess a mobile property, and while the metal is solidifying in the mold the sand must have a static property to produce castings that are precise in dimensions.

The mobile property of a sand has been termed by many as the rammability or moldability of a sand. The static property which imparts load carrying property, may be measured by creep deformation.

Sand initially is in a static phase, but as soon as force is applied in ramming, a movement or flow occurs which ruptures the bond, allowing for re-arrangement of the sand grains.

When force, that is pressure, is applied and movement occurs during ramming then one accomplishes work.

$$\text{Force} \times \text{Distance} = \text{Work}$$

The work required to place the sand grains firmly and uniformly around the pattern will dictate to a great extent the ramming energy required to obtain a given mold hardness and absence of false mold wall voids. The toughness of the sand will affect the amount of work that has to be expended on the sand.

The factor of sand toughness of a sand becomes of prime importance in all forms of ramming. Sand toughness of a particular sand should be held at a predetermined level and controlled within reasonable limits of the chosen value to insure the production of well rammed molds.

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## SAND TOUGHNESS MEASUREMENT

Many methods of measuring sand toughness are in use today. Three methods will be briefly described.

One of the older methods of measuring sand toughness is by using a compressive strength machine equipped with a recorder to draw the stress-strain diagram of a sand. Another procedure for this method is to simply equip a compressive strength machine with an ultimate green deformation indicator. The indicator mechanism is preferably equipped with a center fulcrum linkage which stops the indicator reading when the rate of collapse first exceeds the rate of loading. This is a precise method, but it requires careful operation since one is measuring to 0.001 in.

When a stress-strain diagram is employed, the problem of locating the point on the graph where strain first exceeds the stress on some sands is open to human judgment. With either method, the stress (lb/sq in. compressive strength) is multiplied by the strain (deformation) and multiplied by 1000 yielding a product that expresses the sand toughness; a factor that influences the work required in ramming.

A sand toughness measurement method that is in use in England[1] is termed the shatter test. In the shatter test, an AFS sand specimen is dropped 6 ft on a steel anvil. In this test, the force is secured in the form of kinetic energy from the drop, causing a movement to rupture the specimen. The fractured sand specimen pieces are caught on a screen with ½-in. mesh. The weight of sand specimen of sufficient size to be retained on the ½-in. mesh is expressed in percentage of the original specimen weight as the shatter index, a toughness measurement.

### *Jolt Toughness Method*

The third method of measuring sand toughness is a new development in applying it to molding sand. This method is termed jolt toughness. The method employs the principle of the AFS Core Impact Test.[2] The only two changes made are in the size of the overhang and specimen holder. The length of overhang when the test is used for molding sand is increased to $^{9}/_{16}$-in. from $^{7}/_{16}$-in. as used on core sands. The molding sand specimen is solidly held in the anvil holder by means of a spring loaded strap.

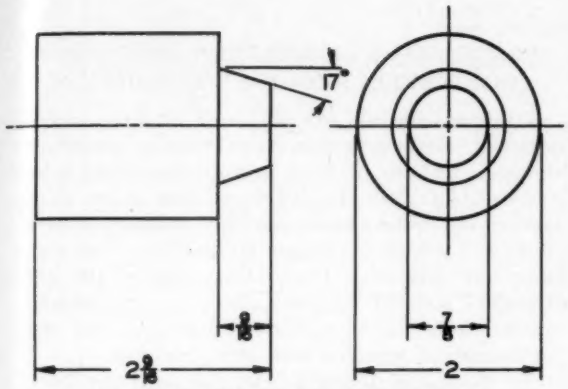



Fig. 1 — Dimensional drawing of the test specimen used to determine the jolt toughness of molding sands.

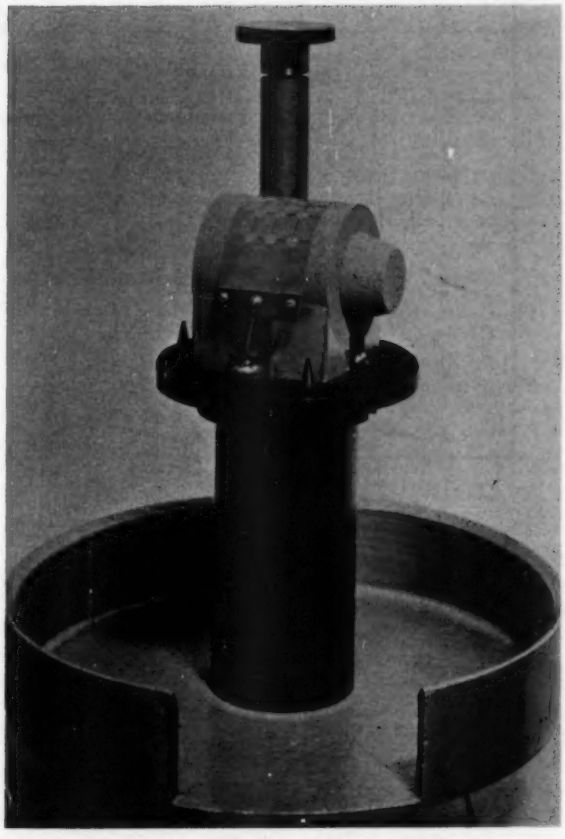


Fig. 2 — Jolt toughness tester used to measure the toughness of sands in this investigation.

The molding sand sample dimensions are shown in Fig. 1. The sand specimen is formed in the conventional AFS manner, with the exception that a ring-shaped pattern is placed in the specimen tube pedestal assembly prior to placing the sand in the tube. The ramming force of the AFS sand rammer causes the sand to flow into the restricted overhang section. As the work required to cause the sand to flow varies, the ramming perfection of the overhang restricted section is affected.

The rammed sand specimen is strapped to a curved-shaped anvil of the jolt impact tester, as illustrated in Fig. 2. The motor drive causes the anvil to repeatedly rise and drop a distance of 0.2 in. The number of jolts required to cause the overhang section to drop off the sand specimen is read from a counter and recorded as the jolt toughness. The authors are of the opinion that this new test, which measures the sand toughness property on a restricted portion of a rammed specimen, offers a test method for studying the toughness property of molding sands.

The relationship that exists between the shatter test and the jolt impact test for fireclay and Western bentonite bonded sands is illustrated in the graphs of Fig. 3.

When the shatter index exceeds 70 per cent for the fireclay bonded sand, the shatter test has a slow rise as toughness increases. This same condition exists for the Western bentonite bonded sand when the shatter index exceeds 90. Thus, the sensitivity of the shatter test evaluation of sand toughness on the tough sand diminishes, while the jolt impact test retains its sensitivity over a wider range of toughness.

### *Sand Toughness Values*

Sand toughness values, as obtained by multiplying the compressive strength by ultimate green deformation, are related to jolt toughness values, as shown by graphs in Fig. 4. The relationship for 80 per cent Southern bentonite, 8.0 per cent Western bentonite and 15 per cent fireclay bonded sands, between jolt toughness and compression deformation sand toughness, indicates that both the strength multiplied by deformation and jolt test methods measure the same property when the sand is rammed to different hardnesses. The jolt impact test, however, is the simpler test to perform. The sands used for the tests in Fig. 4 were all tempered with sufficient water to yield an ultimate green deformation of 0.018 in.

Holding the ultimate green deformation constant yielded a temper point where sufficient moisture was present to soften or plasticize the bond to the same degree. Ultimate green deformation is a useful tool for describing and controlling a degree of temper.

The stress-strain (strength × deformation) sand toughness correlates well with the shatter index, Fig. 5. The rate of shatter index decreases, indicating a reduced sensitivity for the tough sands.

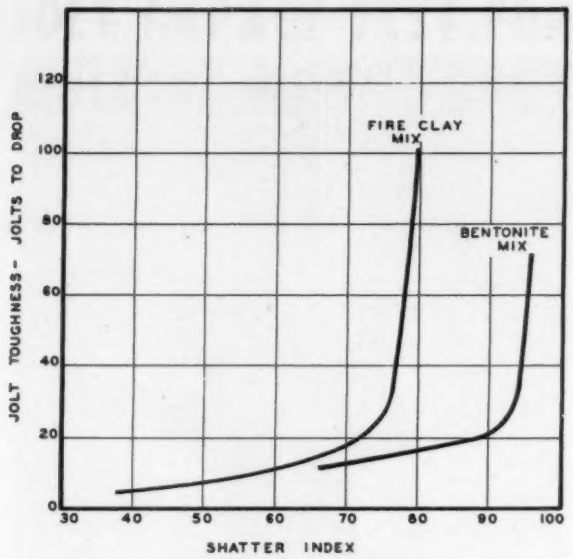



Fig. 3 — The relationship between shatter test values and jolt toughness values.

## GRAIN FINENESS EFFECT ON JOLT TOUGHNESS

In the interest of simplification, the grain fineness of sand will be described by using the AFS Grain fineness number. Bonding sand mixes consisting of sand grains of different AFS Grain fineness numbers with 6 per cent Western bentonite and tempered to 0.018 in. ultimate green deformation yielded sands having little differences in jolt toughness, as illustrated in Fig. 6. For all practical purposes, jolt toughness is not affected by a change in grain size.

## PERCENTAGE OF WESTERN BENTONITE BOND EFFECT ON JOLT TOUGHNESS

An interesting and practical sand condition may be obtained when the percentage of Western bentonite is increased and the ultimate green deformation is held constant, for example, 0.018 in. The bond in the sand is not fully wetted but only sufficiently wetted to yield a constant ultimate deformation, i.e., a constant total plasticity. Under such control, the green strength increases substantially at a rate which is similar to the rate of increase of jolt toughness, as the percentage of Western bentonite increases.

An increase of Western bentonite increases the jolt toughness. The amount of jolt toughness increase will depend on the degree of temper, when other conditions are held constant. In the case of the sand mixes described for Fig. 7, the degree of temper was held in check by tempering to constant ultimate green deformation. Thus, limiting the increase in jolt toughness making for relatively easy ramming of the highly bonded sands.

## MOISTURE AND ADDITIVES EFFECT ON JOLT TOUGHNESS

Water may be termed an additive, and under this premise sand mixes containing additives such as 2 per cent wood flour, 15 per cent silica flour, were mixed with different amounts of water additions, the base sand being a Michigan City Lake Sand of 58 AFS Grain fineness number. The effect of these additives on the jolt toughness is illustrated in Fig. 8.

Noting first the jolt toughness of a 6.0 per cent Western bentonite bonded sand with water as the only additive, it becomes apparent that moisture addi-

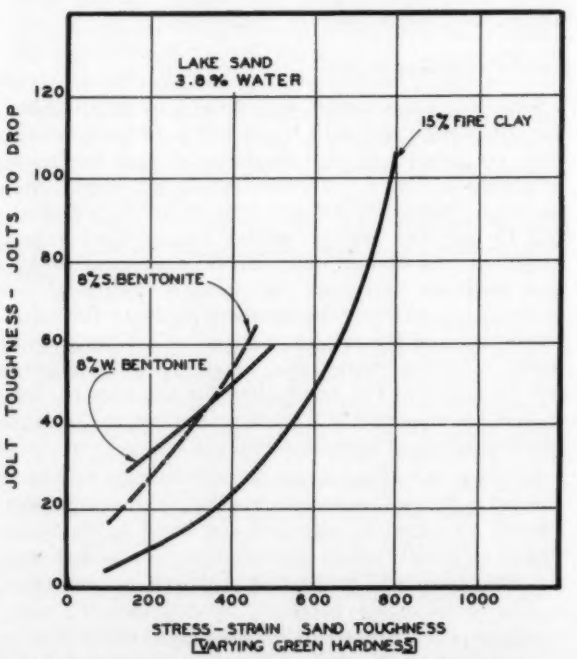



Fig. 4 — The relationship between the stress-strain sand toughness and jolt toughness.

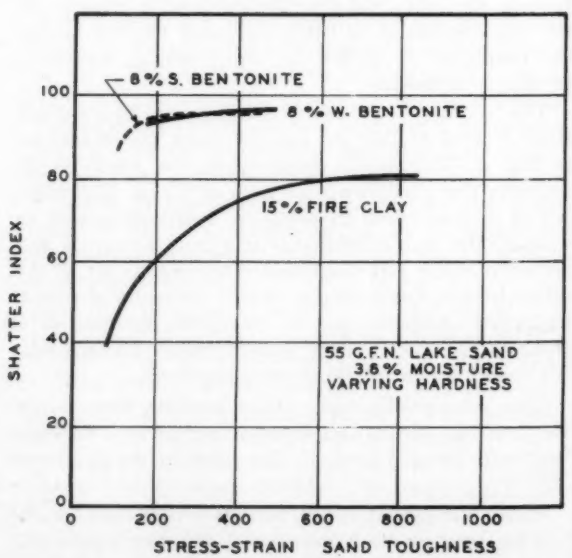



Fig. 5 — The relationship between shatter index and stress-strain sand toughness.

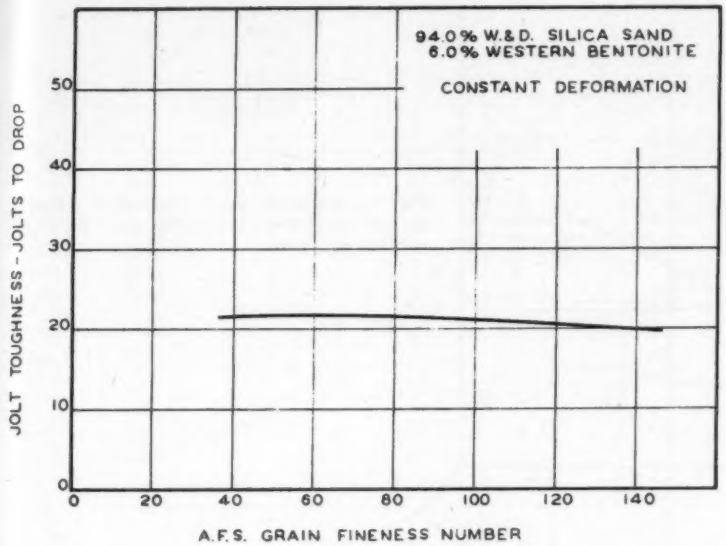



Fig. 6 — The jolt toughness of a bentonite bonded sand is not greatly affected by a change in the grain fineness up to 140 mesh grain.

tions below temper point of 0.018 in. ultimate green deformation (U.G.D.) influences the jolt toughness greatly. The 0.018 in. U.G.D. points are marked by *X* on the graphs, Fig. 8. A slight drop in jolt toughness is experienced above 2.5 per cent moisture.

When 2 per cent wood flour is added to the previous mix the jolt toughness is reduced for the particular cellulose material used, and the sand is tempered below the 0.018 in. U.G.D. point. As moisture is increased above the 0.018 in. U.G.D. point, the jolt toughness is increased above the jolt toughness of a straight bentonite bonded sand. A sand of low toughness can be produced by working the sand on the dry side of temper. The addition of 15 per cent silica flour produces a sand of low jolt toughness.

Continuing the study of the effect of additives on jolt toughness, by referring to Fig. 9, it may be noted that an addition of 5.0 per cent sea coal increases the jolt toughness. The addition of 2 per cent cereal increases the jolt toughness rapidly as moisture is increased. Ample moisture must be present, however, to plasticize the cereal to make it effective. Running moisture low may actually cause the cereal-bentonite bonded sand to possess a low jolt toughness. Additives materially affect jolt toughness of sands.

## MOLD HARDNESS AFFECTS THE JOLT TOUGHNESS

An increase of mold hardness will increase the strength of a sand while it will not alter the ultimate green deformation value. An increase of mold hardness, particularly so, above 75 green hardness, will markedly increase the jolt toughness whether the

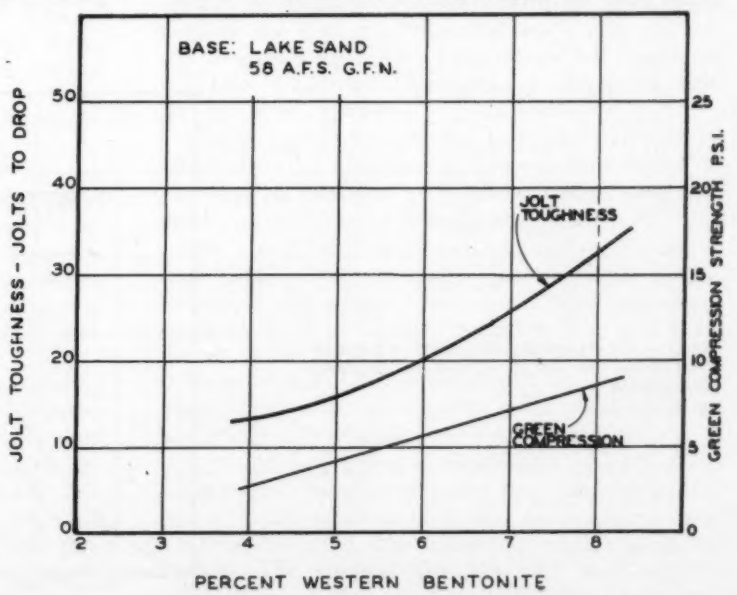



Fig. 7 — An increase of Western bentonite bond in a sand increases the jolt toughness.

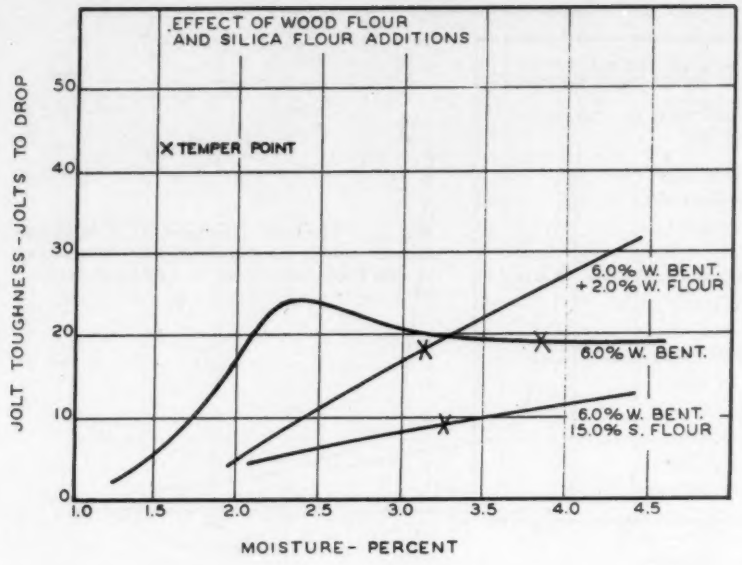



**Fig. 8 — Moisture content materially changes the jolt toughness of bentonite sand with and without additives.**

sand is bonded with Western bentonite, Southern bentonite or fireclay (Fig. 10).

The bentonite bonded sands produce a greater jolt toughness more readily than some fireclay bonded sands. However, they also produce a higher mold hardness for a given ramming energy. Some fireclays contain upward of 40 per cent inert fine material that retards jolt toughness, similar in action as silica flour retarding jolt toughness.

## INFLUENCE OF JOLT TOUGHNESS IN RAMMING OF MOLDS

Many of the casting losses are produced by the presence on the mold face of false void spaces that are not comparable with the grain fineness. False void spaces make for penetration and rough finish. A false void space is secured when adjacent sand grains do not pack together. They leave a void space that is not comparable to the fineness of the sand grains.

A flowability indicator attached to the top of a sand rammer may be used to measure the degree of false void presence. The greater the indicator flowability percentage the smaller the false void spaces for a given grain fineness. This correlation holds true for many sands.

The comparison between the jolt toughness and the above mentioned flowability percentage is illustrated in Fig. 11. The graph in Fig. 11 shows that as the jolt toughness decreases, flowability increases. Stated in another manner, an excess of jolt toughness hinders the ramming of a good mold face with a minimum of false void spaces. The percentage of moisture used in the sand mixes for Fig. 11 are marked on the graph.

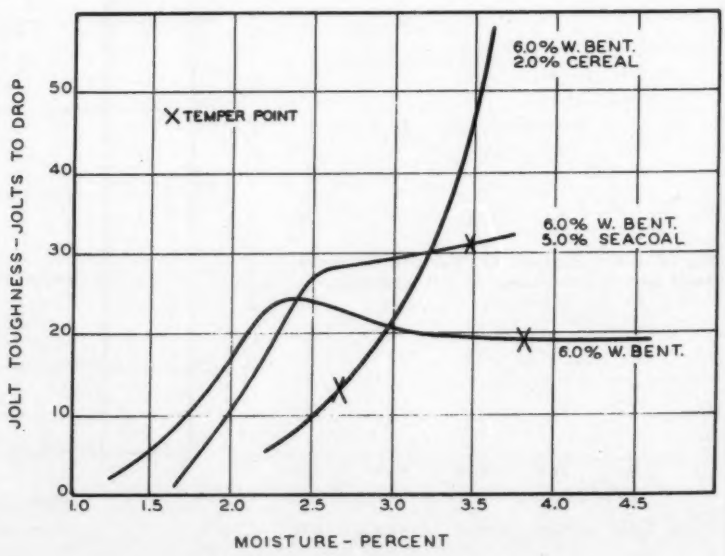



**Fig. 9 — Cereal and sea coal addition materially changes the jolt toughness of bentonite bonded sand.**

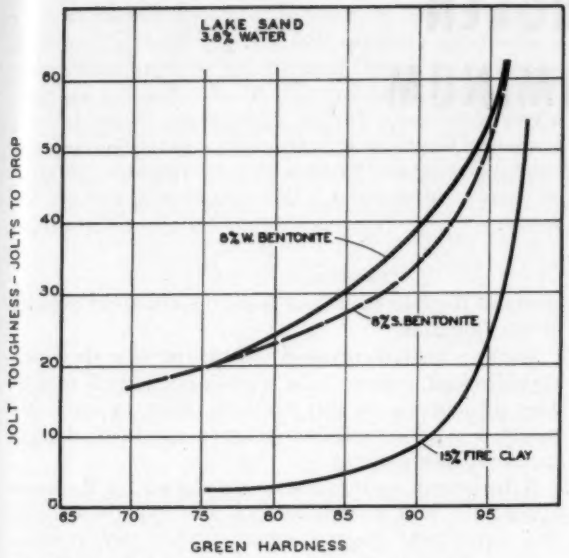



Fig. 10 — An increase in mold hardness increases the jolt toughness of a bentonite bonded sand.

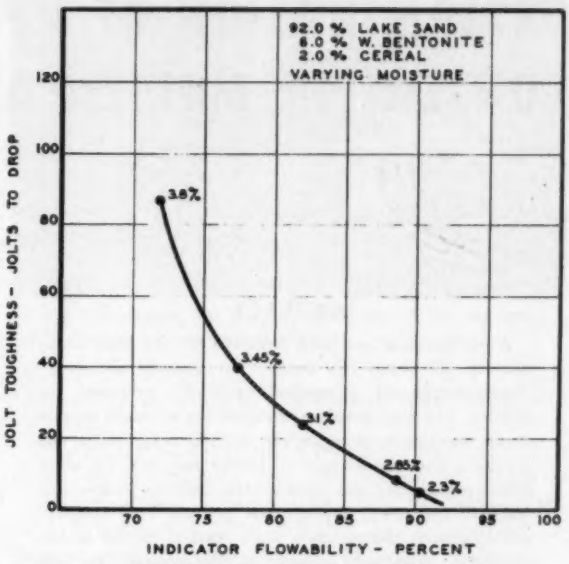



Fig. 11 — As the jolt toughness of a bentonite cereal-bonded sand decreases, the indicator flowability values increase, reducing false void spaces on mold wall surfaces.

## CONCLUSIONS

A relationship exists between the shatter test and the jolt toughness test. On toughness above 70 for fireclay bonded sand, and 80 for bentonite bonded sand, the shatter test shows a slow increase in toughness. The jolt toughness test on the high toughness sands read progressively higher.

The jolt toughness does not vary greatly between Western and Southern bentonite bonded sands. The toughness of fireclay bonded sands show a different degree of toughness from bentonite bonded sands. The stress-strain sand toughness values bear a close relationship to jolt toughness values. The two tests undoubtedly measure the same mechanical property of a sand.

The jolt toughness of sand is not greatly affected by a change in grain fineness. An increase in bonding ingredient present in a sand that is plasticized with sufficient moisture increases jolt toughness. Bentonite bonded sands without additives rapidly develop a jolt toughness, and then reduce in toughness as temper is approached. The addition of silica flour materially reduces the jolt toughness.

The addition of certain wood flour materially reduces the jolt toughness in moisture range below temper. When moisture is above temper, the jolt toughness is increased. Many cellulose materials are great water carriers. The addition of cereal with low moisture materially reduces jolt toughness, but rapidly increases jolt toughness as moisture increases. The addition of sea coal at temper increases the jolt toughness.

The jolt toughness of sands increases as mold hardness increases. The greater the mold hardness, the more rapidly sand toughness increases. This is particularly true in the case of fireclay bonded sands. The sand offers greater resistance to flow as hardness approaches the high mold hardness, for example, 90 and above.

Sand toughness increases less rapidly in the low mold hardness range than in the higher mold hardness range. In this range, fireclay bonds increase more readily than do the bentonites. A relationship exists between the indicator flowability test values and the jolt toughness for a given sand. A low jolt toughness will allow reaching a high indicator flowability with the AFS sand rammer.

Tempering sands with just sufficient moisture to soften the clay substance to a controlled degree of plasticity, as measured by ultimate deformation, will yield a control over jolt toughness.

The use of additives produced high quality mold wall surfaces when the temper point by ultimate deformation was selected at a conservative value, for example, 0.018 to 0.020-in. range for iron and nonferrous sands. Herein lies a possibility on producing superior castings. A combination of an additive and improper moisture can lead to lower mold hardness and greater false void spaces.

It is well to study each sand mix separately—running careful laboratory tests and then ramming molds, noting mold hardness and false void spaces, then pouring the mold and making a careful study of castings. The quality of castings can be materially improved for a given standardized ramming by following a program discussed herein.

# INITIAL BUBBLE TEST FOR DETERMINATION OF HYDROGEN CONTENT IN MOLTEN ALUMINUM

by D. J. Neil and A. C. Burr


## ABSTRACT

**An instrument has been designed for the determination of the hydrogen content of molten aluminum alloys, employing a method originally proposed by Dardel. The test involves maintaining a small sample of an aluminum alloy in the molten state while the pressure over the sample is slowly reduced. At some pressure less than one atmosphere, bubbles of gas will nucleate and break through the surface. This pressure and specimen temperature were related to the actual specimen hydrogen content as determined by the vacuum subfusion method, over a wide range of gas content. Sievert's Law was verified, and the constant of proportionality relating $\sqrt{p}$ and the hydrogen solubility was found to vary with alloy composition as expected. Nomograms were constructed wherewith hydrogen content could be determined quickly from the test data.**

**It is suggested that this test offers advantages over other tests for the determination of the hydrogen content of aluminum alloy melts, and that such determinations can be made with a precision of $\pm 0.05$ ml $H_2$ at N.T.P./100g or better.**


## INTRODUCTION

In recent years great emphasis has been placed on quality control of wrought aluminum alloys. This is especially true where the finished products are critical components of aircraft or missiles. It is generally accepted that the hydrogen content of aluminum alloys is one of the factors which must be controlled, since in the processes of casting, thermal treatments and working, internal defects may occur which result directly or indirectly from an undesirably high melt hydrogen content.

Remedial measures such as gaseous fluxing of furnace charges will effect reductions in the hydrogen content, but since reproducible results are difficult to obtain, it is desirable, if not necessary, to be able to measure the hydrogen content of the molten metal.

Various methods have been used to this end over the years. The most common has been the Straube-Pfeiffer test or modifications of it, in which a small sample of molten aluminum alloy is permitted to solidify under a reduced pressure. During solidification hydrogen is rejected from solution and appears as porosity within the specimen. A qualitative estimate of the gas content is obtained from the convexity of the free surface or from the apparent density of the specimen.

Another method involves the casting of a specimen of controlled volume in a resin-bonded shell mold.[1] The proponents of this method claim to have obtained a good correlation between specimen density and hydrogen content.

A hydrogen content meter developed by Ransley[2] appears to be a promising tool, but experience with this equipment suggests that further development work is necessary to achieve characteristics necessary for convenient and low maintenance operation.

The apparatus described in the following paragraphs was developed jointly by Aluminium Laboratories Ltd. and the Aluminum Co. of Canada, Ltd. The method upon which the design is based was proposed initially by Dardel.[3,4]*

## THEORY OF OPERATION

It is well known that the solubility of hydrogen in molten aluminum alloys varies with temperature and molecular hydrogen pressure according to the relationship

$$\log_{10} S = 0.5 \log_{10} p - \frac{A}{T} + B \qquad (1)$$

where

S = solubility in ml/100 g.
p = pressure in mm Hg.
T = absolute temperature (K).

and A and B are positive constants.

Values of the constants *A* and *B* for pure aluminum have been published by Ransley and Neufeld,[5] Opie and Grant[6] and Hofmann and Maatsch.[7] Opie and Grant also presented values of *A* and *B* for the binary systems Al-Cu and Al-Si.

Consider a sample of an aluminum alloy held above the melting point at temperature *T* and containing *S* cu cm of hydrogen. If the pressure over the sample were reduced below *p*, hydrogen would precipitate from solution and appear as bubbles. This aspect of the problem has been discussed by Dardel.[3]

Since it has been demonstrated that hydrogen solubility in aluminum alloys is affected by compositional

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*Since preparation of this paper an account of the application of the initial bubble test to hydrogen determination in light alloys has come to the attention of the authors. The reader is referred to M. V. Sharov and A. P. Gudchenko, "A Study of the Interaction of Hydrogen with Light Alloys during the Melting Process," *Metallurgicheskie Osnovy Lit'ya Legkikh Splavov*, Oborongiz, Moscow, p. 306 (1957).



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changes, a calculation of the hydrogen content from the temperature $T$ and pressure $p$ at which bubbles of gas appear, is possible only if the constants $A$ and $B$ are known. Another approach is to obtain chill cast specimens corresponding to each set of values of $T$ and $p$. The hydrogen content of these specimens is then determined by the vacuum subfusion or hot extraction method.[8] If all determinations of the pressure at which gas bubbles appear were made at the same temperature a calibration curve could be plotted directly; however, this is difficult in practice. Since for a given gas content and a temperature above the liquidus

$$\sqrt{p}\, e^{-\frac{k}{T}} = \text{constant} \qquad (2)$$

where

$$k = \frac{A}{\log_{10} e} \qquad (3)$$

then a knowledge of the constant $A$ only, would enable one to adjust observed pressure data to a common reference temperature $T_r$. Ransley and Neufeld[5] found $A$ to be 2760 while Opie and Grant[6] report 2550 for pure aluminum. These authors also report values of $A$ from 2550 to 3150 for the binary systems Al-Cu (up to 32 per cent Cu) and Al-Si (up to 16 per cent Si).

Since

$$\sqrt{p_o}\, e^{-\frac{k}{T_o}} = \sqrt{p_r}\, e^{-\frac{k}{T_r}} \qquad (4)$$

where subscripts $o$ and $r$ signify observed and reference respectively, the effect of changes in the value of $A$ may be calculated. The results are presented in Table 1.

TABLE 1 — EFFECT OF k ON CORRECTED PRESSURE DATA

| | $k = 5872$ | $k = 6355$ | $k = 7253$ |
|---|---|---|---|
| $k/T_r$ | 6.227 | 6.739 | 7.691 |
| $k/T_o$ | 6.035 | 6.531 | 7.454 |
| $k\left(\frac{1}{T_r}-\frac{1}{T_o}\right)$ | 0.192 | 0.208 | 0.237 |
| $e^{k\left(\frac{1}{T_r}-\frac{1}{T_o}\right)}$ | 1.212 | 1.231 | 1.267 |

$T_o = 973$ K for illustration.
$T_r = 943$ K.

On the basis of the relatively small effect on the corrected values of $\sqrt{p}$, 6355 was chosen as a reasonable value of $k$ for application to commercial aluminum alloys.

## HYDROGEN CONTENT TESTER

The present form of the tester is illustrated in Fig. 1. Four basic components are involved, of which the vacuum pump is not shown. The instrument cabinet contains the pressure and temperature gages as well as a vacuum manifold, vacuum control valves and necessary outlets and switches to provide power to the vacuum pump and heater. The other two components are the heater unit and cover unit shown, respectively, in the right and left foreground of Fig. 1. The vacuum pump used has a free air capacity of 10 litres/min and a guaranteed vacuum of 0.3 microns Hg. Any pump of these characteristics will work with the tester.



Fig. 1 — Tester for the determination of the gas content of molten aluminum alloys.

The heater unit contains a boron nitride crucible of about 50 ml capacity, on the outside of which is wound directly a chromel $A$ heater of 325 watts, rated at 115 volts A.C. The crucible heater assembly is located within a cast aluminum vessel. Heat losses are reduced by approximately 1½-in. of insulation between the crucible and the vessel walls and bottom. The crucible is held in place by means of a recess machined in the stainless steel cover on the heater unit. Electrical connections are brought out through the handle by means of hermetic seals.

The cover unit contains a glass viewing port sealed in position with O-rings, and a chromel-alumel thermocouple sheathed in stainless steel. The thermocouple leads pass through the handle of the cover unit and inside the rubber vacuum line to the manifold in the instrument cabinet, where they are brought out through hermetic seals to the temperature gage. The flange of the cover unit contains a recessed O-ring which mates with the flange integral with the heater unit to provide the vacuum seal.

The vacuum gage has an absolute pressure range of 0 - 200 mm Hg, and the temperature gage is compensated with a suitable thermistor to correct for variations in ambient temperature.

## OPERATION

After having brought the crucible to the operating temperature and coated the stainless steel thermocouple sheath with a wash to prevent attack by molten aluminum, the instrument is ready for use.

Using a clean dry ladle a sample of molten metal is transferred into the tester crucible. The cover unit is placed over the heater unit, thus introducing the thermocouple into the specimen, and providing for continuous temperature measurement. The heater is so designed that metal held in the crucible will remain molten at a temperature which will eventually rise to approximately 760 C (1400 F). Having first closed the vacuum release valve the vacuum pump is switched on and the needle valve adjusted so that the pressure decreases at the rate of approximately 5 - 10 mm Hg/sec. Since the pressure gage is scaled from 0 - 200 mm Hg (absolute), the pointer will not

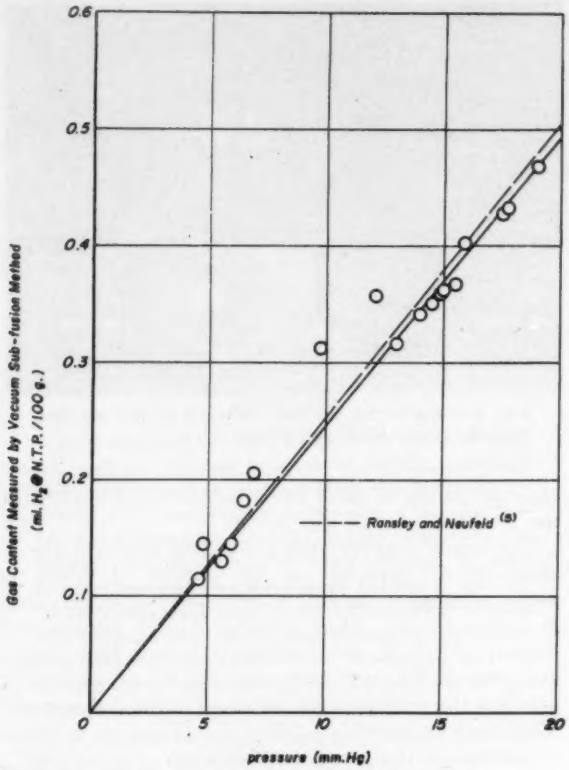



Fig. 2 — Calibration curve for aluminum 2S alloy.

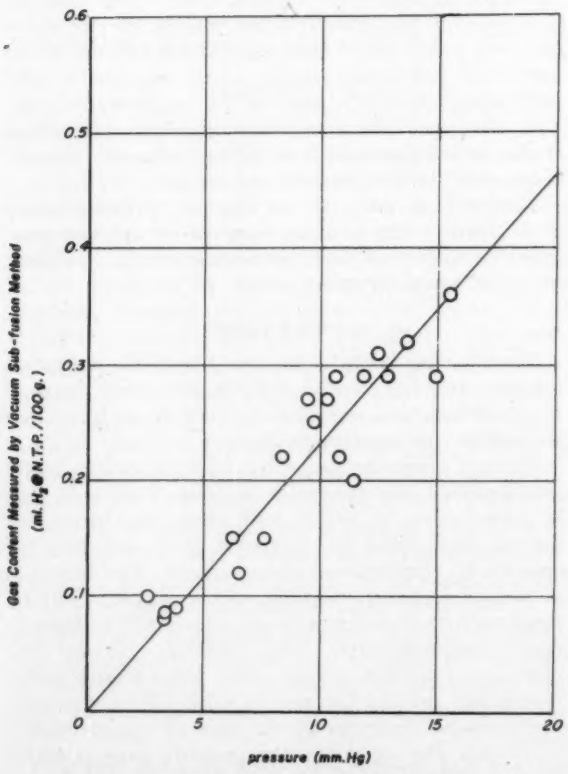



Fig. 3 — Calibration curve for aluminum 3S alloy.

move until the pressure has been reduced below 200 mm.

While the pressure is being reduced the specimen surface is observed with care. At some point bubbles of hydrogen will appear and the pressure and temperature are immediately noted. A further reduction in pressure will usually produce more bubbles to confirm the original observation. The pressure is then returned to atmospheric by switching off the vacuum pump and opening the vacuum release valve. It has been found desirable to make at least three observations on any melt since spurious indications occur occasionally. These spurious indications may result from air entrapped in dross adhering to the surface of the crucible. This possibility suggests that the crucible should be kept as clean as possible. It is also considered that spurious results might occur as a result of the presence of nuclei of solid foreign material within the specimen.

## RESULTS

Small experimental melts were made up in aluminum alloys 2S, 3S, 57S and 75S compositions (Table 2) and chill-cast specimens were obtained covering a range of gas levels. Corresponding pressure and temperature data were obtained using the initial bubble test. The chill cast specimens were analyzed for hydrogen content by the vacuum subfusion method and the initial bubble test data were corrected to a reference temperature of 670 C (1400 F) using the relationship

$$p = p_o \, e^{2k\left(\frac{1}{T_r} - \frac{1}{T_o}\right)} \qquad (5)$$

where the subscripts have already been defined. The subfusion hydrogen content values were plotted against the square root of the corrected pressure values for the four alloys and the results are shown in Figs. 2, 3, 4 and 5.

TABLE 2 — CHEMICAL COMPOSITIONS FOR ALUMINUM ALLOYS EXAMINED*
(weight per cent)

| Alloy | Aluminum Assoc. Equivalent | Cu | Fe | Mg | Mn | Si | Zn | Ti | Cr | Other Total |
|---|---|---|---|---|---|---|---|---|---|---|
| 2S | 1100 | 0.20 | ‡ | — | 0.05 | ‡ | 0.10 | — | — | 0.15 |
| 3S | 3003 | 0.20 | 0.70 | — | 1.5 | 0.60 | 0.10 | — | — | 0.15 |
| 57S | 5052 | 0.10 | ** | 2.8 | 0.10 | ** | 0.10 | — | 0.35 | 0.15 |
| 75S | 7075 | 2.0 | 0.70 | 2.9 | 0.30 | 0.50 | 6.1 | 0.2 | 0.40 | 0.15 |

*All values quoted are upper limit.
‡Fe + Si = 1.0% Max.
**Fe + Si = 0.45% Max.

## DISCUSSION

It will be noted that the results indicate the anticipated linear relationship between hydrogen content and $\sqrt{p}$. It will also be noted that the slopes of the best-fit lines have been altered as a result of compositional changes. Figure 2 shows the line plotted from data of Ransley and Neufeld[5] and the agreement is quite satisfactory. This agreement suggests that confidence may be placed in the method.

The scatter in the results appears to be more pronounced in the case of the stronger alloys. It has been decided that this fact can be attributed in part to the nature of the specimen oxide surface and the re-

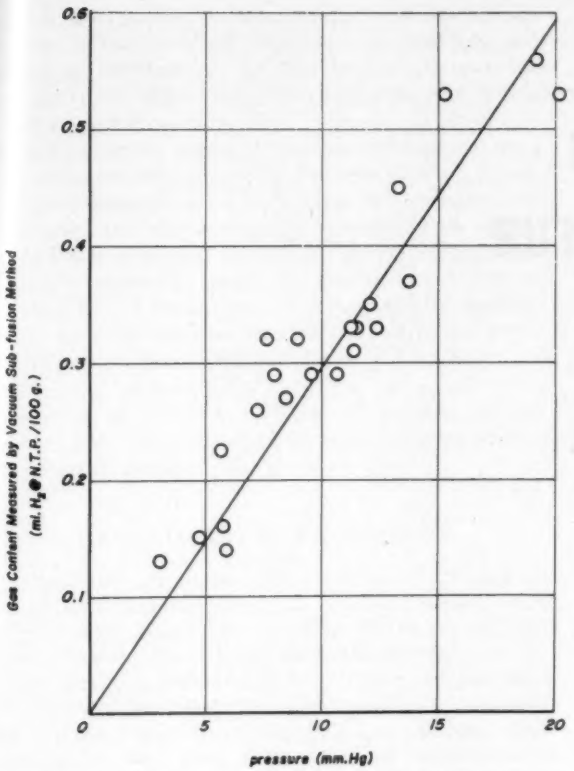



Fig. 4 — Calibration curve for aluminum 57S alloy.

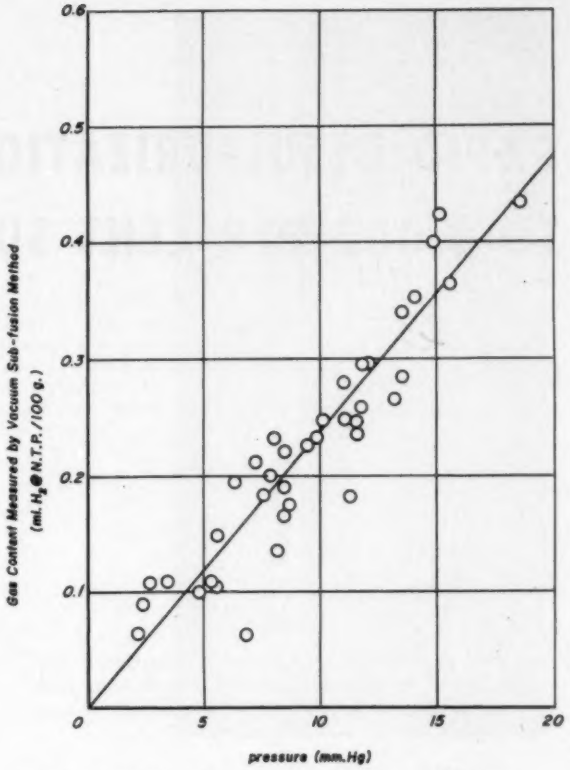



Fig. 5 — Calibration curve for aluminum 75S alloy.

lative difficulty in observing the first bubbles. In spite of the scatter, melt gas contents can be determined by this method within ± 0.05 ml $H_2$/100 g and furthermore, an individual determination can be made in approximately 2 min.

To facilitate interpretation of the pressure and temperature data from the test, it is possible to construct an appropriate nomogram. The nomogram scales are based on the previously mentioned assumption concerning the effect of temperature on hydrogen solubility. A sample nomogram is shown in Fig. 6.

The present evidence suggests that individual calibration curves may be necessary, especially in the case of major compositional changes. It is known for example, that silicon and copper additions depress the hydrogen solubility in aluminum while magnesium has the opposite effect.

The attractive feature of the method is the fact that actual melt hydrogen content in conventional units can be obtained in a matter of a few minutes. One must bear in mind, however, that the accuracy of gas content measurements appears to be more or less inversely proportional to the time involved in the determination. In spite of this the tester has been demonstrated to be a useful tool in melt gas content control.

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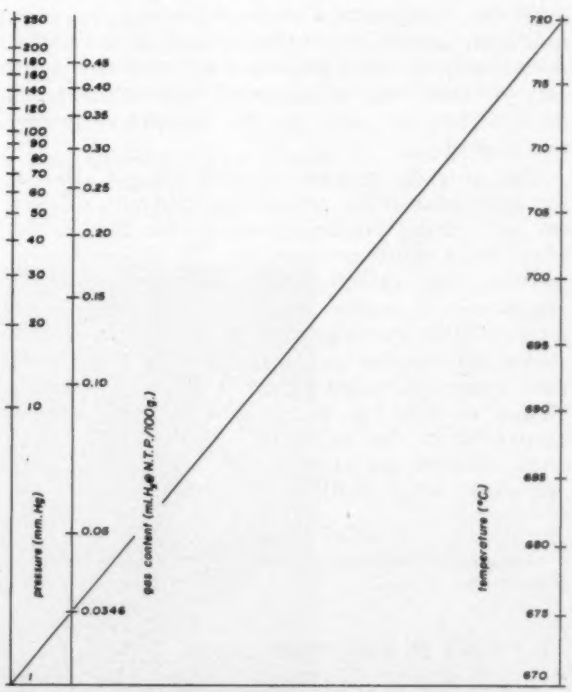



Fig. 6 — Nomogram for determination of hydrogen content of aluminum 2S alloy melts using tester data.

# RAPID DESULFURIZATION TO 0.002 PER CENT SULFUR

by Edward J. Dunn, Jr.


## ABSTRACT

**The development of a simple desulfurization procedure that rapidly and consistantly reduces sulfur from initial values as high as 0.024 per cent to final residuals of 0.002/0.003 per cent is presented.**


## INTRODUCTION

Various investigations have shown that increasing amounts of sulfur in a steel composition decreases its ductility, toughness, hardenability and increases surface defects,[1] red-shortness[2] and porosity in welding.[3] Since the majority of these investigations were conducted with steels containing normal and higher sulfur residuals comparatively little data are available to show the influence of low sulfur levels (below 0.010 per cent) on the mechanical properties of steel.

The probable reasons for such limited data at the low sulfur levels are (1) the difficulty of present air melting practices to reproduce final sulfur levels below 0.010 per cent and (2) the alternative methods, such as using high purity melting stock and/or special vacuum techniques are not economically feasable for large tonnage production. Nevertheless, investigators such as Hodge et al[4] have shown that transverse toughness in a steel composition similar to 4330 can be as much as 50 per cent* greater when the sulfur is lowered from initial levels of 0.010 per cent to final residuals of 0.005 per cent, while further improvements might be expected at sulfur levels of 0.002 per cent sulfur.

The following investigation describes the development of an effective, reproducible and practical technique for the rapid desulfurization of steels to levels as low as 0.002 per cent sulfur, such that improvements in mechanical properties which are related to sulfur contents may be commercially realized.

Because of the author's immediate interest in silicon modified aircraft steels, effort in this investigation was concentrated on the desulfurization of a 4340 composition modified with 1/1.5 per cent Si.

## GENERAL CONSIDERATIONS

Sulfur in steel is reduced in the arc furnace by the use of a strongly basic and reducing carbidic slag, and is generally represented by chemical reactions such as:

$$2CaO + CaC_2 + 3FeS = 3CaS + 2CO + 3Fe$$[5]

This reaction occurs at the slag-metal interface, and is limited in its effectiveness because of the small metal-to-slag contact and the poor sulfur diffusion through the apparent metal and slag boundary layers. This, consequently, necessitates long refining times under the reducing slag in order to reduce the sulfur to any appreciable degree.

It can be assumed that a more rapid and effective sulfur removal can be accomplished if the desulfurizing reaction is brought into intimate contact with the greater bulk of the metal. One way to accomplish this is by immersing an element or compound (which forms an insoluble stable sulfide) into the metal bath, and permitting the sulfide reaction to occur through the mass of the metal rather than at the slag-metal interface.

Of the more familiar sulfide forming elements, Ce, Ca and Mg can be considered to form some

*These values were compared at the 40 Rc and the 300 F test temperature.

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(Work on this paper was performed when the author was associated with Watertown Arsenal, Rodman Laboratory, Watertown, Mass.)

of the most stable sulfides at steel making temperatures, as indicated by their relative standard free energy of formations,[6] i.e., the greater the negative free energy of formation of a compound, the greater is its probable stability.

Of these three sulfides, CeS has the highest negative standard free energy of formation at steel making temperatures with CaS somewhat smaller and MgS with the lowest negative standard free energy. Since Ce is available in limited quantities and at a relatively premium price, the formation of CeS by the use of Ce would not be economically feasible, while the formation of MgS by the use of elemental Mg presents a possible explosive hazard.

Thus an investigation into the effect of a Ca immersion in a steel bath holds the greatest immediate promise of a practical and economical technique for rapid desulfurization of steel.



Fig. 1 — Complete holding device showing cup container at bottom, rod attachment, weight and crane loop on top.

## EXPERIMENTAL PROCEDURE

Using the required conditions presently known for effective desulfurization in the arc furnace, such as a strongly basic and reducing slag,[5] and adding to this the supposition that desulfurization can be more readily accomplished if an element (or compound) which forms a stable sulfide is added directly into the metal bath, this procedure was devised:

1. Melt down in a basic-lined furnace.
2. Skim off melt down slag to remove as many acid melt down constituents as possible.
3. Reduce or deoxidize the bath directly, rather than via a reducing slag, by the addition of a normal amount of Al (0.1 per cent) (This was done at 2900 F).
4. Immediately add unslaked lime to a depth of one to 2 in. to form a strongly basic slag.
5. At a temperature high enough to prevent freeze-up (2900 F), take a metal sample for a complete chemical analysis (including sulfur), and immediately inject Ca into the metal bath at arbitrary amount of 0.6 per cent of the charge.

Because of the cost of metallic Ca, it was immediately substituted by a cheaper form of Ca, i.e., CaMnSi and the bulk of the investigation was conducted with CaMnSi which was added in sufficient quantities to produce a 1.0/1.5 per cent Si residual (2 per cent CaMnSi). This Ca or CaMnSi was injected into the bath by means of an expanded metal screen cage attached to a long weighted bar and manipulated by a crane (Figs. 1-4).

6. Tap immediately (The additions cooled the metal approximately 80 F, resulting in a desirable tapping temperature of approximately 2820 F).

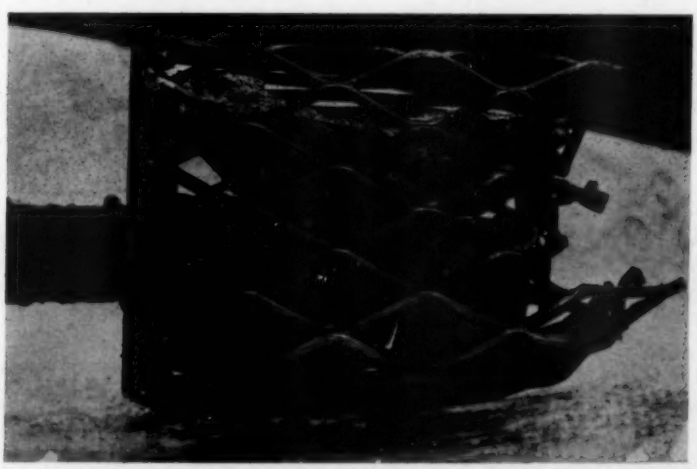


Fig. 2 — Closeup of cup container with CaMnSi inside.

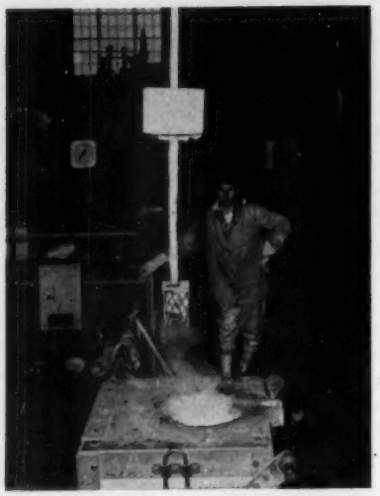


Fig. 3 — The Ca-MnSi and its holding device before lowering into furnace.



Fig. 4 — Reaction caused by CaMnSi injection.

Induction furnaces of 300 and 600 lb capacities were used in the development of the desulfurization procedure, afterwhich it was proved on 1000 and 2400 lb induction furnace melts.

## RESULTS AND DISCUSSION

The proposed method of desulfurization quickly and effectively reduced the sulfur content of a medium alloy steel from average initial sulfur contents of 0.020 per cent to final residuals of 0.006 per cent (Table 1). Excessive ingot porosity and a drastic Al reduction was observed with this desulfurization, and it was necessary to increase the Al deoxidation from 0.1 to 0.15 per cent to eliminate this porosity. With such a modification in the procedure, not only was the porosity eliminated but desulfurization was also increased. For example, when the Al addition was increased to 0.15 per cent the same Ca or CaMnSi addition reduced the original average sulfur contents of 0.022 per cent to final average sulfurs of 0.002 per cent (Table 2).

This desulfurization occurred rapidly, for an analysis of the metal pin samples taken immediately before and after the Ca or CaMnSi addition showed that the greater bulk of the sulfur was immediately reduced while only a little desulfurization occurred during tapping, i.e., a 0.002 per cent sulfur reduction generally occurred during tapping. Along with the sulfur and Al reduction in the metal and their apparent subsequent increase in the slag, an increase of Fe, Mn and Si in the slag was also suggested by typical slag analysis taken before and after 2 per cent CaMnSi additions (Table 3). The reported reduction of CaO in the slag after the addition is apparently the reflection of an increase in one or all of the other constituents in the slag.

TABLE 1 — DESULFURIZATION WITH INITIAL MELTING PRACTICE

| Heat No. | Per Cent | | | | | | | | | | Wt. of Addition/Wt. of Charge, % | Type of Addition |
|---|---|---|---|---|---|---|---|---|---|---|---|---|
| | C | Mn | Si | Ni | Cr | Mo | P | S | Al | Al (added) | | |
| 1 | 0.24 (B.A.) * | 0.51 | 0.23 | 1.78 | 0.88 | 0.34 | — | 0.019 | 0.08 | 0.1 | 2 | CaMnSi |
| | 0.24 (A.A.) ** | 0.71 | 1.02 | 1.76 | 0.81 | 0.31 | — | 0.006 | 0.02 | | | |
| 2 | 0.32 | 0.77 | 0.28 | 1.76 | 0.89 | 0.33 | 0.012 | 0.016 | 0.05 | 0.1 | 2 | CaMnSi |
| | 0.30 | 1.12 | 1.22 | 1.72 | 0.80 | 0.31 | 0.009 | 0.005 | 0.02 | | | |
| 3 | 0.33 | 0.82 | 0.24 | 1.83 | 0.84 | 0.31 | 0.009 | 0.024 | 0.07 | 0.1 | 2 | CaMnSi |
| | 0.33 | 1.11 | 1.65 | 1.80 | 0.90 | 0.29 | 0.009 | 0.005 | 0.02 | | | |
| 4 | 0.36 | 0.86 | 0.24 | 1.72 | 0.82 | 0.32 | 0.012 | 0.017 | 0.05 | 0.1 | 2.3 | CaMnSi |
| | 0.36 | 1.20 | 1.44 | 1.67 | 0.84 | 0.30 | 0.012 | 0.003 | 0.02 | | | |
| 5 | 0.42 | 0.66 | 0.15 | 1.93 | 0.97 | 0.32 | 0.010 | 0.022 | 0.07 | 0.1 | 2.3 | CaMnSi |
| | 0.42 | 0.92 | 1.25 | 1.84 | 0.97 | 0.29 | 0.011 | 0.005 | 0.03 | | | |
| 6 | — | — | — | — | — | — | 0.010 | 0.025 | — | 0.1 | 0.6 | Ca |
| | 1.12 | 0.86 | 0.08 | 1.72 | 0.79 | 0.25 | 0.007 | 0.005 | 0.01 | | | |
| 7 | 0.22 | 0.84 | 0.10 | 1.90 | 1.07 | 0.33 | 0.010 | 0.022 | 0.05 | 0.1 | +0.6 | CaMnSi |
| | 0.22 | 0.92 | 0.31 | 1.87 | 1.00 | 0.32 | 0.010 | 0.006 | 0.03 | | 0.3 | plus Ca |

* Before addition.

** After addition and taken in the ladle.

TABLE 2 — INCREASED ALUMINUM EFFECT

| Heat No. | Per Cent C | Mn | Si | Ni | Cr | Mo | P | S | Al | Al (added) | Wt. of Addition/Wt. of Charge, % | Type of Addition |
|---|---|---|---|---|---|---|---|---|---|---|---|---|
| 8 | 0.34 (B.A.) * | 0.38 | 0.15 | 1.90 | 5.05 | 0.27 | 0.008 | 0.020 | 0.13 | 0.15 | 2 | CaMnSi |
| | 0.34 (A.A.) ** | 0.84 | 1.70 | 1.85 | 4.78 | 0.25 | 0.008 | 0.002 | 0.05 | | | |
| 9 | 0.47 | 0.68 | 0.16 | 1.90 | 0.76 | 0.25 | 0.008 | 0.006 | 0.16 | 0.15 | 0.6 | Ca |
| | 0.45 | 0.69 | 0.21 | 1.80 | 0.78 | 0.26 | 0.007 | 0.003 | 0.05 | | | |
| 10 | 0.24 | 0.75 | 0.05 | 1.91 | 0.92 | 0.27 | 0.006 | 0.019 | 0.11 | 0.15 | +0.3 | CaMnSi |
| | 0.24 | 0.80 | 0.33 | 1.93 | 0.88 | 0.27 | 0.008 | 0.003 | 0.06 | | 0.3 | Ca |
| 11 | 0.69 | 0.35 | 0.09 | 1.75 | 0.85 | 0.25 | 0.009 | 0.005 | 0.10 | 0.15 | 2 | CaMnSi |
| | 0.67 | 0.87 | 1.59 | 1.63 | 0.83 | 0.23 | 0.009 | 0.002 | 0.07 | | | |
| 12 | 0.40 | 0.45 | 0.03 | 1.85 | 0.87 | 0.26 | 0.008 | 0.022 | 0.07 | 0.15 | 2 | CaMnSi |
| | 0.40 | 0.72 | 0.93 | 1.83 | 0.87 | 0.25 | 0.009 | 0.003 | 0.02 | | | |
| 13 | 0.39 | 1.00 | 0.06 | 1.30 | — | 0.29 | 0.009 | 0.023 | 0.11 | 0.15 | 2 | CaMnSi |
| | 0.40 | 1.29 | 1.12 | 1.26 | — | 0.28 | 0.009 | 0.003 | 0.09 | | | |

* Before addition.
** After addition and taken in the ladle.

TABLE 3 — TYPICAL SLAG ANALYSIS TAKEN BEFORE AND AFTER THE CaMnSi ADDITION

| Heat | Per Cent CaO | MgO | MnO | FeO | $SiO_2$ | $P_2O_5$ | $Al_2O_3$ | S |
|---|---|---|---|---|---|---|---|---|
| A | 80.1* | 1.1 | 0.4 | 6.1 | 4.1 | 0.07 | 2.9 | 0.414 |
| | 73.6** | 1.2 | 0.4 | 9.7 | 5.3 | 0.10 | 2.8 | 0.430 |
| B | 76.5 | 0.2 | 0.1 | 4.4 | 3.1 | 0.04 | 2.0 | 0.15 |
| | 73.9 | 0.2 | 0.4 | 11.0 | 4.7 | 0.14 | 2.9 | 0.29 |
| C | 74.3 | 0.2 | 0.2 | 3.2 | 2.0 | 0.03 | 1.5 | 0.125 |
| | 69.5 | 0.2 | 0.5 | 13.2 | 7.5 | 0.09 | 2.0 | 0.210 |

* Before the CaMnSi addition.
** After the CaMnSi addition.

TABLE 4 — INCREASED LIME ADDITIONS EFFECT

| Heat No. | Per Cent C | Mn | Si | Ni | Cr | Mo | P | S | Al | Lime Added to Bath Surface, % |
|---|---|---|---|---|---|---|---|---|---|---|
| 14 | 0.37 (B.A.) * | 0.34 | 0.10 | — | — | 0.26 | 0.010 | 0.024 | 0.11 | 0 |
| | 0.36 (A.A.) ** | 0.62 | 1.08 | 1.75 | 0.96 | 0.28 | 0.008 | 0.017 | 0.06 | |
| 15 | 0.37 | 0.35 | 0.10 | 1.76 | 0.96 | 0.28 | 0.008 | 0.024 | 0.11 | 1.6 |
| | 0.37 | 0.67 | 1.38 | 1.76 | 0.96 | 0.28 | 0.008 | 0.006 | 0.05 | |
| 16 | 0.32 | 0.30 | 0.10 | 1.76 | 0.93 | 0.26 | 0.010 | 0.019 | 0.12 | 2.0 |
| | 0.32 | 0.59 | 1.20 | 1.74 | 0.90 | 0.25 | 0.001 | 0.003 | 0.07 | |
| 17 | 0.36 | 0.41 | 0.19 | 1.86 | 0.83 | 0.25 | 0.009 | 0.022 | 0.11 | 2.5 |
| | 0.36 | 0.81 | 1.44 | 1.74 | 0.87 | 0.25 | 0.009 | 0.003 | 0.05 | |
| 18 | 0.36 | 0.33 | 0.18 | 1.88 | 0.93 | 0.26 | 0.008 | 0.024 | 0.10 | 3.0 |
| | 0.35 | 0.72 | 1.23 | 1.78 | 0.92 | 0.25 | 0.008 | 0.004 | 0.04 | |
| 19 | 0.37 | 0.33 | 0.12 | 1.93 | 0.98 | 0.27 | 0.007 | 0.024 | 0.11 | 3.2 |
| | 0.38 | 0.62 | 1.19 | 1.78 | 0.96 | 0.29 | 0.009 | 0.004 | | |
| 20 | 0.39 | 0.39 | 0.12 | 1.81 | 0.98 | 0.25 | 0.007 | 0.025 | 0.10 | 3.5 |
| | 0.39 | 0.72 | 1.08 | 1.75 | 0.98 | 0.24 | 0.009 | 0.003 | 0.03 | |
| 21 | 0.33 | 0.35 | 0.09 | 1.86 | 0.96 | 0.25 | 0.007 | 0.021 | 0.12 | 4.0 |
| | 0.32 | 0.74 | 1.22 | 1.79 | 0.94 | 0.26 | 0.009 | 0.003 | 0.05 | |
| 22 | 0.36 | 0.38 | 0.08 | 1.81 | 0.85 | 0.25 | 0.007 | 0.022 | 0.10 | 5.0 |
| | 0.35 | 0.67 | 0.99 | 1.81 | 0.87 | 0.24 | 0.008 | 0.003 | 0.03 | |
| 23 | 0.38 | 0.40 | 0.12 | 1.81 | 0.99 | 0.26 | — | 0.025 | 0.11 | 6.0 |
| | 0.36 | 0.68 | 1.16 | 1.78 | 0.99 | 0.25 | — | 0.001, 0.002 | 0.05 | |

(The heats were 300 lb melts in which 0.15% Al + 2% CaMnSi was added)

* Before addition.
** After addition and taken in the ladle.

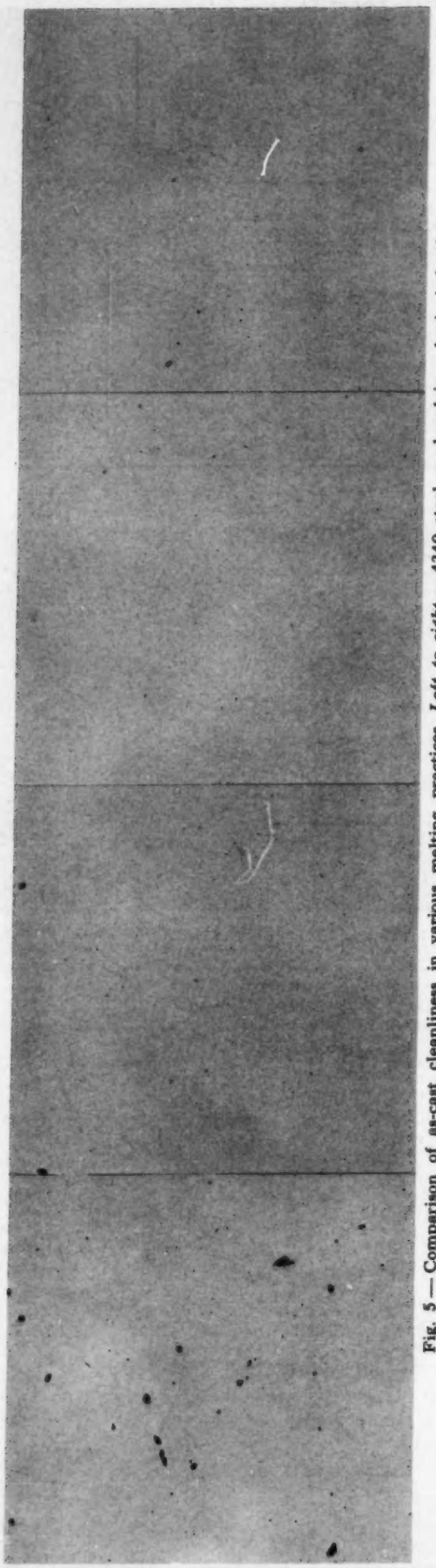


Fig. 5 — Comparison of as-cast cleanliness in various melting practices. *Left to right* — 4340 steel produced in a basic induction furnace, 4340 steel produced in a basic arc furnace, 4340 (1.5 per cent Si) steel produced in a basic induction furnace using the desulfurizing method of this investigation and 4340 steel produced in a basic induction furnace under vacuum. Unetched. 100 ×.

In addition to using a sufficient Al addition, and the proper technique for adding Ca or CaMnSi, a sufficient quantity of lime must be added to the bath surface before effective desulfurization can be obtained. As Table 4 indicates, for identical practices except for the lime additions, insignificant sulfur reduction was effected when lime was not available at the metal surface and the melt was left completely exposed to the atmosphere. But as the lime addition was increased, increased desulfurization was possible such that at a 6 lb/100 lb charge, desulfurization to levels as low as 0.001/0.002 per cent was obtained.

It is noteworthy to mention the fact that not only did this procedure produce steel with extremely low sulfur residuals, but the cleanliness of the steel was improved considerably and can be compared to vacuum induction melted steel (Fig. 5). The inclusions in these micrographs were not identified, but it is readily seen that the type of inclusions resulting from the desulfurization practice were of the more desirable Type 1, while the arc and the normal induction furnace melts contained considerable amounts of undesirable Type 3 inclusion.

When the desulfurizing procedure developed in this paper was tested on 1000 lb and 2400 lb melts similar results were obtained.

## SUMMARY AND CONCLUSION

Sulfur content of a medium carbon, medium alloy steel can be quickly, effectively and consistently reduced to 0.002/0.003 per cent whenever this procedure is followed:

1. Melt down in a basic lined furnace.
2. Remove melt down slag.
3. Deoxidize the metal bath with 0.15 per cent Al (more recent investigations indicate that greater quantities of Al may be necessary in low carbon melts).
4. Add lime in minimum quantities of 2 lb/100 lb charge (6 lb/100 lb charge is required if 0.001/0.002 per cent sulfur is desired).
5. At a sufficiently high enough temperature to prevent freezeup, immerse 2 lb/100 lb of charge of CaMnSi into the steel bath.
6. Tap at the desired temperature.

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# SAMPLING OF SAND

AFS Grading, Fineness and Distribution Committee (8-F) Report


## ABSTRACT

**The importance of sampling and controlling raw materials is increasing as the foundries modernize. Due to frequent requests for more information on sampling, Committee 8-F of AFS Sand Division and the Grading and Fineness Committee of the N.I.S.A. conducted a survey of sampling techniques and the probable error that can be expected in the sieve tests.**

**The survey showed that there is no universal, simple method of sampling. Each laboratory, foundry or mining operation has its own special technique or method which is satisfactory for its own conditions but not satisfactory for everyone. Nonmetallics will segregate according to grain distribution, density, moisture content and method of loading. The most uniform sample can be taken while the material is being loaded for shipment.**


## INTRODUCTION

In the casting industry, as well as in other industries, trained technicians run the tests, but untrained men from the unloading crew often get the samples of raw materials. Sampling is equally as important as the testing, and every precaution should be taken to obtain samples that show the true nature and condition of the materials they represent. Test results are valuable only when the tests are performed on a sample that is representative. Sampling and testing require practical experience and thought, to be carried out efficiently.

Samples are taken for two purposes:

1. To represent as nearly as possible an average of the materials sampled and to know the variation that is normally expected.
2. To subject materials to additional tests for controlling or improving methods and products.

Often a sample and a specimen are considered to be the same, but this is an error. A representative sample is a definite quantitative representative amount, containing the same proportions of each essential constituent. A specimen is a portion that contains a noticeable amount of each essential constituent of the mass from which it is taken, but it is intended only as a qualitative representative amount. The eye cannot accurately detect various amounts of materials so it can not distinguish unlikeness. A test is the only way to judge a shipment.

Grading, Fineness and Distribution Committee (8-F): T. W. Seaton, *Chairman*, J. G. Smillie, *Vice Chairman*, R. W. Bennett, W. D. Chadwick, C. F. Daly, L. F. Dennie, R. Dietert, G. DiSylvestro, H. S. Fagan, F. P. Goettman, W. L. Hemingway, W. Mahlig, L. Marinelli, R. E. Morey, G. Nestor, J. A. Schumann, E. W. Smith, N. J. Stickney, W. C. Swanson, T. A. Tarquinio, L. E. Taylor, S. Walker and S. A. Wick.

## BULK SAMPLING OF DAMP MATERIALS

Sampling of materials is an important phase of laboratory testing. There are various methods and procedures for getting samples and each one should have some good points. It is most important that the same method of procedure be followed for all samples. Sampling of damp sand and/or a clay bearing sand is not as difficult as a clay free, dry sand. If difficulties are found in sampling sand due to excess segregation, it is suggested that the shippers sample and test data be used.

There are many ways of taking samples from various types of equipment. It is recommended that open cars, covered hoppers, and trucks be sampled as shown in Fig. 1.

At least ten equal increments of at least one lb each shall be taken from various equally spaced places on the car or truck, as shown in Fig. 1. An excavation of about one ft in depth shall be made at each sampling place or point, and the sample shall be taken from the bottom of the excavation.

Commercially available samplers are designed for this purpose. Some of the types available are shown in Fig. 3.

A box car of sand is usually loaded on both ends leaving the doorways clear. Because of the motion of the car enroute the material is usually flat or even throughout the car. When this happens the material in the car is sampled with the same procedure as shown in Fig. 1. If the material is still in a heap or piled on both sides of the car, at least six places or points shall be sampled from each end; about one lb from each location (Fig. 2).

## BAG SAMPLING

A sample from a bag shipment shall be taken with as much accuracy as possible. The number of bags to be sampled shall be the cube root of the number of bags in the shipment. At least one lb shall be taken

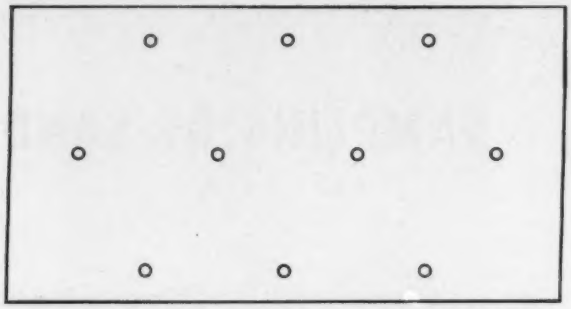


Fig. 1 — Open cars, covered hoppers and trucks of sand recommended sampling method. Ten equal increments are taken, about one ft deep.

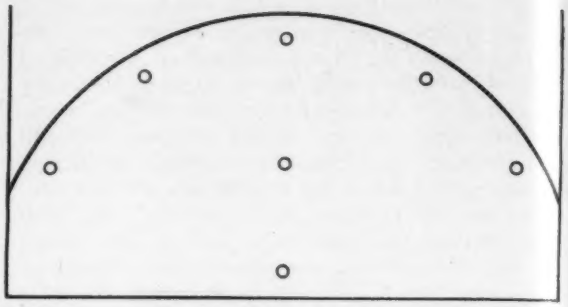


Fig. 2 — If the material is in a heap or piled on both sides of the car, this method is recommended. At least six points are sampled; seven points are shown.

from each bag in such a way that part of the sample be taken below the surface.

If the combined samples are large, they can be reduced in size by the quartering method. Further reduction in size is done with a sand splitter. A sample should always be split to the approximate desired weight. Altering the weight by means of adding or reducing small amounts shall not be done.

Some of the samplers that can be purchased or made are shown in Fig. 3.

## SAMPLING AND REPRODUCIBILITY

The problem of sampling damp material is not as great as sampling dry material. One method is to use a 2 in. diameter x 6 ft long tube, thrusting it into the sand in a perpendicular manner and taking the 2 in. core sample from the same locations as shown in Fig. 1.

In order to learn the probable error that can be expected in the sieve test, 20 damp samples and 20 dry samples were sent to 20 laboratories for testing. Each laboratory received a different bag containing approximately 2 lb of damp silica sand. These bags were marked 1, 2, 3, etc. Each laboratory also received a bottle containing approximately 50 grams of dry silica sand. These bottles were marked A-1, A-2, A-3, A-4, B-1, B-2, etc. Bottle *A* went to four laboratories, bottle *B* went to four laboratories, etc.

The silica sand selected was washed and graded, and in the damp condition. The sand was reduced in size with a sand splitter having nine ½-in. slots on each side, Fig. 4.

Approximately 80 lb of damp silica sand was split five times until there were 32 samples of approximately 2 lb each. These samples were put into bags and sent to 20 committee members. The members were instructed to dry sample in an oven, reduce sand to approximately 50 grams, and run sieve test.

Two damp sand samples from above group were mixed together and dried in an oven at 220 F for 2 hr. The dry sand was split five times with a sand splitter, until each sample was approximately 50 grams. Five 50 gram samples were selected and put into small glass bottles. Each bottle was sent to four laboratories. The personnel of the committee was instructed to put the contents of the bottle into their sieves without trying to reduce the sample to a certain weight. After sieving and weighing, the weights were computed on the basis of 100 grams. This procedure eliminated sampling error.

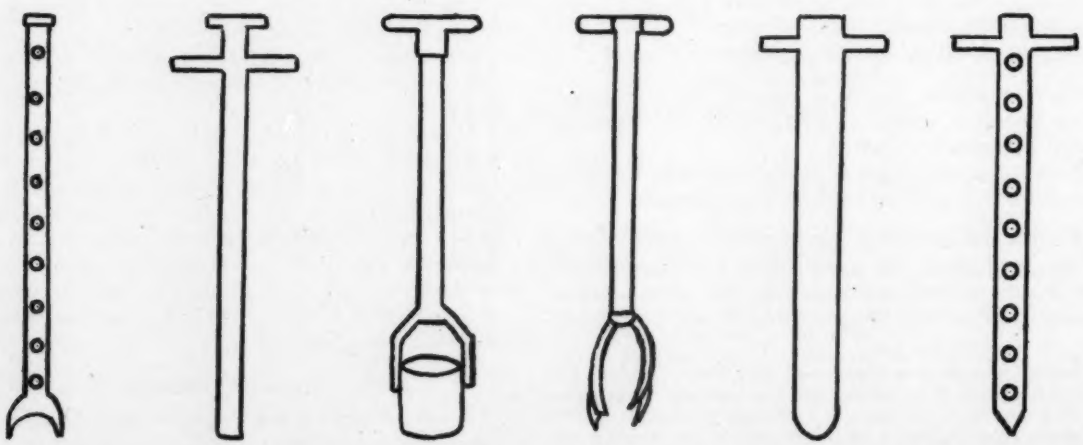


Fig. 3 — Some of the different shaped sand samplers which are available.

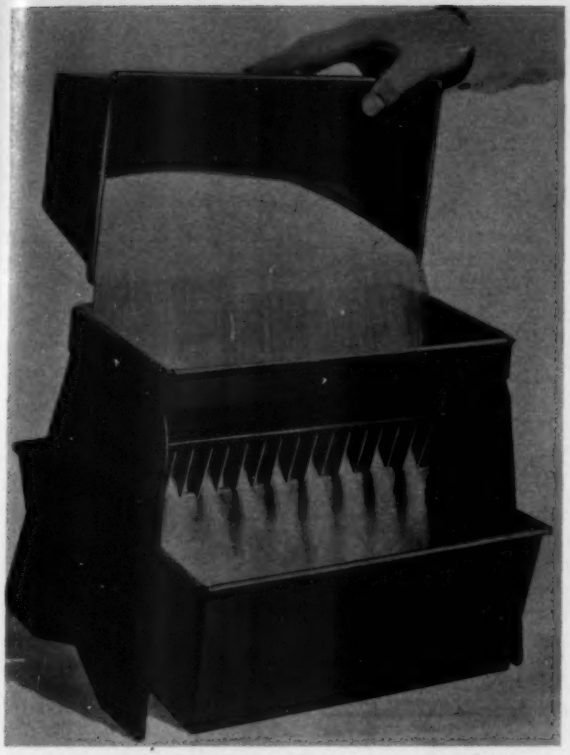


Fig. 4 — Sand was reduced in size with a sand splitter having nine ½-in. slots on each side.

## DISCUSSION OF THE DATA

Charts 1, 2 and 3 show the sieve test data collected from the 20 participating laboratories.

The data listed under columns numbered 1, 2, 3, 4, etc., represents data on the 2 lb samples.

The data listed under 1-A, 2-A, 3-A, 4-A, 1-B, 2-B, 3-B, etc., are data collected on the samples of approximately 50 grams. All of the samples showing the letter *A* after them represent tests by four laboratories on the same 50 gram sample. The same is true of the B's, C's, D's and G's. In other words, four different laboratories ran the same 50 gram sample.

The type of vibrating equipment used by the various laboratories is shown under their reported data. Sieving time is indicated, the type of sieves on which the tests were run is shown and the method of preparing the smaller sample from the 2 lb sample is shown.

Chart 4 is a summation of the data, and is broken down to show the range of the tests and the ranges with respect to the type of sifting device employed.

Tables *A* and *B* show the data recorded in charts 1, 2 and 3, as a cumulative analysis for all of the tests. The high and low ranges for the damp samples (2 lb) are shown at the bottom of Table *A*. The high and low ranges for the dry samples (50 gram) are shown at the bottom of Table *B*. The averages for both dry and damp samples are also shown.

The cumulative data was then plotted on semi-log graph paper in order to obtain additional information. The cumulative plot enables the "sorting coefficient" and skewness of a sand to be determined and compared with other sands.

In order to calculate the sorting coefficient and skewness of a sand, the 25 per cent quartile ($Q_1$), Median (M) and 75 per cent quartile ($Q_3$) must be determined.

Figure 5 shows how these figures are obtained. It shows the cumulative plot of the average sieve analysis of the 20 damp samples. At the bottom are listed the U.S. Series Sieves used to run the sieve analysis. They are plotted on the basis of sieve opening size in microns. At the top of the graph is shown the corresponding micron sizes. The vertical plot is the cumulative per cent from 0 to 100 per cent.

The 25 per cent Quartile ($Q_1$) is that point at which the 25 per cent retained line intersects the curve and is then dropped vertically to meet the base and the value is recorded in microns. The median (M) is obtained by extending the 50 per cent retained line in the same manner. The 75 per cent Quartile ($Q_3$) is obtained by extending the 75 per cent line.

The sorting coefficient ($S_o$) is a measure of the distribution, and is obtained from the formula:

$$S_o = \sqrt{\frac{Q_1}{Q_3}}$$

Where particles are all of the same size, the sorting coefficient would be 1.0. As distribution broad-

CHART 1 — SIEVE TESTS ON DAMP AND DRY SAMPLES

| U.S. No. | Microns | 1 | 1A | 2 | 2A | 3 | 3A | 4 | 4A | 5 | 5B | 6 | 6B | 7 | 7B |
|---|---|---|---|---|---|---|---|---|---|---|---|---|---|---|---|
| 6 | 3360 | | | | | | | | | | | | | | |
| 12 | 1680 | | | | | | | | | | | | | | |
| 20 | 840 | | | | | | | | | 0.02 | 0.02 | | | | |
| 30 | 590 | | | 0.2 | 0.2 | 0.1 | 0.2 | 0.1 | | 0.05 | 0.06 | Tr | Tr | | 0.2 |
| 40 | 420 | 2.0 | 1.7 | 2.0 | 1.9 | 1.7 | 1.8 | 1.9 | 2.0 | 2.02 | 1.84 | 2.4 | 1.9 | 1.9 | 1.6 |
| 50 | 297 | 14.0 | 13.4 | 12.6 | 12.8 | 12.7 | 13.2 | 13.8 | 15.3 | 12.58 | 12.25 | 13.0 | 13.1 | 13.5 | 12.6 |
| 70 | 210 | 37.0 | 34.9 | 32.3 | 32.4 | 31.8 | 31.2 | 35.0 | 35.4 | 33.14 | 31.73 | 35.8 | 36.5 | 38.7 | 35.6 |
| 100 | 149 | 37.0 | 41.3 | 41.7 | 41.1 | 41.3 | 41.0 | 40.6 | 36.4 | 40.55 | 41.51 | 37.2 | 36.8 | 36.6 | 40.0 |
| 140 | 105 | 8.2 | 7.2 | 10.2 | 10.5 | 11.1 | 11.1 | 7.2 | 9.4 | 10.46 | 11.05 | 10.0 | 10.2 | 8.4 | 9.0 |
| 200 | 74 | 1.2 | 1.2 | 1.0 | 1.0 | 1.1 | 1.4 | 1.0 | 1.0 | 1.00 | 1.21 | 1.4 | 1.1 | 0.7 | 0.8 |
| 270 | 53 | 0.1 | 0.2 | 0.1 | 0.2 | 0.1 | | 0.2 | 0.1 | 0.09 | 0.13 | 0.1 | 0.2 | | |
| Pan | | 0.1 | | | | | | | | 0.05 | 0.10 | 0.1 | 0.2 | 0.2 | 0.2 |
| Vibrator | | Rotate-Tap | | Rotate-Tap | | Rotate-Tap | | Rotate-Tap | | Rotate-Tap | | Rotate-Tap | | Rotate-Tap | |
| Time | | 15 min | | — | | 15 min | | 15 min | | — | | 15 min | | 15 Min | |
| Sieves | | Certified | | Standard | | Standard | | Certified | | Matched | | — | | — | |
| Splitter | | Tyler | | Sand | | Sand | | Tyler | | Tyler | | Mechanical | | — | |

## CHART 2 — SIEVE TESTS ON DAMP AND DRY SAMPLES

| U.S. No. | Microns | 8 | 8B | 9 | 9D | 10 | 10D | 11 | 11D | 12 | 12D | 13 | 13G | 14 | 14G |
|---|---|---|---|---|---|---|---|---|---|---|---|---|---|---|---|
| 6 | 3360 | | | | | | | | | | | | | | |
| 12 | 1680 | | | | | | | | | | | | | | |
| 20 | 840 | | | | | | | | | | | 0.02 | | | |
| 30 | 590 | | 0.2 | 0.08 | 0.04 | Tr | Tr | 0.10 | 0.05 | 0.04 | — | 0.03 | 0.06 | 0.2 | 0.2 |
| 40 | 420 | 2.0 | 1.3 | 2.32 | 2.20 | 2.2 | 2.2 | 1.75 | 1.80 | 2.22 | — | 2.63 | 2.50 | 2.0 | 2.9 |
| 50 | 297 | 13.0 | 12.6 | 15.26 | 15.22 | 12.2 | 12.2 | 13.55 | 13.65 | 14.05 | — | 14.75 | 14.34 | 15.0 | 16.5 |
| 70 | 210 | 38.2 | 36.1 | 34.78 | 34.33 | 35.0 | 34.8 | 33.60 | 33.70 | 41.65 | — | 44.32 | 43.10 | 38.4 | 38.1 |
| 100 | 149 | 38.2 | 40.7 | 34.38 | 38.32 | 40.8 | 39.6 | 37.60 | 37.80 | 32.24 | — | 31.45 | 32.80 | 34.4 | 33.9 |
| 140 | 105 | 7.2 | 7.6 | 7.64 | 8.02 | 8.8 | 8.2 | 12.00 | 11.30 | 8.60 | — | 5.60 | 6.00 | 8.8 | 7.5 |
| 200 | 74 | 1.0 | 1.3 | 1.30 | 1.56 | 0.8 | 1.6 | 1.10 | 1.30 | 1.12 | — | 1.00 | 1.00 | 1.0 | 0.7 |
| 270 | 53 | | 0.2 | 0.16 | 0.22 | 0.1 | 0.1 | 0.10 | 0.20 | 0.04 | — | 0.18 | 0.06 | 0.1 | 0.1 |
| Pan | | | | 0.08 | 0.08 | 0.1 | 0.1 | 0.20 | 0.20 | Tr | — | 0.02 | 0.14 | 0.1 | 0.1 |
| Vibrator | | Rotate-Tap | | Rotate-Tap | | Rotate-Tap | | Rotate-Tap | | Coombs | | Coombs (No tappers) | | Coombs | |
| Time | | — | | 15 min | | 15 min | | — | | 15 min | | 15 min | | — | |
| Sieves | | — | | Standard | | Standard | | Standard | | Standard | | Standard | | Standard | |
| Splitter | | Jones | | Quartered | | Simple | | Tyler | | Rolled & Quartered | | — | | Quartered | |

## CHART 3 — SIEVE TESTS ON DAMP AND DRY SAMPLES

| U.S. No. | Microns | 15 | 15G | 16 | 16G | 17 | 17C | 18 | 18C | 19 | 19C | 20 | 20C |
|---|---|---|---|---|---|---|---|---|---|---|---|---|---|
| 6 | 3360 | | | | | | | | | | | | |
| 12 | 1680 | | | | | | | | | | | | |
| 20 | 840 | 0.01 | | | | | | | | | | | |
| 30 | 590 | 0.06 | 0.042 | 0.1 | 0.1 | | | 0.2 | | Tr | | 0.1 | Tr |
| 40 | 420 | 2.00 | 2.184 | 1.6 | 1.9 | 2.2 | 2.0 | 1.6 | 2.0 | 1.8 | 1.9 | 1.8 | 1.9 |
| 50 | 297 | 13.26 | 14.018 | 12.0 | 13.3 | 13.6 | 11.8 | 13.4 | 14.0 | 11.0 | 11.5 | 10.8 | 12.2 |
| 70 | 210 | 37.09 | 36.419 | 38.1 | 39.5 | 37.4 | 37.2 | 32.4 | 32.6 | 31.7 | 34.5 | 37.9 | 37.3 |
| 100 | 149 | 37.91 | 37.613 | 38.8 | 37.2 | 38.6 | 39.6 | 41.6 | 41.2 | 43.8 | 42.0 | 38.8 | 38.8 |
| 140 | 105 | 8.86 | 8.812 | 8.3 | 6.9 | 7.2 | 8.0 | 9.6 | 9.2 | 10.0 | 8.7 | 9.2 | 8.6 |
| 200 | 74 | 0.71 | 0.761 | 0.8 | 0.8 | 1.0 | 1.2 | 0.8 | 0.8 | 1.4 | 1.2 | 0.9 | 0.7 |
| 270 | 53 | 0.05 | 0.029 | 0.2 | 0.2 | | 0.2 | 0.2 | 0.2 | 0.2 | 0.2 | 0.2 | 0.5 |
| Pan | | 0.05 | 0.122 | 0.1 | 0.1 | | Tr | 0.2 | | Tr | | 0.1 | — |
| Vibrator | | Coombs (Tappers on all sieves) | | — | — | Rotate-Tap | | Rotate-Tap | | Rotate-Tap | | Rotate-Tap | Rotate-Tap |
| Time | | 15 min | | — | — | — | | — | | 15 min | | 15 min | 15 min |
| Sieves | | Standard | | Master | | — | | Standard | | Standard | | — | — |
| Splitter | | | | | | Jones | | Quartered | | Jones | | | Quartered |

(Data for Fig. 5)
Sieve Test

| U.S. Series Sieve No. | Microns | Cumulative % |
|---|---|---|
| 6 | 3360 | |
| 12 | 1680 | |
| 20 | 840 | |
| 30 | 590 | |
| 40 | 420 | 2.1 |
| 50 | 297 | 15.3 |
| 70 | 210 | 51.6 |
| 100 | 149 | 39.9 |
| 140 | 105 | 93.3 |
| 200 | 74 | 99.8 |
| 270 | 53 | 99.9 |
| Pan | — | |

| $Q_1$ | M | $Q_3$ | $S_o$ | $S_k$ |
|---|---|---|---|---|
| 261 | 210 | 170 | 1.241 | 1.006 |

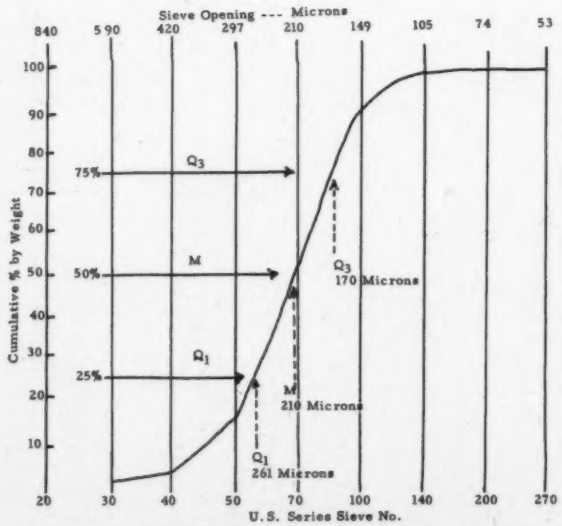



Fig. 5 — Cumulative data of average sieve analysis of 20 damp samples.

TABLE A — CUMULATIVE RESULTS — DAMP SAMPLES

| Sample | 6 | 12 | 20 | 30 | 40 | 50 | 70 | 100 | 140 | 200 | 270 | Pan |
|---|---|---|---|---|---|---|---|---|---|---|---|---|
| 1 | | | | | 2.0 | 16.0 | 53.0 | 90.4 | 98.6 | 99.8 | 99.9 | 100.0 |
| 2 | | | | 0.2 | 2.2 | 14.8 | 47.1 | 88.8 | 99.0 | 100.0 | 100.0 | 100.0 |
| 3 | | | | 0.1 | 1.8 | 14.5 | 46.3 | 87.6 | 98.7 | 99.8 | 99.9 | 100.0 |
| 4 | | | | 0.1 | 2.0 | 15.8 | 50.8 | 91.4 | 98.6 | 99.6 | 99.8 | 100.0 |
| Avg. 1-4 | | | | 0.1 | 2.0 | 15.3 | 49.3 | 89.6 | 98.7 | 99.8 | 99.9 | 100.0 |
| 5 | | | | 0.1 | 2.1 | 14.7 | 47.8 | 88.4 | 98.8 | 99.8 | 99.9 | 100.0 |
| 6 | | | | | 2.4 | 15.4 | 51.2 | 88.4 | 98.4 | 99.8 | 99.9 | 100.0 |
| 7 | | | | | 1.9 | 15.4 | 54.1 | 90.7 | 99.1 | 99.8 | 99.8 | 100.0 |
| 8 | | | | | 2.0 | 15.0 | 53.2 | 91.4 | 98.6 | 99.6 | 99.8 | 100.0 |
| Avg. 5-8 | | | | | 2.1 | 15.1 | 51.6 | 89.7 | 98.7 | 99.8 | 99.9 | 100.0 |
| 9 | | | | 0.1 | 2.5 | 18.4 | 54.6 | 90.5 | 99.5 | 99.8 | 99.9 | 100.0 |
| 10 | | | | | 2.2 | 14.4 | 49.4 | 90.2 | 99.0 | 99.8 | 99.9 | 100.0 |
| 11 | | | | 0.1 | 1.8 | 15.4 | 49.0 | 86.6 | 98.6 | 99.7 | 99.8 | 100.0 |
| 12 | | | | | 2.3 | 16.3 | 58.0 | 90.2 | 98.8 | 99.9 | 100.0 | 100.0 |
| Avg. 9-12 | | | | | 2.2 | 16.1 | 52.7 | 89.4 | 99.0 | 99.8 | 99.9 | 100.0 |
| 13 | | | | 0.1 | 2.7 | 17.4 | 61.8 | 93.2 | 98.8 | 99.8 | 100.0 | 100.0 |
| 14 | | | | 0.2 | 2.2 | 17.2 | 55.6 | 90.0 | 98.8 | 99.8 | 99.9 | 100.0 |
| 15 | | | | 0.1 | 2.1 | 15.3 | 52.4 | 90.3 | 99.2 | 99.9 | 99.9 | 100.0 |
| 16 | | | | 0.1 | 1.7 | 13.7 | 51.8 | 90.6 | 98.9 | 99.7 | 99.9 | 100.0 |
| Avg. 13-16 | | | | 0.1 | 2.2 | 15.9 | 55.4 | 91.0 | 98.9 | 99.8 | 99.9 | 100.0 |
| 17 | | | | | 2.2 | 15.8 | 53.2 | 91.8 | 99.0 | 100.0 | 100.0 | 100.0 |
| 18 | | | | 0.2 | 1.8 | 15.2 | 47.6 | 89.2 | 98.8 | 99.6 | 99.8 | 100.0 |
| 19 | | | | | 1.8 | 12.8 | 44.5 | 88.3 | 98.3 | 99.7 | 99.9 | 100.0 |
| 20 | | | | 0.1 | 1.9 | 12.7 | 50.6 | 89.4 | 98.8 | 99.7 | 99.9 | 100.0 |
| Avg. 17-20 | | | | 0.1 | 1.9 | 14.1 | 49.0 | 89.7 | 98.7 | 99.8 | 99.9 | 100.0 |
| Avg. of 20 samples | | | | | 2.1 | 15.3 | 51.6 | 89.9 | 98.8 | 99.8 | 99.9 | 100.0 |
| Range — High | | | | | 2.7 | 18.4 | 61.8 | 93.2 | 99.5 | 100.0 | 100.0 | 100.0 |
| Range — Low | | | | | 1.7 | 12.7 | 44.5 | 86.6 | 98.3 | 99.6 | 99.8 | 100.0 |

TABLE B — CUMULATIVE RESULTS — DRY SAMPLES

| Sample | 6 | 12 | 20 | 30 | 40 | 50 | 70 | 100 | 140 | 200 | 270 | Pan |
|---|---|---|---|---|---|---|---|---|---|---|---|---|
| 1A | | | | | 1.7 | 15.1 | 50.0 | 91.3 | 98.5 | 99.7 | 99.9 | 100.0 |
| 2A | | | | 0.2 | 2.1 | 14.9 | 47.3 | 88.4 | 98.9 | 99.9 | 100.1 | 100.1 |
| 3A | | | | 0.2 | 2.0 | 15.2 | 46.4 | 87.4 | 98.5 | 99.9 | 100.0 | 100.0 |
| 4A | | | | | 2.0 | 17.3 | 52.7 | 89.1 | 98.5 | 99.5 | 99.6 | 100.0 |
| Avg. 1A-4A | | | | | 1.9 | 15.6 | 49.1 | 89.0 | 98.6 | 99.8 | 99.9 | 100.0 |
| 5B | | | | 0.1 | 1.9 | 14.2 | 45.9 | 87.4 | 98.5 | 99.7 | 99.8 | 99.9 |
| 6B | | | | | 1.9 | 15.0 | 51.5 | 88.3 | 98.5 | 99.6 | 99.8 | 100.0 |
| 7B | | | | 0.2 | 1.8 | 14.4 | 50.0 | 90.0 | 99.0 | 99.8 | 99.8 | 100.0 |
| 8B | | | | 0.2 | 1.5 | 14.1 | 50.2 | 90.9 | 98.5 | 99.8 | 100.0 | 100.0 |
| Avg. 5B-8B | | | | 0.1 | 1.8 | 14.4 | 49.4 | 89.2 | 98.6 | 99.7 | 99.8 | 100.0 |
| 9D | | | | | 2.2 | 17.5 | 51.8 | 90.1 | 98.1 | 99.7 | 99.9 | 100.0 |
| 10D | | | | | 2.2 | 14.4 | 49.2 | 88.8 | 97.0 | 98.6 | 98.7 | 98.8 |
| 11D | | | | | 1.8 | 15.5 | 49.2 | 87.0 | 98.3 | 99.6 | 99.8 | 100.0 |
| 12D | | | | | | | | | | | | |
| Avg. 9D-11D | | | | | 2.1 | 15.8 | 50.1 | 88.6 | 97.8 | 99.3 | 99.5 | 99.6 |
| 13G | | | | 0.1 | 2.6 | 16.9 | 60.0 | 92.8 | 98.8 | 99.8 | 99.9 | 100.0 |
| 14G | | | | | 3.0 | 19.6 | 57.7 | 91.6 | 99.1 | 99.8 | 99.9 | 100.0 |
| 15G | | | | | 2.2 | 16.2 | 52.7 | 90.3 | 99.1 | 99.8 | 99.9 | 100.0 |
| 16G | | | | 0.1 | 2.0 | 15.3 | 54.8 | 92.0 | 98.9 | 99.7 | 99.9 | 100.0 |
| Avg. 13G-16G | | | | | 2.4 | 17.0 | 56.3 | 91.7 | 99.0 | 99.8 | 99.9 | 100.0 |
| 17C | | | | | 2.0 | 13.8 | 51.0 | 90.6 | 98.6 | 99.8 | 100.0 | 100.0 |
| 18C | | | | | 2.0 | 16.0 | 48.6 | 89.8 | 99.0 | 99.8 | 100.0 | 100.0 |
| 19C | | | | | 1.9 | 13.4 | 47.9 | 89.9 | 98.6 | 99.8 | 100.0 | 100.0 |
| 20C | | | | | 1.9 | 14.1 | 51.4 | 90.2 | 98.8 | 99.5 | 100.0 | 100.0 |
| Avg. 17C-20C | | | | | 2.0 | 14.3 | 49.7 | 90.1 | 98.8 | 99.7 | 100.0 | 100.0 |
| Avg. of 19 samples | | | | | 2.0 | 15.3 | 51.0 | 89.8 | 98.6 | 99.7 | 99.8 | 99.9 |
| Range — High | | | | | 3.0 | 19.6 | 60.0 | 92.8 | 99.1 | 99.9 | 100.1 | 100.1 |
| Range — Low | | | | | 1.5 | 13.4 | 45.9 | 87.0 | 97.0 | 98.6 | 98.7 | 98.8 |

CHART 4 — RESULTS OF SIEVE TESTS

| U.S. Series | Avg. 1 to 20 | Avg. all rotate-taps | Avg. all Coombs | Range all Tests | Range all rotate-taps | Range all Coombs |
|---|---|---|---|---|---|---|
| 20 | Tr | Tr | Tr | 0.0- 0.02 | 0.0- 0.02 | 0.0 - 0.02 |
| 30 | 0.1 | 0.1 | 0.1 | 0.0- 0.2 | 0.0- 0.2 | 0.03- 0.2 |
| 40 | 2.0 | 2.0 | 2.1 | 1.3- 2.9 | 1.3- 2.4 | 1.6 - 2.9 |
| 50 | 13.3 | 13.2 | 13.8 | 10.8-16.5 | 10.8-15.3 | 12.0 -16.5 |
| 70 | 36.5 | 35.6 | 39.5 | 31.2-44.3 | 31.2-41.7 | 36.4 -44.3 |
| 100 | 37.9 | 38.5 | 35.6 | 31.5-43.8 | 32.2-43.8 | 31.5 -38.8 |
| 140 | 8.8 | 9.1 | 7.9 | 5.6-12.0 | 7.2-12.0 | 5.6 - 8.9 |
| 200 | 1.0 | 1.0 | 0.9 | 0.7- 1.6 | 0.7- 1.6 | 0.7 - 1.0 |
| 270 | 0.1 | 0.1 | 0.1 | 0.0- 0.2 | 0.0- 0.2 | 0.03- 0.2 |
| Pan | 0.1 | Tr | Tr | 0.0- 0.2 | 0.0- 0.2 | 0.02- 0.1 |

TABLE C — GRAPHIC RESULTS FOR DAMP SAND

| Sample | $Q_1$ | M | $Q_3$ | $S_o$ | $S_k$ |
|---|---|---|---|---|---|
| 1 | 269 | 215 | 174 | 1.242 | 1.012 |
| 2 | 260 | 206 | 169 | 1.240 | 1.034 |
| 3 | 257 | 203 | 167 | 1.241 | 1.027 |
| 4 | 265 | 212 | 173 | 1.238 | 1.020 |
| Avg. 1-4 | 263 | 209 | 171 | 1.240 | 1.023 |
| 5 | 260 | 206 | 168 | 1.244 | 1.029 |
| 6 | 267 | 212 | 172 | 1.247 | 1.022 |
| 7 | 269 | 217 | 177 | 1.242 | 1.010 |
| 8 | 268 | 216 | 178 | 1.227 | 1.022 |
| Avg. 5-8 | 266 | 213 | 174 | 1.240 | 1.021 |
| 9 | 276 | 218 | 174 | 1.259 | 1.010 |
| 10 | 261 | 209 | 172 | 1.232 | 1.032 |
| 11 | 264 | 209 | 168 | 1.252 | 1.015 |
| 12 | 275 | 224 | 180 | 1.237 | 0.987 |
| Avg. 9-12 | 269 | 215 | 174 | 1.245 | 1.011 |
| 13 | 279 | 230 | 186 | 1.223 | 0.981 |
| 14 | 273 | 219 | 176 | 1.245 | 1.001 |
| 15 | 268 | 213 | 174 | 1.241 | 1.027 |
| 16 | 262 | 212 | 174 | 1.228 | 1.015 |
| Avg. 13-16 | 271 | 219 | 178 | 1.234 | 1.008 |
| 17 | 269 | 215 | 177 | 1.232 | 1.029 |
| 18 | 258 | 205 | 169 | 1.235 | 1.037 |
| 19 | 251 | 200 | 166 | 1.230 | 1.041 |
| 20 | 260 | 211 | 173 | 1.225 | 1.011 |
| Avg. 17-20 | 260 | 208 | 171 | 1.231 | 1.029 |
| Avg. 1-20 | 266 | 213 | 173 | 1.238 | 1.018 |
| Range—High | 279 | 230 | 186 | 1.259 | 1.041 |
| Range—Low | 251 | 200 | 166 | 1.223 | 0.981 |

ens, the sorting coefficient increases. Most sands will fall between 1.14 and 2.50.

Skewness coefficient ($S_k$) is a measure of the slope of a curve, and is calculated from the formula:

$$S_k = \frac{Q_1 \, Q_3}{M^2}$$

Tables *C* and *D* show the figures obtained from the graph and $S_o$ and $S_k$ for all the sands tested. The average values are shown and the high and low range is shown.

## CONCLUSIONS

The data obtained for this work were obtained under what would be considered ideal conditions. Samples were prepared with extreme care and sieve tests were run under the most desirable conditions.

These facts developed as a result of this study:

1. Exact duplication of sieve tests between different laboratories under the most closely controlled conditions is difficult, if not impossible.

TABLE D — GRAPHIC RESULTS FOR DRY SAND

| Sample | $Q_1$ | M | $Q_3$ | $S_o$ | $S_k$ |
|---|---|---|---|---|---|
| 1A | 262 | 210 | 177 | 1.218 | 1.050 |
| 2A | 257 | 205 | 171 | 1.227 | 1.045 |
| 3A | 255 | 203 | 167 | 1.235 | 1.033 |
| 4A | 268 | 214 | 174 | 1.241 | 1.018 |
| 5B | 260 | 203 | 166 | 1.251 | 1.045 |
| 6B | 268 | 212 | 170 | 1.255 | 1.014 |
| 7B | 265 | 210 | 170 | 1.249 | 1.020 |
| 8B | 263 | 210 | 170 | 1.238 | 1.014 |
| 9D | 270 | 212 | 173 | 1.250 | 1.038 |
| 10D | 261 | 209 | 171 | 1.235 | 1.021 |
| 11D | 262 | 208 | 168 | 1.250 | 1.017 |
| 12D | | | | | |
| 13G | 274 | 227 | 184 | 1.221 | 0.978 |
| 14G | 278 | 223 | 180 | 1.242 | 1.005 |
| 15G | 268 | 214 | 174 | 1.241 | 1.017 |
| 16G | 268 | 218 | 179 | 1.223 | 1.009 |
| 17C | 261 | 212 | 174 | 1.225 | 1.010 |
| 18C | 262 | 207 | 170 | 1.241 | 1.039 |
| 19C | 257 | 206 | 170 | 1.230 | 1.028 |
| 20C | 261 | 212 | 174 | 1.225 | 1.009 |
| Total Avg. | 264 | 211 | 172 | 1.237 | 1.022 |
| Range—High | 278 | 227 | 184 | 1.255 | 1.050 |
| Range—Low | 255 | 203 | 166 | 1.218 | 0.978 |

2. The inability to obtain duplication of tests must of necessity be taken into consideration whenever specifications for sand are discussed.
3. Extreme care in sampling is a must. Careless sampling can only result in misleading and erroneous information.
4. Sieving efficiency for various types of shakers is dependent on time. The "end-point" of sieving or the reproducibility varies with each type of shaker.

These recommendations are made by this committee as a result of this study:

1. Samples should be obtained by the methods presented to insure obtaining representative samples.
2. A sample splitter should be used to split the sample down to the approximate test sample size. Never add to or remove from the split sample to obtain a sample of a definite weight. This will destroy the value of the sample.
3. It is necessary to weigh only to the nearest tenth of a gram.
4. If an analysis does not meet expected results or specifications, obtain another sample. Do not retest the same sample.
5. Better accuracy of weighing can be obtained by cumulative weighing than by weighing material retained on individual sieves separately.
6. The variations in sieve tests shown in this report are what can be normally expected under the most closely controlled conditions.
7. It can reasonably be expected that there will be differences in tests run by producers and consumers on shipments of raw materials. These variations should be determined and used as a basis for setting up specifications.
8. It is suggested that consumers accept the samples obtained at point of production as being representative of the shipment and where possible, accept suppliers test data.

# CORE PRODUCTION IN HOT CORE BOXES

by Philippe Jasson


## ABSTRACT

**It is now over three years that core production in hot boxes has been in operation at the author's company for mass production of automobile parts. Different techniques to those at present being developed in the United States, were elaborated. In this report, the company's realizations in this field are described.**


## INTRODUCTION

As early as the beginning of 1956, the author's company was called upon to define what is now called core production in hot boxes, and it is of interest to recall the reasons which brought the company to elaborate upon this process.

The foundryman's ambition is to improve the quality of his moldings while reducing his costs. In automobile foundry, owing to the size of the cored castings, the cost of producing the cores has a considerable influence on the cost price of the molded part. However, it can easily be proved that, in the standard core production, the cost price of a core depends essentially on the cost of man power employed in accessory operations such as – preparation of wires, fitting of dryer, loading and unloading drying oven, finishing and grinding the cores to size and checking and measuring before molding again.

These operations are rendered necessary owing to distortions which occur while stripping the core or during curing. Hardening the core in the box therefore appears to be the rational method of obtaining core precision and to reduce the cost thereof, and this by the suppression of the accessory operations. It is to be noted, however, that all the new processes of core production, which have appeared in recent years, apply the method of hardening the sand in contact with the core box. Whether it be shell molding, silicate molding carbon dioxide hardened, air set sand molding, and even some investment molding techniques, the mass of sand is hardened before being separated from the pattern.

However, one must agree that each of these processes, considered separately, has its advantages and its drawbacks, and the company has endeavored to create a process which would be the synthesis of modern tendencies, while maintaining, however, a certain number of imperative conditions which are well founded, as proved by traditional methods. In particular, the company has tried to retain:

1. The production of solid cores, the body of which is capable of supporting the outer parts bearing the stress of the metal when casting. In fact, thin walled cores or parts of molds must generally be reinforced, excepting in the case of small castings, by sand packing, shot or gravel, in order to restore assemblies capable of resisting without distortion the pressure of melted metal.
2. The production of a core equally hardened on all its faces. It is a fact that it is not only the core part in contact with the metal which is of importance in the precision of an assembly. The remainder of the core serves either as a bearing surface or as a support, and in consequence, has a large part in this precision.
3. The production of a core having a high superficial hardness, that is having a similar strength to that of the conventional oil core, if not higher. This high strength (estimated at the minimum at 250 psi tensile strength) is in fact necessary in cases of cores of small section, submitted to heavy stresses in the course of remolding manipulations and during casting. Such cohesion values are not obtained with silicate sands, $CO_2$ hardened. Their specifications after hardening are much nearer those of dried molding sands than those of oil cores.
4. The use of organic binders. The advantages of which over mineral binders are well known – low resistance to contraction of the cooling metals, easy shake out, no sand vitrification giving an improved finish to the moldings.

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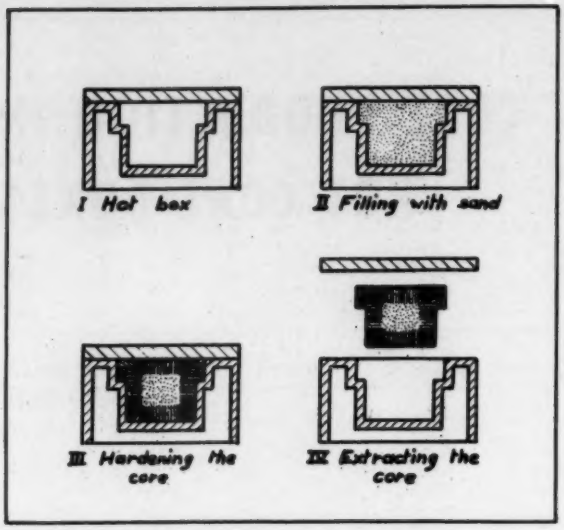



Fig. 1 — Principle of the hot box core process.

5. The use of a liquid binder. This notion is essential for questions of performance. A liquid binder easily enrobes the grains of sand in a uniform film; furthermore, it gives to the mixture a low green strength, thus allowing ramming operations by the usual methods — blowing, projection, squeeze, vibration.
6. Hardening in the box in the quickest time possible (2 min max.) compatible with the rational use of pattern and mass production in automobile foundries. This speed of execution is not obtained with the air set sand technique which requires immobilizing patterns sometimes for several hours.

## HOT CORE BOX PRINCIPLE

Under the circumstances, the ideal would have been to use an organic binder, capable of instantaneous hardening when cold following a similar reaction to that of carbon dioxide on sodium silicate. Such a binder not having been placed at the foundryman's disposal at least for the present by the chemical industry, the only range of suitable products, practically, remain the thermosetting group, i.e., products which are liquid at ambient temperature and which under the influence of an increased temperature are capable of becoming a three dimensional molecular network which render them solid and nonfusable. If it is added that heat transmission phenomenons are carried out much more easily if the source of the heat is a solid instead of a gas, which means that a mass of sand heats up much more rapidly when in contact with a hot solid, than when placed in the midst of a hot gas, one arrives quite naturally at defining the principle of core production in hot boxes.

The principle of this process, detailed in Fig. 1 is:

a) A core box (I) pre-heated to a certain temperature, is filled with mixture of sand and liquid thermosetting binder (II).
b) In contact with the hot box, the sand sets through the polymerizing of the binder, to a certain depth which varies according to the temperature of the box and the curing time (III).
c) From the box a core, which can be considered as finished, is then extracted, as it has acquired its final hardness (IV). The box is thus freed for a further operation.

One must insist on the necessity of baking in an entirely closed box, so as to obtain the hardening of the entire surface of the core. The closing of the box can take place either before or after the filling operation according to the ramming process used. Furthermore, one will note that a core obtained through this process will appear as a compact mass hardened on all its faces, the center being in some state or other, hardened or not by the polymerization of the binder.

## BINDER CHOICE

It must be understood, the carrying out of such a process requires for the binder supplementary specifications which further reduce the possible choice in the thermosetting group;

1. The baking speed. All thermosetting binders do not polymerize at the same speed. In practice, the most interesting binders are those which are capable of hardening at the lowest temperature.
2. Absence of thermoplasticity in the range of temperatures likely to be attained with the hot box contact, otherwise there is a risk of distortion when stripping, thus destroying the advantages inherent to hardening in the box.
3. The price. As the process produces solid cores, the cost price of the prepared sand should be of the same order as that of conventional sands bonded by cereals and oil.

Before this assembly of properties, the company was obliged to eliminate the following thermohardenables:

a) siccative oils, as their hardening is slow and furthermore they are thermoplastic.
b) polyesters and epoxides, the cost of which is extremely high in France.
c) phenoplasts, also expensive in France, although there are certainly possibilities in the direction of certain solutions or aqueous emulsions.

Finally, only two categories of products remained, corresponding to the general characteristics stated. On the one hand there were the binders derived from sugar. Particularly cheap, those binders are capable of rapid hardening with the addition of appropriate catalysts and their supplyers are apt to supply in regular qualities. On the other hand, the urea-formaldehyde resins. Easily obtainable, also inexpensive and capable of hardening at low temperatures.

But it is equally clear that this list is not complete, as owing to the principle of core production in hot boxes itself, requiring only one fundamental specification, that of quick polymerization under heat effect, a great number of products could be perfected

to suit this process.* However, up to now the company has only used sugar and urea-formaldehyde based binders. These binders give to the sands particular specifications.

## SPECIFICATION OF SANDS BOUND BY HOT BOX HARDENABLE BINDERS

From these two groups of binders—sugar and urea-formaldehyde, a range of formulas containing, either only sugar or only urea-formaldehyde, according to the type of production have been developed. However, mainly in use are mixed formulas containing both sugar and U.F.

The principal specifications of the binders used are:

1. Water contents fairly high from 30 to 50 per cent. Presence of water is indispensable to obtain a liquid binder at normal temperature, but it is inconvenient in the course of the baking of the core, as it must be evaporated.
2. A low viscosity, which enables an easy distribution.
3. The presence of an acid catalyst for polymerizing.

### *Sand Preparation*

The preparation of the sand is an extremely simple operation, as it consists only in distributing a low viscosity binder in the sand and any kind of core sand mixer is likely to suit. Care must be taken only to avoid the excessive evaporation of the water in the binder. The basic sand should be of a high grade of mineralogical purity. The presence of clay, mica and organic matter greatly reduces the dry strength. In practice sands are used, either new or wet reclaimed, of an AFS fineness number between 5 and 105. The agglomeration of olivine and zircon sands is also obtained without difficulty.

### *Green Properties*

Hot box sands present a low green strength. Their compression strength is between 0.3 and 0.6 psi, and are therefore excellent for blowing. Mixtures should be kept away from air, as their contents in water must be calculated with precision. Lack of water reduces hardness after baking, while excess of water slows the baking down.

Bench life depends entirely on the quantity of acid catalyst. Whereas sugar based mixtures have a practically endless preservation, it is not so with those containing urea-formaldehyde. In practice, the quantity of acid catalyst is determined so that a 24 hr preservation is assured.

### *Dry Properties*

Dry properties can be considered under the double point of view of strength and baking speed. On Fig. 2 will be found baking curves for four binders:

a) Linseed oil in the proportion of 0.8 per cent.
b) Formophenolic resin for shell molding in the proportion of 1.5 per cent.

*In particular, furanic resins which are being greatly developed just now will not be examined in this report.

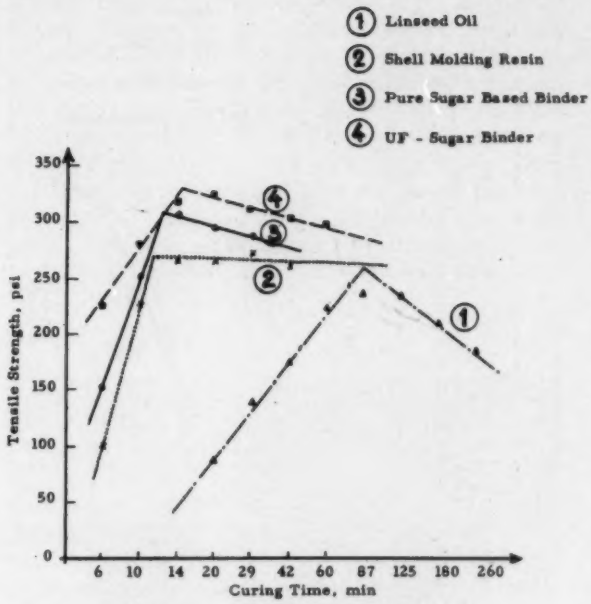



Fig. 2 — Baking curves for four core binders.

c) A sugar base binder for hot boxes in the proportion of 2 per cent.
d) A sugar and U.F. base binder for hot boxes in the proportion of 1.5 per cent.

The first three binders are dried at 225 C (437 F) in an ordinary oven, the UF-sugar mixtures at only 125 C (257 F). It is possible to establish that:

a) From the strength point of view, a tensile strength of 300 psi is obtained with 1.5 per cent of UF-sugar mixture, and of 2 per cent of sugar base binder whereas 260 psi is the maximum with 1.5 per cent shell resin and 8 per cent linseed oil.

The excellent dry strength of the binders, thus made obvious, is due undoubtedly to their excellent distribution in liquid form.

b) From the point of view of baking speed, maximum strength is obtained in 87 min with linseed oil, in 12 or 13 min with the other binders, which show effectively the aptitude of hot box binders to harden quickly, their baking speed being of the same order as that of the phenoplasts widely used with hot pattern contacts. This means that, when using hot boxes, the constitution of the hard crust can take place practically at the same speed as in the case of shell molding. In practice, one can count on obtaining a hard shell of 8-10 mm (0.315 to 0.3937 in.) in thickness in only 2 min baking.

It is also to be noted that the mixture UF-sugar is the quickest and that, only at a temperature of 125 C (257 F), the constitution of the hardened shell is therefore hastened. Finally, the catalyst quantity dosing allows the mixture to continue to harden to the center after the core has been extracted from the box.

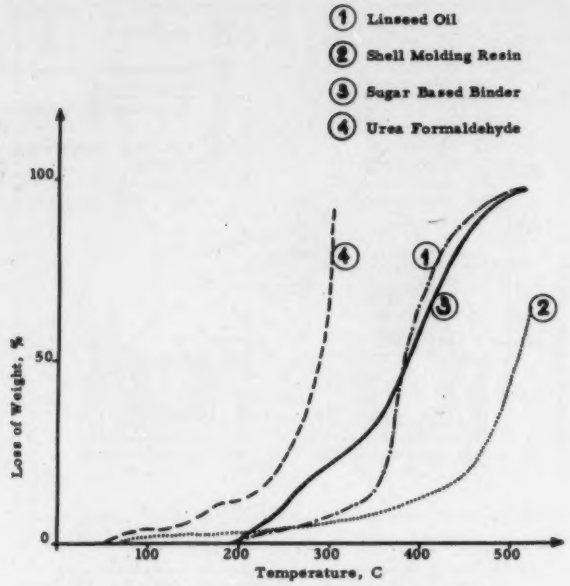



Fig. 3 — Thermal analysis of thermosetting binders.

### *Behavior While Casting*

The diversity of mixtures which are used is precisely due to their varied behavior during the casting. This behavior proceeds from the thermoanalysis curves on Fig. 3. Pure urea-formaldehyde resins deteriorate quickly under heat effect. Their use is therefore reserved for light alloy castings. Sugars have a thermic decomposition curve nearing that of linseed oil. They are, in our range of hot box binders, those with the best heat resistance. They allow the casting of small steel parts, even of low carbon steel. U.F. sugar mixtures show intermediary characteristics and can be used for ferrous metal castings. They are the binders generally used for the casting of parts in gray iron, ductile iron, malleable in small and medium thicknesses, often used in automobile foundry.

### *Collapsibility and Casting Finish*

The shake out of cores obtained in hot boxes does not offer any particular problem. Collapsibility is generally good, whether it be with pure U.F. in aluminum foundry or with UF-sugar mixtures in ferrous foundry. One can even say that one of the most frequent problems is to avoid premature knock out of the cores, which loosens sand. The casting finish is quite acceptable, although in this process, as in the others, it depends essentially on the basic fineness of the grain. Cores produced in coarse sand, and submitted to severe thermal conditions, can easily be dipped in a wash immediately after stripping from the box. The heat stored in the core is generally sufficient to ensure complete drying without further heating.

### *Sand Formula Cost Prices*

The chart, giving cost prices for different sand formulas used at the author's company, shows with precision the economical interest of hot box core production (at least as far as French prices are concerned).

| Sand Formula | Cost of binders in N.F. per 100 kg of sand |
|---|---|
| Sand, linseed oil, dextrin | 2.5 to 3 |
| Sand, silicate hardened with $CO_2$ | 1.7 to 2.3 |
| Sand, shell molding | 10 to 12 |
| Sand, for hot box, UF pure base | 3 |
| Sand, for hot box, sugar base | 2.75 |
| Sand, for hot box, UF-sugar base | 2.50 |

Thus, it is seen that hot box sand formulas have cost prices which compare quite favorably with those of the other formulas used to produce solid cores — on one hand oil-dextrin, on the other silicate CO It is also to be pointed out that, among the sands liable to hardening in contact with hot patterns, there there is a great difference in price between hot box formulas and those for shell molding.

## TECHNOLOGY OF HOT BOX CORE PRODUCTION

Implementing hot box core production consists in carrying out these operations — (a) filling the box with the sand and binder mixture, (b) heating the box to maintain it at a temperature higher than the minimum temperature at which the chosen binder polymerizes and (c) extracting the core from the box. This production cycle having to be carried out at a given rate requires special equipment and boxes.

### *Filling the Boxes*

The ideal way to fill the boxes quickly is, without doubt, by blowing. It can be carried out under excellent conditions owing to the low green bond of the sands for hot boxes and one of the big advantages consists in the possibility to use ordinary blowers. The only really indispensable modification concerns the use of the water cooled blower plate. Where it is not possible to blow the core through a port on the outside of the box, it is necessary that the blow tubes pass through the upper half of the core box while remaining integral with the blower plate. To this end conical nozzles, a type of which appears in Fig. 4 are used.

A washer and spring retainer assembly ensures a certain play in the nozzle as regards the blower plate, and this allows the nozzle to center itself in the core box cover whether hot or cold. Concerning venting when blowing, ordinary slit filters or perforated brass plates are used. The latter system is the better. Finally, it is to be noted that owing to the low green bonding of the sand, the blower holes can be of a small diameter and this allows a correlative reduction in the size of the venting surface. Two models of boxes for blown cores appear on Figs. 5 and 6.

Apart from the blowing, another method is used where greater ramming density is necessary to bring out certain details of the core. By using squeeze and vibration it is possible to obtain ramming densities around 100 lb/cu ft, the lower half of the box having

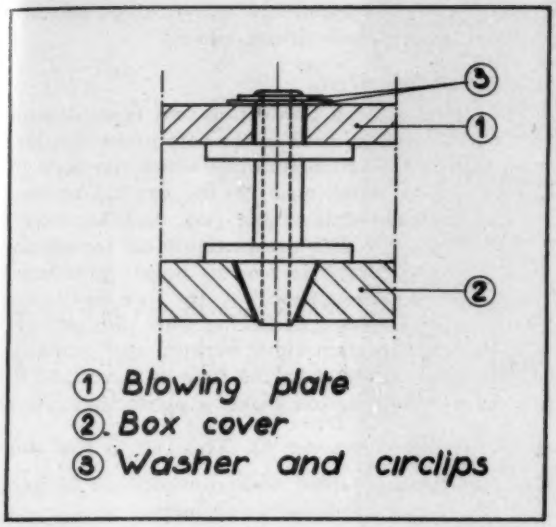



Fig. 4 — Type of blower nozzle.

been filled first by low pressure blowing (20 to 40 psi). This technique is used in particular for the production of flat cores used for stack molding small castings. Figure 7 shows a box built for this type of ramming.

*Heating the Boxes*

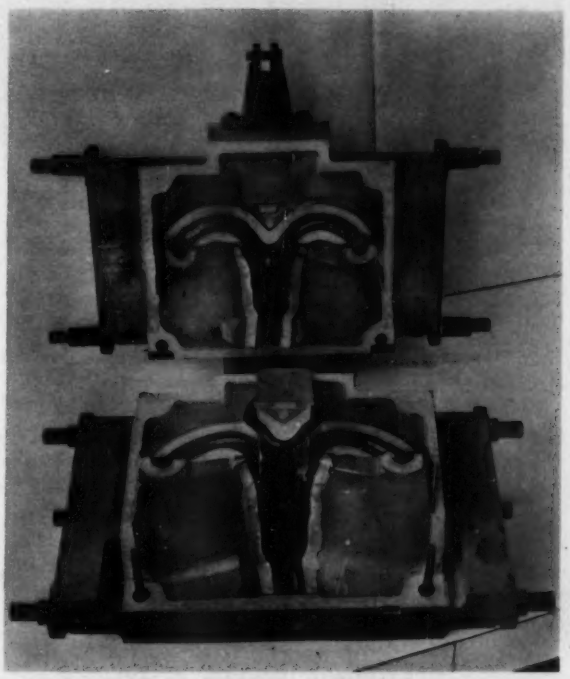
Heating the boxes is one of the important elements in carrying out this process. First of all, one must consider that the introduction of a mass of cold sand into a hot box cools the latter considerably, the sand literally pumping up the calories stored in the box. There is a risk of the box temperature falling below the minimum temperature for the binder polymerization.

As in general one can assume that the heat necessary for the hardening of a solid core is greater than the heat capacity of the pattern, two imperatives result therefrom — (a) use an extremely powerful heating source and (b) apply the heat when the sand is in the box and not before. Under these conditions, the solutions adopted are of two sorts according to the choice of a continuous heat source or of a discontinued one.

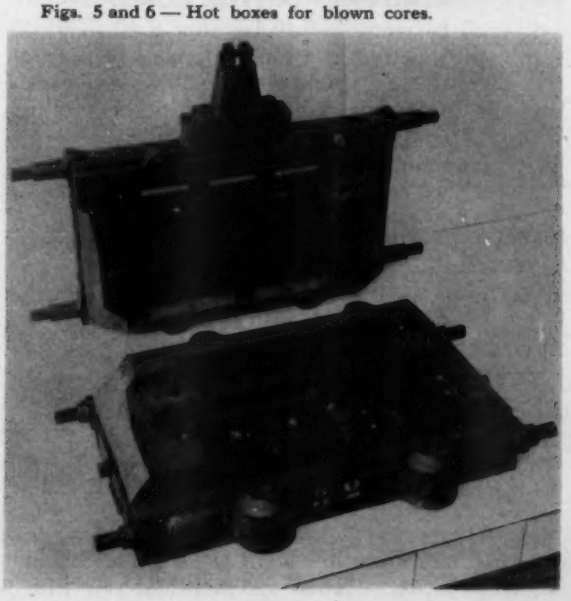
**Heating by electric resistance units incorporated in the box.** This is the most favored medium, as it permits an easy control, and its efficiency is great but it requires special study for each box. It is often difficult to place resistances on core boxes having numerous accessory equipments such as filters, blow holes, stripper pins. It is necessary to incorporate high power resistances in the core boxes in order to be able to control easily. Up to now, this type of heating has been used only for boxes built in aluminum which has a good coefficient of thermal diffusivity. For core weights under 1.5 kg (3.3 lb) the power installed on each box is from 7 to 8 kw, consumption settling around 5 kw.

**Heating by radiation.** In particular this refers to the use of gas heated radiating elements. These elements comprise a ceramic surface which the gas combustion heats to 700-800 C (1292-1472 F), and which radiates directly on the outer surface of the boxes. This type of heating, although discontinued has the advantage of allowing the application of important heating power. The model used gives more than 3000 calories (11.900 Btu hr) for an emission surface of 8½x6½-in.

It happens, furthermore, that the use of radiation solves the problem set by core production in hot boxes particularly well, as the object being to transmit heat to the sand, the box plays an intermediary role, trans-



Figs. 5 and 6 — Hot boxes for blown cores.



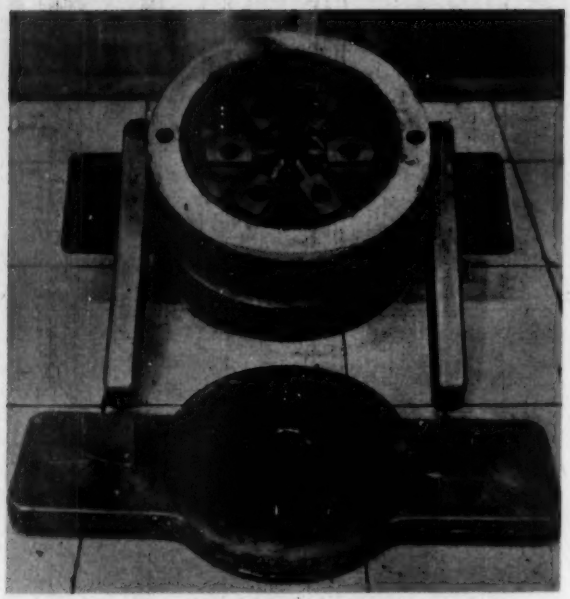


Fig. 7 — Hot boxes designed for ramming by squeeze and vibrations.

ferring by conductibility the calories received by radiation. The drawback in heating by radiation is that it is difficult to control, as the efficiency of an installation depends as much on the transmitter (surface and temperature) as on the receiver (power of absorbtion of the surface) and the distance between the radiant and the box. In practice, control is obtained by varying the gas pressure (this controls the transmitter temperature), and at the same time the distance between the box and the radiant.

In any case, and whatever may be the heating medium employed, baking costs are much below those of conventional core production.

## STRIPPING OF THE CORE

It is to be remembered that after hardening, the core even when hot, is a solid extremely hard on all its faces, and has a tendency to bind owing to expansion and contraction respectively of the box and the core. The best method of stripping without doubt is by means of ejectors. Ejectors are used having the shape of valves normally resting on their seats in the bottom of the box. These ejectors slide in a sleeve with plenty of clearance. At their base, an adjusting screw ensures a simultaneous lift. They have no retracting spring, and are brought back into position either by direct control operated by the box cover or by permanent magnets on the stripping machine.

In any case, one must insist on the fact that most of the stripping difficulties are due to defects in the pattern, badly calculated tapers or even undercuts render ejection difficult if not impossible. That is one of the reasons for which hot boxes require utmost care in preparation, if possible by machining, in alloys presenting a high mechanical resistance at high temperatures. From this point of view, notwithstanding unfavorable thermal specifications, cast iron is much superior to aluminum alloys, even the most resistant (for example silicon alloys).

### *Mass Production Units*

The choice of mass production unit types depends both on the baking speed of the core under consideration and on the production rate which has been decided. It has been indicated that the core baking time, even in the most unfavorable case, does not exceed 2 min. If we take into consideration an installation with only one box in service, its hourly production rate is inversely proportional to the core production total length of time, i.e., baking time plus pre and after baking operation time, blowing and stripping.

Thus for example, a baking time of 2 min and 0.6 min for outside time lost makes a production rate of $r = \frac{60}{2 + 0.6} = 23$ cores per hr. This rate is low compared to those obtained with conventional blowers, that can exceed 150 blowings an hour.

As the baking speed can hardly be reduced, owing to the size of the core and the kind of binding selected, one is obliged to set up installations comprising several identical boxes in operation, some in the course of baking while others are being stripped or blown. Thus a rate of 240 an hr will be obtained for cores requiring 1.5 min of baking with $\frac{240 \times 1.5}{60} = 6$ baking boxes. To these 6 boxes, if one adds one being stripped and one being blown, there will be a total of 8 boxes in service at the same time.

All the mass production installations developed at the author's company were studied in view of the production rates to be carried out, they may be classified into 3 types — two station, four station and six or eight station groups.

**Two station unit (Fig. 8).** This unit comprises two boxes in action; two stripping machines are placed on each side of a blowing machine. The boxes are electrically heated, therefore, by continuous heating. This unit was developed for the production of thin cores, the baking time for which did not exceed double the stripping and filling times. Thus the baking time varying from 40 to 80 sec, production rates arrive at 70 to 120 cores an hr. On this model, a blower plate transfer system permits the simultaneous production of two different cores which can immediately be assembled hot, when taken out of the boxes.

**Four station unit.** This is the most simple unit developed, it includes a blower, a stripping station and a revolving oven with three heating stations. The oven revolves automatically and is controlled by a time switch adjusted to the desired rate. Four core boxes are in service simultaneously, three of which are being baked, while the fourth passes from the stripping to the blower. Handling of the boxes is entirely manual.

Such an installation manned by two men is capable of a production that can attain 150 cores an hr. Its simple mechanization makes it an ideal means of production for small cores. Figures 9 and 10 show two views of the installation in operation. On Fig. 9 can be seen the blowing operation of a core box with an absolutely standard blower. Figure 10 represents



Fig. 8 — Two station production unit.

the manual stripping of a core – the operator utilizes the removable part of the box for extracting.

**Six or 8 station units.** This machine has a turntable with 6 or 8 stations transferring the boxes successively to the blowing station, the baking station (4 or 6 stations) and the stripping station. It is entirely automatic, and with one operator attains production rates of 240 cores an hr. All the movements of this machine, whether hydraulic or pneumatic are controlled by a program controller.

Figure 11 shows the general view of an 8 station machine. On Fig. 12 can be seen, near view, a stripping station, the core box is open, the core is extrated by ejectors. This 6 or 8 station unit is now in standard production. The rate is adjusted from 180 to 240 cores per hr as needs require. The number of stations in the unit depends entirely on the baking time for the core to be produced, 1.5 min minimum for 8 station machines and one min for the 6 station machines.

As a whole, it can be said that hot box coring can be carried out on a wide variety of production units, as shown by those developed at the author's company. Following the rates required, the degree of automation desired, it is evident that it is possible to use quite simple units with relatively important manual handling or entirely automatic mass production units.

However, these machines have one common feature, that of being relatively light and therefore easily movable. Here is an extremely favorable productivity factor, the redeployment of core shops fitted with hot boxes can be easily carried out as needs require. It is not so with conventional core shops which are necessarily retained by a fixed element, the oven.

## FIELD IN WHICH BOX CORING MAY BE USED

The field in which hot box coring can be used is as much that of producing cores proper as that of producing mold parts. On Figs. 13 and 14 appear a certain number of cores realized according to the process and which are used for the making of various parts used in building an engine. One can see cylinder head cores, pipes and water jackets; inlet and exhaust manifold cores; cylinder block timing gear core and pulley grove cores. All these cores are exclusively produced in hot boxes, for production rates which can exceed 10,000 engines per week.

Their weight does not exceed 3 kg (6.6 lb) and they are solid. Their shape is that of conventional cores produced in oil sand. It is evident that for these types of cores which can be used without any finishing operation as soon as they leave the boxes, hot box coring constitutes the most rational means of production. It suppresses the dryers, the core wires and grinding of cores before assembly (in particular in the case of cylinder head cores).

A reduction in manufacturing time is the result thereof, as it is apparent in the chart:

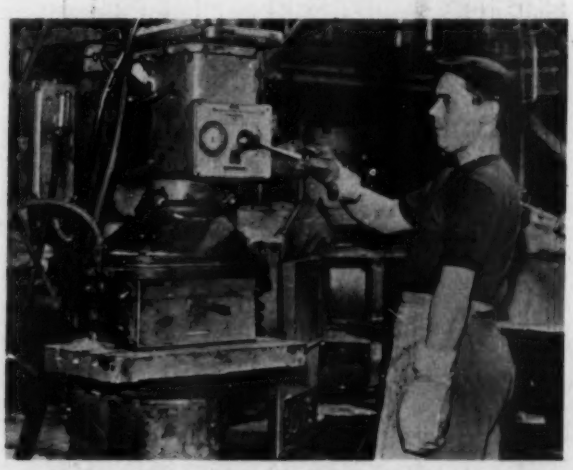


Fig. 9 — Four station unit. View of the blower station.



Fig. 10 — Four station unit stripping station.

| Cores | Core production time in min: Conventional oil core | hot box core | Time ratios % |
|---|---|---|---|
| Cylinder head assembled cores: water jackets; plus pipes (aluminum) | 16.5 | 4.5 | 27.3 |
| Inlet manifold (aluminum) | 1.90 | 0.45 | 23.7 |
| Exhaust manifold (cast iron) | 3.54 | 0.83 | 23.4 |
| Timing gear core (cast iron) | 3 | 0.91 | 30 |

It is thus possible to see that the production times are reduced by three quarters. The actual problem really is that of the limits of the process – how far can one go in weight and dimensions? For the time being, the company has not tried to produce cores with a thickness above 3 in. Under these conditions, the relative importance of the unbaked core body with regard to hardened outside layer is relatively small.

It is not thought that production of more voluminous cores is impossible, but beyond a certain volume it is necessary to take into account the quantity of sand used and the saving obtained by producing partly hollowed cores. Under these conditions the hollowing out of the core must be decided when the box is designed. An experimental realization towards

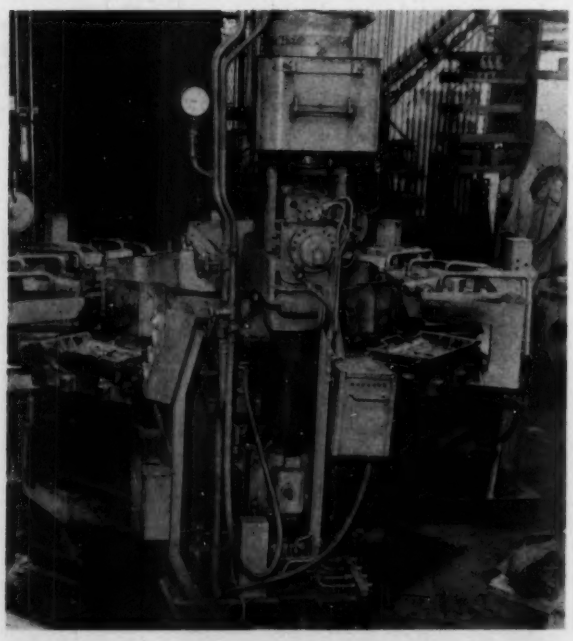


Fig. 11 — Eight station unit general view.

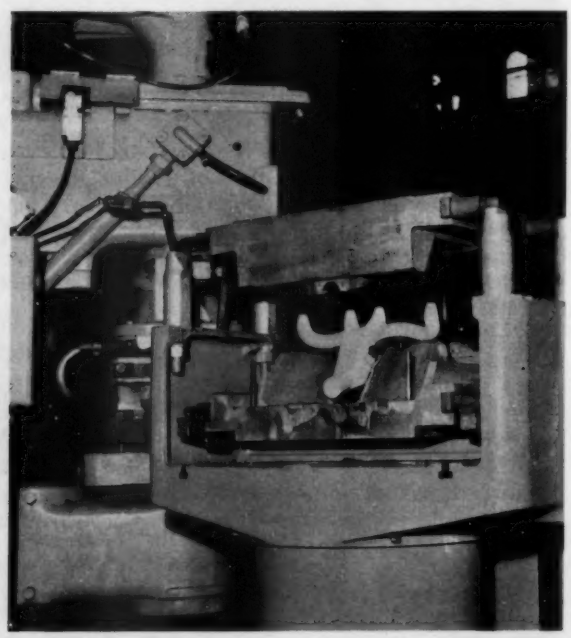
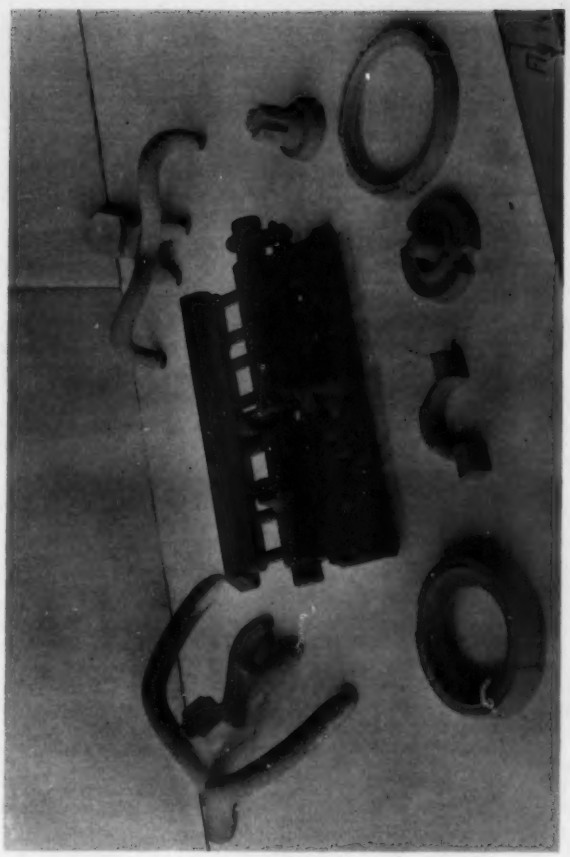
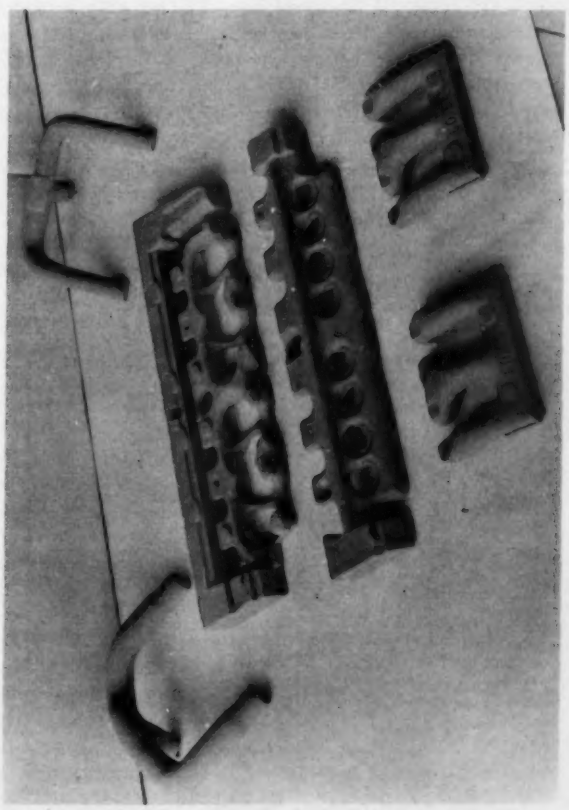


Fig. 12 — Eight station unit, view of the stripping station.

Figs. 13 and 14 — Samples of cores produced in hot boxes.





lightening the core has been carried out in the molding field. Figures 15a and 15b represent a half mold for a crankshaft, cavity side and opposite side.

It can be seen that large cavities have been made on the back of the core, the external shape of the core being maintained with narrow ribs. As it is, the core weighs 17 Kg (37.5 lb). It was obtained by blowing. Two cores of this type, assembled to form a complete mold are enclosed between two metallic plates fastened by clamps and cast without any intermediary packing to support the core part in contact with the metal. This method, in its experiemental stage for the time being, shows most interesting possibilities. It is a simple solution to the clamping problem for shell molds, and it is certain that numerous parts can be made this way.

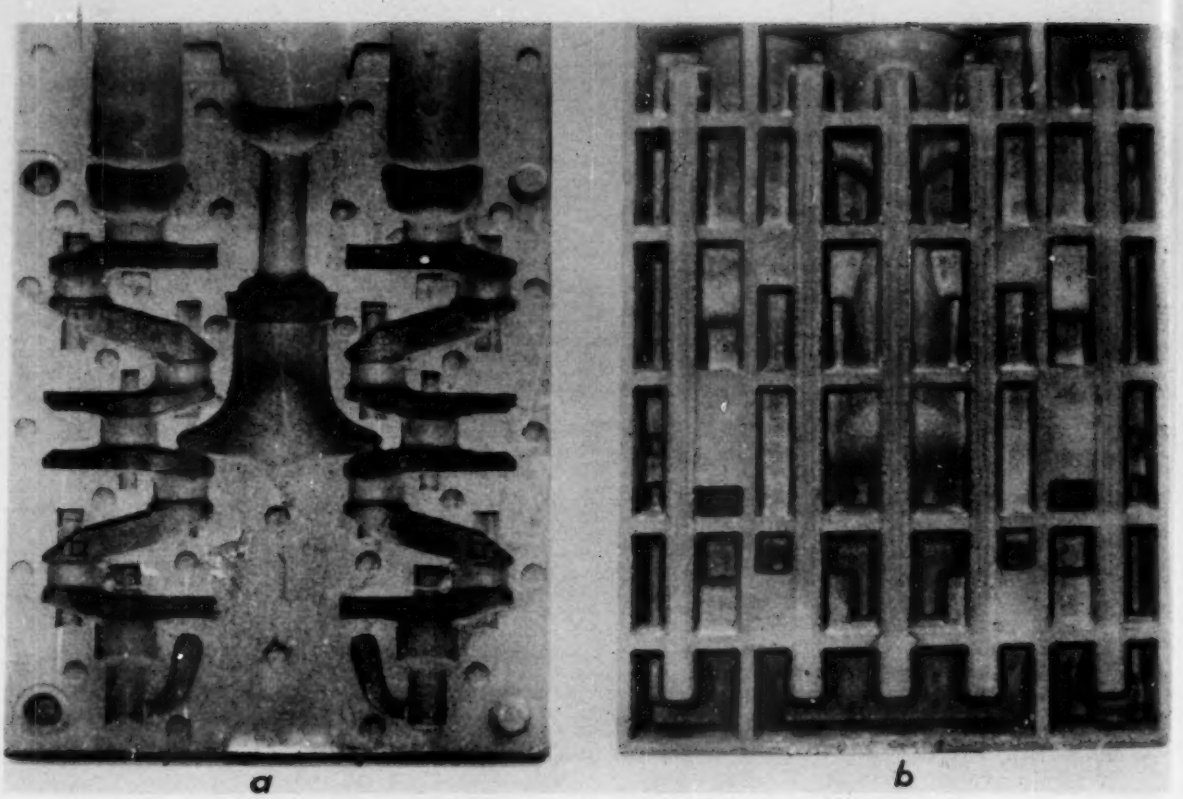
Other molding research has been carried out in the field of small parts in steel or special iron cast in stacks. In this case the company has taken advantage of the possibilities offered by the hot box process for making flat cores of regular geometrical shapes, therefore suitable for stacking and thus for stack molding. Figure 16 shows a certain number of molds of this type, as well as the corresponding castings.

These molds were obtained by the mixed technique which consists in filling the box by low pressure blowing and carrying on the ramming by squeeze and vibration. On Fig. 17 can be seen an application of this method of molding to the production of rocker arms. The two cores when assembled form one stage of the stack, the mold itself being made by the stacking of 12 cores topped by a sprue cup core. The whole stack is held together by clamps. On the right is a group of castings. The field in which hot box coring may be used seems to be relatively wide. It would seem, in particular, that the technique of the mold and core partly hollowed out and ribbed shows certain possibilities in the production of fairly large molds.

## CONCLUSION

The hot box core production process seems to have succeeded in reconciling two tendencies which were in opposition until now, making more precise molded castings, at lower cost prices than those of the conventional methods. This result was obtained with the use of thermosetting binders and efficient and capable of polymerization when in contact with a hot box. It is thus possible to obtain a core or part of a mold which reproduces faithfully the boxes used to form them. Standard mixers and blowers are used in this process, which suppresses many of the numerous operations which are so expensive in the traditional methods.

In favor of this coring method are (1) simplification of sand preparation and suppression of dryers and drying plates, core wires, large core ovens, generally bottle necks in core production shops. These




**Figs. 15a and 15b — Samples of hollowed and ribbed molds produced in hot boxes.**

ovens are replaced by small heater units with a high thermal efficiency, therefore saving fuel and accuracy of core assemblies which normally reduces finishing operations. It is to be noted that these results can only be obtained with high quality core boxes, which must be studied with extreme care.

Examples have established the present possibilities of the process, whether it be for the production of cores for different parts, or for molds for small or medium parts. It is believed the hot method will allow sand molding to maintain its position as a mass production process.

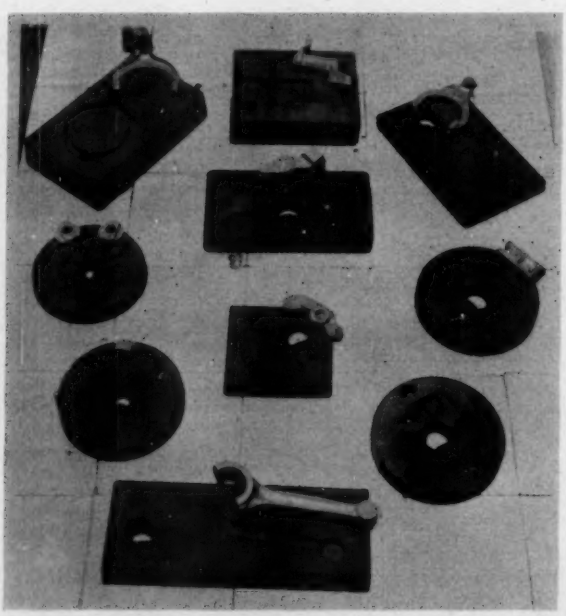


**Fig. 16 — Molds for stack molding.**

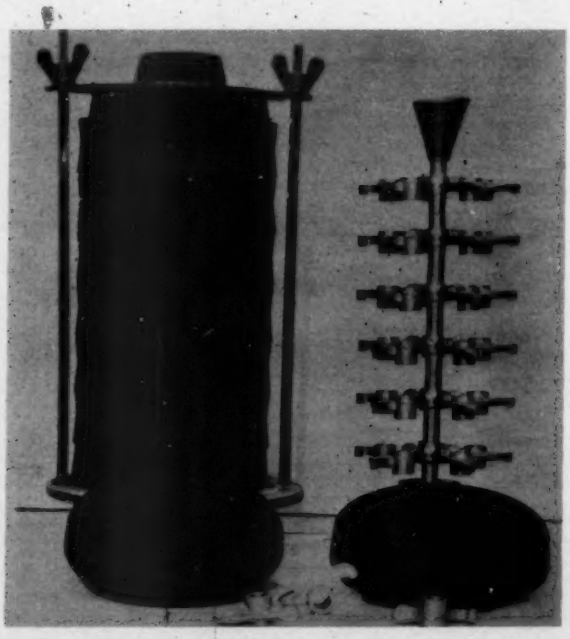


**Fig. 17 — Stack castings of rocker arms.**

# DUCTILE IRON GRAPHITE FORM CLASSIFICATION

by C. K. Donoho


## ABSTRACT

**Graphite type and size charts are presented for the rating of structures of ductile irons. Charts are needed to classify graphite forms of spherulitic shape, and modifications between flake and spherulitic graphite. Matrix structure is estimated by examination of etched specimens under magnification and estimating percentages of carbides, pearlite and ferrite. These charts are offered as an interim suggestion until standards have been established for these graphite shapes and sizes.**


## INTRODUCTION

A.S.T.M. Standard A-247-47 was developed prior to 1941 by a Joint Committee of AFS and A.S.T.M. This standard has served its purpose well, and references to Type *A* and Type *D* graphite, the most frequently used, have become bywords in descriptions of gray iron structures. A-247, of course, pertains only to the types, and more specifically, to the distributions of flake graphite in gray cast iron. Obviously, there is now a need for further classifications to include spherulitic graphite and modifications or transition forms between flake and spherulitic graphite.

## GRAPHITE CHARTS

The first published work that we have seen which includes classification by type and size of spherulitic graphite (Kugel Graphit) is that proposed by the Cast Iron Committee of the German Foundrymen's Association under the chairmanship of Dr. A. Wittmoser. In its complete form, this classification uses A.S.T.M. A-247 for the flake graphite types and sizes, and adds charts for spherulitic graphite and for temper carbon graphite as seen in malleable iron.

The charts for type and size of spherulitic graphite (including transition forms) are reproduced as Figs. 1 and 2.

These can be used satisfactorily to grade or specify ductile irons as to microstructural quality. Type *K* is, of course, the most desirable for optimum mechanical properties, and usually the smaller graphite sizes are preferred. A typical specification might be – at least 75 per cent Type *K* with not more than 15 per cent of Type *P* and size 3 or finer. This rating has been used effectively at the author's company in routine quality control.

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Along with the graphite types and sizes, the average percentages of pearlite, ferrite and carbides, if present, are also recorded. This can be used conveniently to describe ductile iron metallography for practically all grades of ductile or semi-ductile irons.

## INTERNATIONAL COMMITTEE STUDIES

There is also an International Committee on Methods of Testing Cast Iron, Prof. A. deSy, *Chairman*, which is studying graphite classifications, and hopes to have a final draft approved soon. Other graphite classification schemes include the Italian specification UNI-3775, and Czechoslovakian specification CSN-42-0461. The International Committee is considering all these as well as the German system in an effort to produce a truly International Specification on graphite classification.

Through Dr. A. B. Everest of the Mond Nickel Co. and AFS headquarters, preliminary charts from this work have been seen, but since these have not had final approval of the I.S.O., they cannot be reproduced.

In the studies of graphite forms encountered in ductile iron, the chart, shown as Fig. 3, was evolved. This is similar to the German classification chart in that five types are shown. However, our Type V is not included in the German classification. That Type V does occur is shown by the micrograph, Fig. 4. There is some question that Type V occurs frequently enough to warrant inclusion in the standard chart.

Type I is the most desirable type for most purposes, but the presence of Type II does not affect properties adversely to any marked degree. Type III is the quasi-flake or vermicular graphite which is usually associated with an insufficient magnesium content. Some Type IV graphite is seen when magnesium content is excessively high, or in the presence of certain subversive elements. At present, the mechanism or the cause of the formation of Type V is uncertain.

Roman numerals were selected for the type designations because this avoids any confusion with A-297 letter designations for flake graphite distribution.

## GRAPHITE SIZE CLASSIFICATION

Figure 5 is an attempt to effect a size classification based on inches and maximum diameter of the graphite particles. However, this proved to be rather awkward, and the German size chart based on area and millimeters is currently preferred.

Figure 6 is a micro of an iron with predominantly

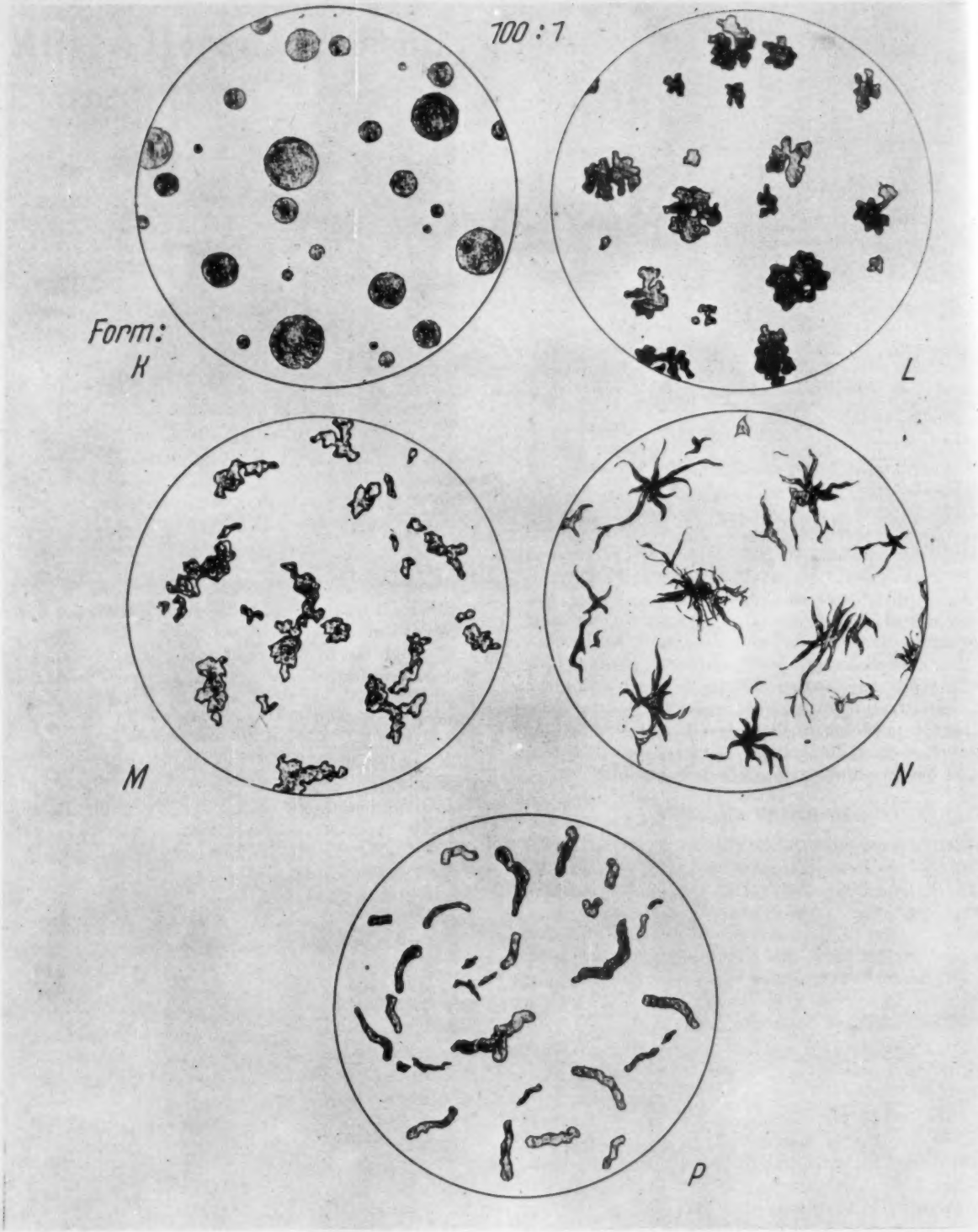



Fig. 1 — Types of graphite proposed by the Cast Iron Committee of the German Foundrymen's Society.

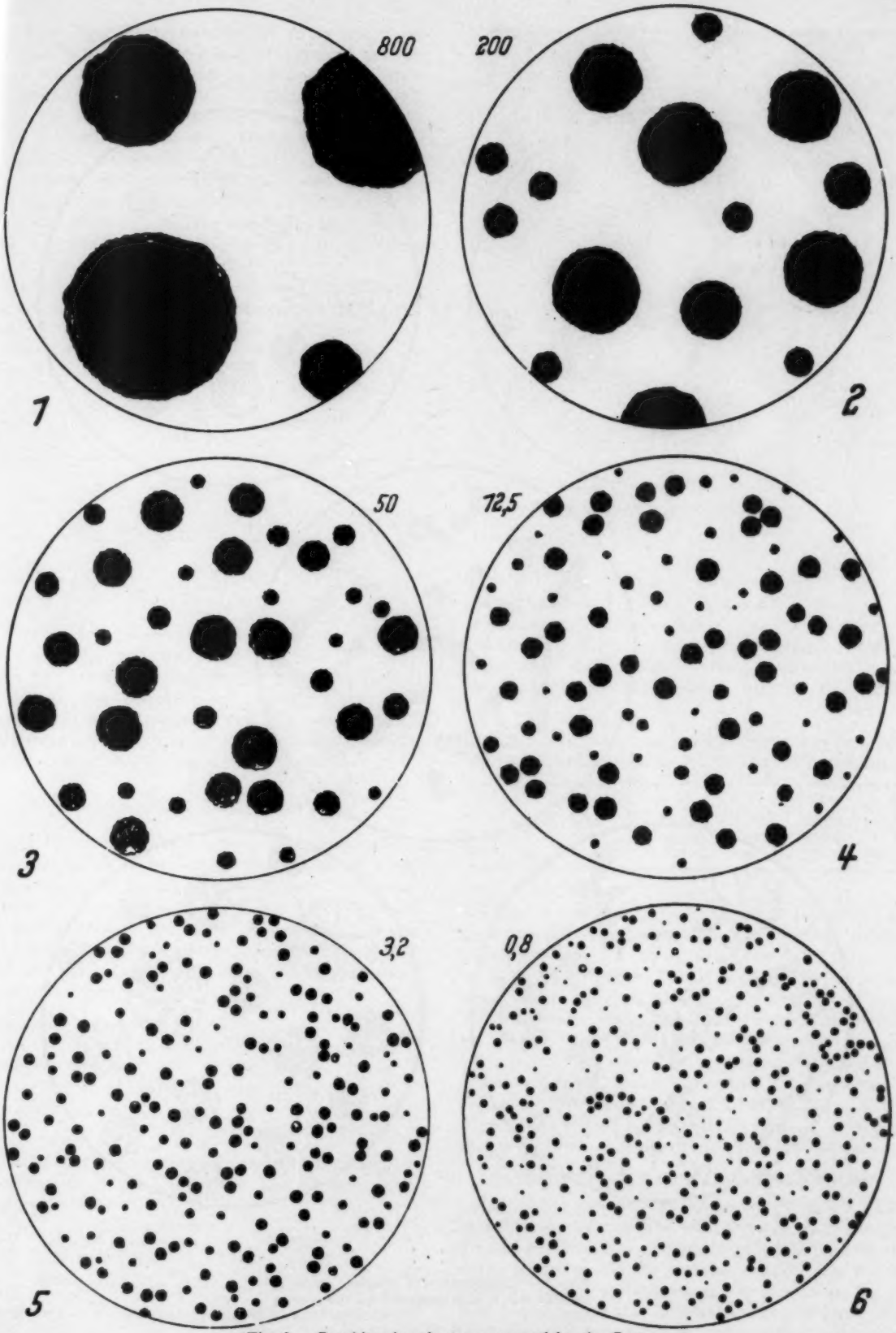



Fig. 2 — Graphite size chart as proposed by the Cast Iron Committee of the German Foundrymen's Society. 100 ×.

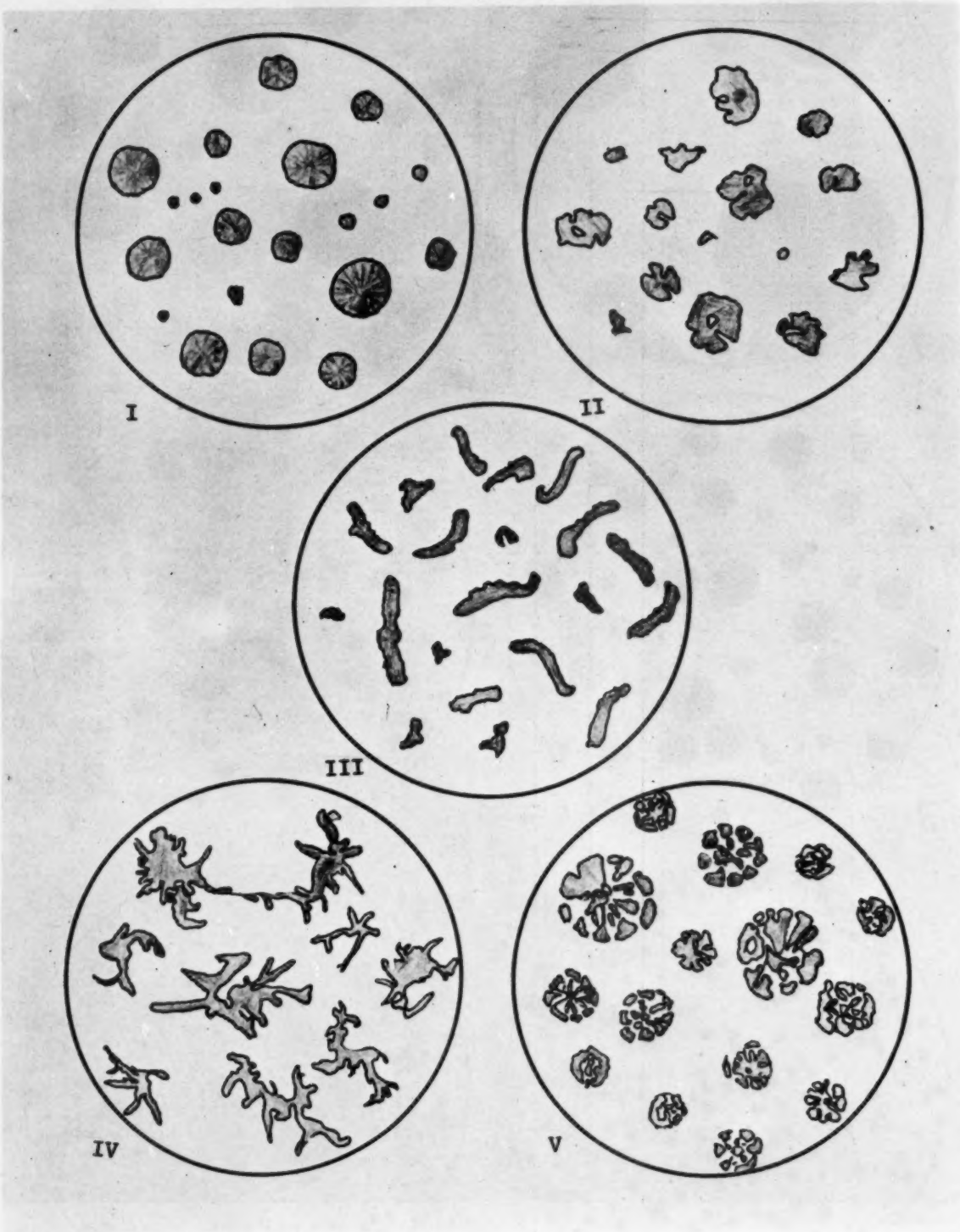



Fig. 3 — Classification of graphite forms in ductile cast iron as proposed by the author. 100 ×.

Type I (or Type *K*) graphite. At 100 ×, the size rating by Fig. 2, was 4. With less than 10 per cent pearlite as-cast, the tensile properties in the as-cast condition were somewhat higher in ductility than usually found. Figure 7 is a micro of a ductile iron just slightly deficient in magnesium content. Figure 8 is a micro of a ductile iron with excessive magnesium content showing occurrence of Type IV graphite as-cast. These last three micros were taken at 200 × in order to show clearly the graphite type. For size rating, the samples were, of course, viewed at 100 ×. For matrix rating, the samples were etched in nital.

In Table 1 are shown the metallographic ratings of these three samples, and the corresponding as-cast tensile properties as obtained from one in. keel block coupons. Graphite type ratings are by Fig. 3 and size ratings by Fig. 2.

TABLE 1

| Structure, Unetched | Graphite | | Matrix, % | | | As-Cast Tensile Properties | | |
|---|---|---|---|---|---|---|---|---|
| | Type | Size | Carbides | Pearlite | Ferrite | T. S. ×10-3 | Y. S. ×10-3 | Elong., % |
| Fig. 6 | I | 4 | 0 | 10 | 90 | 81.6 | 62.6 | 19.1 |
| Fig. 7 | I, 60% III, 40% | 4 | 0 | 30 | 70 | 81.2 | 60.5 | 9.6 |
| Fig. 8 | I, 90% IV, 10% | 4 | 0 | 20 | 80 | 83.7 | 65.4 | 5.9 |

It is to be noted, from samples in Figs. 7 and 8, that considerable quantities of graphite types other than Type I can be tolerated and still meet 80-60-03 requirements. The sample of Fig. 6 meets the tensile requirements of 60-45-15 as well as 80-60-03.

## REGRESSION EQUATIONS

Grant Thomas* of Deere and Co. has shown statistically that up to 10 per cent of vermicular or Type III graphite may be tolerated with only minor reductions in physical properties. He has calculated from statistical data the effect of vermicular (Type III) graphite on tensile strength, yield strength and elongation of as-cast ductile iron. These regression equations are as follows:

Change of T. S., psi
$= -1380\text{-}189$ (% vermicular graphite)

Change of Y. S., psi
$= -1250\text{-}65$ (% vermicular graphite)

Change of Elong., %
$= -0.70\text{-}0.12$ (% vermicular graphite)

Assume that normal as-cast ductile iron of all spherulites, or Type I, graphite has the following average properties:

T. S., psi ..... 90,000
Y. S., psi ..... 65,000
Elong., % ..... 12

These are not unusual properties, and represent what is normally obtained on as-cast tensile tests machined from keel block coupons when the graphite is predominantly spherulitic.

*G. F. Thomas, "How Do You Know It Is Ductile?" AFS TRANSACTIONS, *vol. 68*, p. 182 (1960).

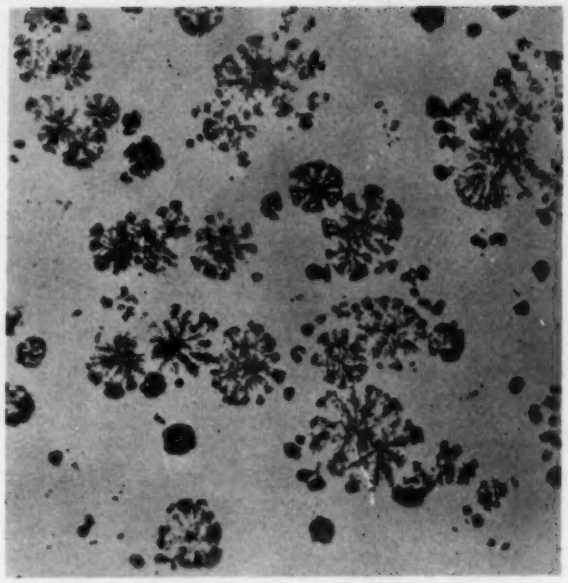


Fig. 4 — Type V graphite. Unetched. 100 ×.

Using the Thomas regression equations, the properties to be expected with varying percentages of vermicular, or Type III, graphite can be calculated.

TABLE 2 — VERMICULAR GRAPHITE EFFECT ON TENSILE PROPERTIES OF AS-CAST DUCTILE IRON

| Vermicular Graphite, % .. | 10 | 20 | 30 | 40 |
|---|---|---|---|---|
| T. S., psi ..... | 86,730 | 84,840 | 82,950 | 81,060 |
| Y. S., psi ..... | 63,100 | 62,450 | 61,800 | 61,150 |
| Elong., % ..... | 10.1 | 8.9 | 7.7 | 6.5 |

From Table 2, with the assumptions made, it is seen that the calculations show that 80-60-03 requirements can be met with up to 40 per cent vermicular, or Type III, graphite. This is confirmed by an actual test, Table 1, Fig. 7. Indications are that a small amount of the "spiky" or Type IV graphite is more detrimental to ductility than is the vermicular type.

Of course, the effects of compositional variables on ductile iron properties are of considerable importance. These effects have been purposely minimized in this discussion which deals primarily with structural effects on irons of generally normal compositions.

## SUMMARY AND RECOMMENDATIONS

Graphite type and size charts are presented which may be used to rate the structures of ductile irons. Present preference is to rate the graphite type by the chart, shown as Fig. 3, and the graphite size by the German square millimeter chart, shown as Fig. 2. The graphite type may be estimated at any convenient magnification, but 100 × is the standard magnification for size rating.

Matrix structure is estimated by examining etched

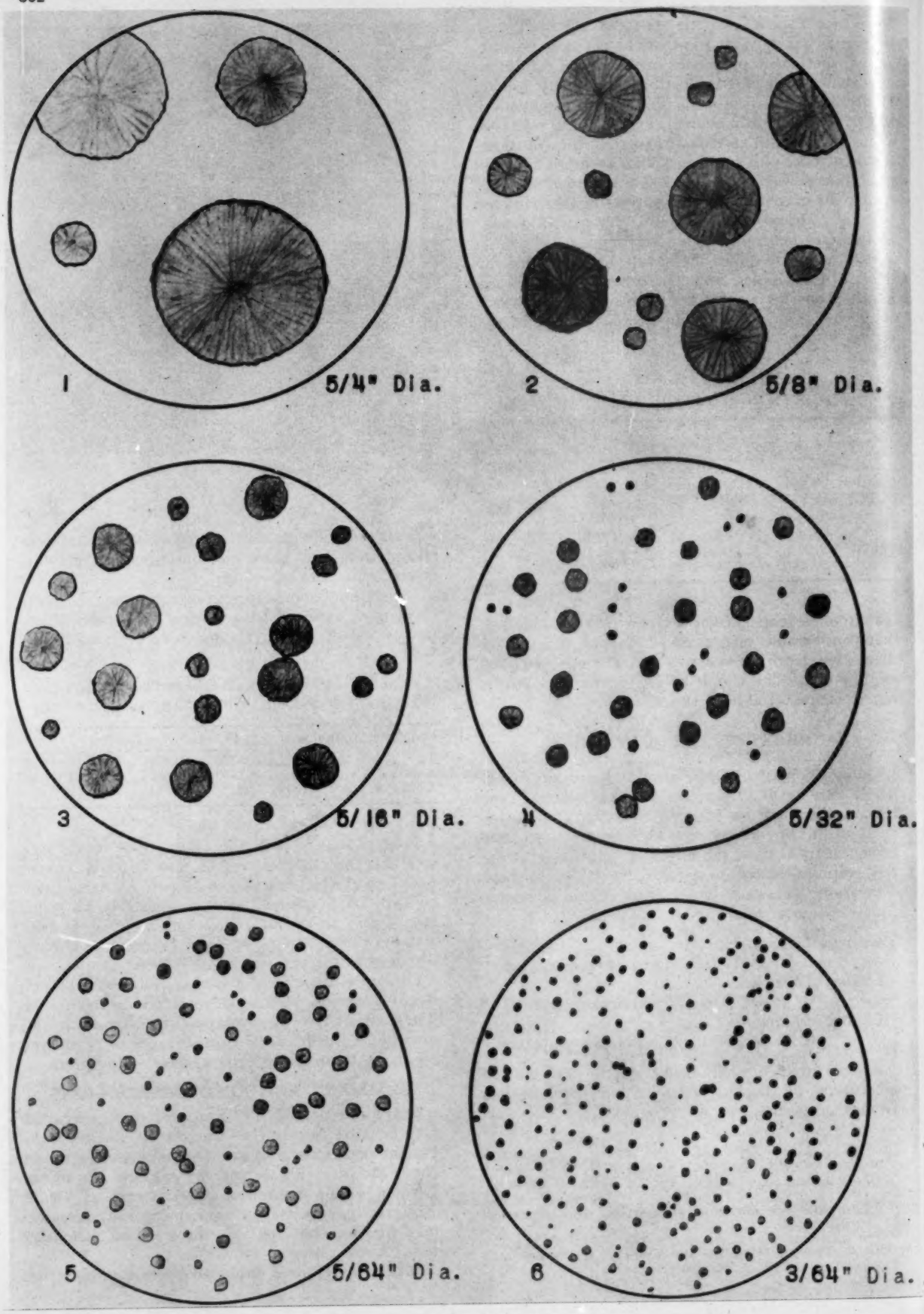

1
5/4" Dia.
2
5/8" Dia.
3
5/16" Dia.
4
5/32" Dia.
5
5/64" Dia.
6
3/64" Dia.

Fig. 5 — Classification of graphite size in ductile iron based on inches and maximum diameter. 100 ×.

specimens at suitable magnification, and estimating the percentages of carbides, pearlite and ferrite. It is shown that sizable percentages of graphite types other than Type I (Fig. 3) can be tolerated without serious effect on tensile properties.

It is proposed that this discussion be considered as an interim suggestion until the International Committee's standard for the classification of graphite has been approved by I.S.O. At that time, the American organizations concerned can decide whether to accept the International Standard or proceed to develop a separate American Standard.

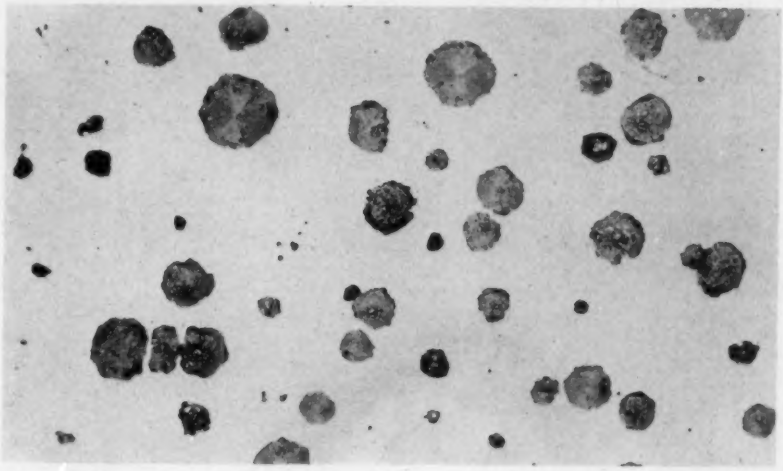


Fig. 6 — Ductile iron with Type II (or K) graphite. Unetched. 200 ×.

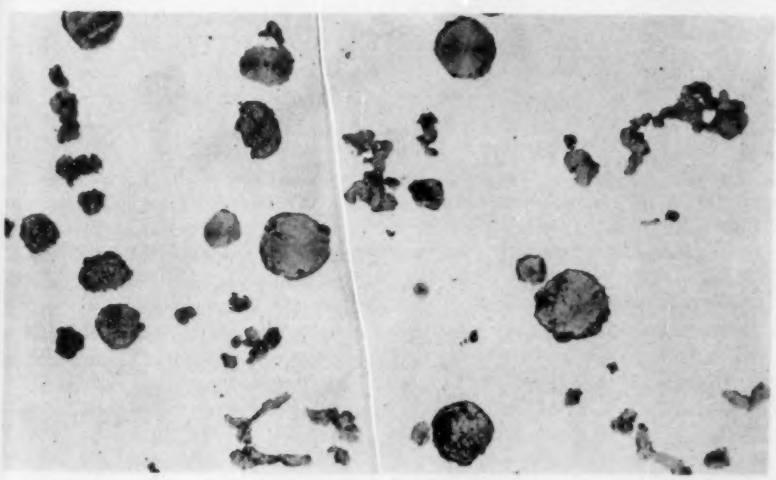


Fig. 7 — Ductile iron with 60 per cent Type I, 40 per cent Type III graphite. Unetched. 200 ×.

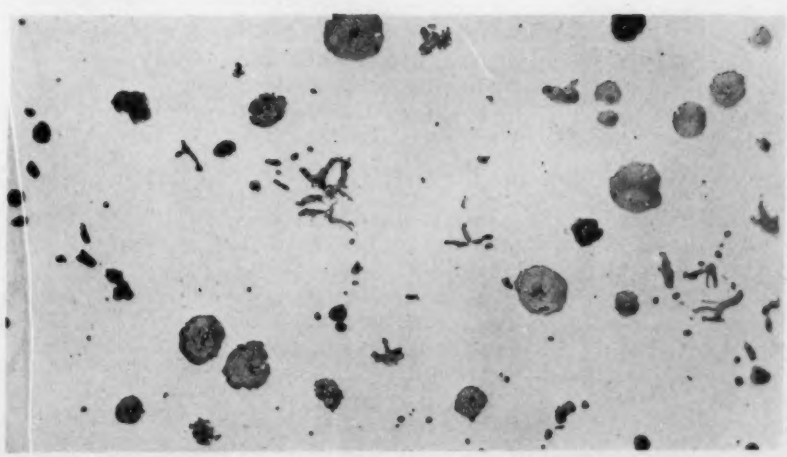


Fig. 8 — Ductile iron with 90 per cent Type I, 10 per cent Type IV graphite. Unetched. 200 ×.

# PLASTIC PATTERNMAKING

## *The polyurethane concept*

by Henry A. Burton


### ABSTRACT

**Throughout the years, patternmaking has remained static in respect to use of wear resistant materials for the foundry industry. In all probability, hard chromium plating is the ultimate in abrasion resistance of patterns and core boxes for high production runs. However, objectionable features of this particular process have been the extremely high cost factor and size limitation, coupled with expensive pattern equipment necessary for the application of hard chromium surfacing. This article outlines and describes methods whereby highly resilient plastics and metals help provide ideal abrasion resistant characteristics and low cost hitherto deemed impossible in processing of foundry pattern equipment.**


### INTRODUCTION

Introduction of the epoxides to the foundry industry has greatly stimulated the imagination in pattern construction particularly where abrasion resistance is concerned, and is principally influenced by the versatility of the materials in liquid form. Complexity of shape or size is still further enhanced by the elimination of machining to the end product. This signifies to some extent that abrasion resistive materials may be introduced to the pattern or core box at indicated or anticipated wear sections with little or no effort on the part of the builder.

Experiences in the past with erosion and destructive action of sand slingers, or erosion caused by the blow-down action in core blowing, point out the folly of subjecting wood patterns or core boxes to such treatment on the basis of medium or heavy production. However, there are at hand today a number of abrasive resistant materials to overcome erosion to some degree.

These abrasion resistant materials take the form of isocyanates, wear resistant epoxy resins, hard chromed plate, etc. These materials over a period of time have proved their worth in combating erosion, not only through the medium of sand slinging or core blowing, but also in other phases of operation throughout the foundry. Erosion to high production metal patterns subject to sand slinging has led to experimenting with materials foreign to normal patternmaking procedure.

#### *Ethers and Esters*

Metallic filled plasticized epoxy resins seem to have been positioned between wood and aluminum in respect to wear resistance of pattern equipment. Synthetic materials are not cure-alls in foundry erosion problems, Yet, in some aspects of patternmaking, they excel over aluminum or other materials in the rigid state. Prolonged evaluation over the course of the past six years has made it abundantly clear that the epoxides surpass aluminum in core box construction material for core blowing. Core boxes with a surface area up to 300 sq in. have produced cores in excess of 30,000, even without erosion in core box cavities opposite to blow down holes of core blowers.

It is indeed reasonable to assume the answer to erosion free plastic pattern equipment lies in the general direction of superior grade plasticizers, providing resilience necessary to overcome impact shock. From all outward appearances, epoxides more or less resemble a rigid material. Nevertheless, barcol impressor testing indicates a hardness within the range of 40-45 barcol. Whereas aluminum, on the other hand, depicts readings in the neighborhood of 70.

This barcol differential should establish epoxy resins as a highly resilient material. It then naturally becomes apparent why synthetics stand up so well under core blowing conditions, with subsequent reduction of wear through resilience.

#### *Polyester Resins*

Little thought had been given in the past to polyester resins, primarily due to their extremely high shrinkage. However, here is a synthetic material which can be of exceptional value and assistance to the patternmaker. Filled polyesters have long found a ready market in the automotive body repair field, principally for their splendid adhesion qualities, coupled with fast curing properties (up to a 10 min cure time).

What makes polyesters so interesting and attractive to the patternmaker stems from the fact that adhesion properties are such as to withstand chipping under sand slinging conditions. Surface preparation includes metal rough up and solvent cleaning as the

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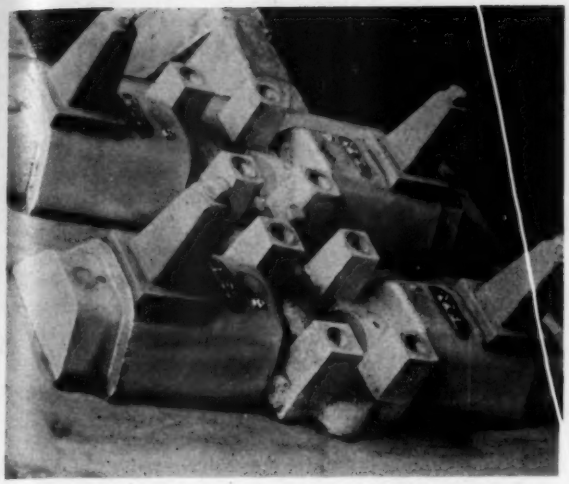


Fig. 1 — Blind heads mounted for high production.

only requirements necessary in obtaining a good bond. Extremely fast curing properties with dimensional stability after cure further enhance filled polyesters in patching and repairing of wood or metal pattern equipment.

All in all, the fact remains there are definite applications for the material, even if material imbalance manifests a high incidence of shrinkage, coupled with the fact they do not adhere to the epoxides.

## INDUSTRIAL RUBBERS

Prior to the entrance of plastics to the foundry scene, some work had been accomplished utilizing industrial rubber as a resilient material in combating wear to pattern equipment. However, it was easily seen the material had limited scope, inasmuch as rubber, in order to fashion a pattern into any desired shape or form, required the employment of expensive die forming equipment. Rubber, therefore, became economically impractical for prototype pattern work.

An outstandingly successful application was found in the shape of blind heads for jolt machine work. The elasticity of the material enabled the use of ram-up cores, virtually eliminating core breakage. Figure 1 shows blind heads mounted for high production. Considering that the pattern equipment has completed over eight years of satisfactory service without wear or further adjustment should demonstrate the fact that elastomeric materials have a value in the foundry.

Another application for elastomeric materials in the shape of rubber open heads is seen in Fig. 2 set up for jolt machine work. In some cases, industrial rubber has been used to advantage in overcoming the wear experienced to heads under slinger operations. There are, of course, limitations in part, such as caused by the flexibility of the material in solid form.

For the want of another term, limitations of materials could aptly be described as "material imbalance." This so called imbalance is frequently encountered in patternmaking materials. A case in point is that of the subject matter under discussion, lowered durometer hardness results in higher abrasion resistance, increasing elasticity and, by the same token, effects greater distortion under impact. On the other hand, increased durometer hardness sharply decreases the abrasion resistance of the material.

Wishful thinking has led to the thought that if any synthetic resilient material could be developed along the lines of ideal abrasive resistant characteristics, such as exemplified by low durometer rubber, it would prove of immense interest and satisfaction to the patternmaker in combating erosion problems.

The arrival of a two component cross linking polyisocayanate castor oil based material, apparently possessing to all appearances the same characteristics as a flexible rubber, suggests this material may be a possible answer to major erosion problems in the foundry.

## SYNTHETIC ELASTOMERS

Isocyanates or diisocyanates, more commonly referred to as polyurethanes or urethanes are gradually making an appearance and slowly gaining acceptance in the foundry.

The history of urethanes is quite interesting. They first appeared well over a hundred years ago. It is only in late years that they have gained any importance commercially. One hears of ethyl urethane in medicine as a producer of sleep. Other applications of solid elastomers are in the form of synthetic rubber tires, bushings, ring seals and the like.

On the other hand, flexible and rigid urethane foams are being applied extensively in aircraft and automobiles as paddings, fillers and so forth. In fact, there are a thousand and one commercial products on the market where they are used to advantage. Tri-

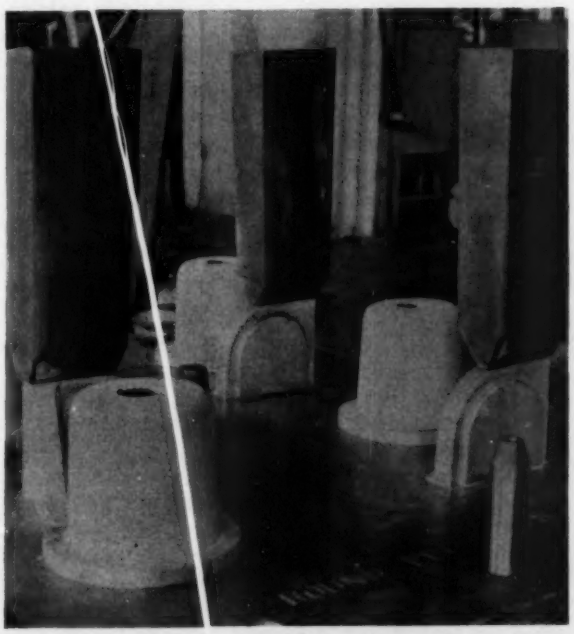


Fig. 2 — Elastomeric materials in the shape of rubber open heads set up for jolt machine work.

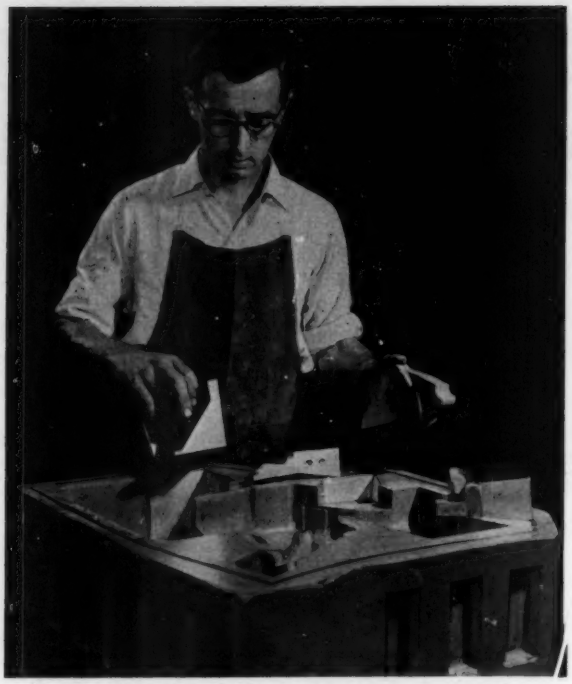


Fig. 3 — Core box showing hard chromium plated facings in cavity and loose pieces to reduce loose piece stick.

isocyanates are used in coatings such as paints and lacquers to provide wear resistance and act as a dirt repellant. It would not be at all surprising if they eventually show up as satisfactory coatings for wood patterns.

The physical properties of urethanes are such they lend themselves quite easily for their ability to withstand heavy impact shock, which makes them particularly interesting and adaptable in the field of abrasion resistance for the foundry.

Investigation and evaluation of the remarkable elastomeric or rubbery like material shows a great deal of promise under test. When applied in gelcote form, modulus of elasticity is so low when backed up by rigid materials, it is of no practical importance nor gives any cause for concern whatsoever. This implies no visual dimensional change under load to the ultimate end product.

### *Abrasion Resistance*

Repeated experimentation with elastomeric polyurethanes as a pattern medium has in all cases manifested superior abrasion resistance over a number of rigid materials in use today. However, this superior resistance to abrasion does not mean the materials are adaptable to all phases of operation in patternmaking. Unfortunately, material imbalance restricts the use of urethanes as a possible solution to extreme wear in quite a number of unusual ways.

Urethanes are hygroscopic in every sense of the word, this condition definitely retards speed in reproduction to a marked degree. It indicates all moisture must be completely eliminated from the form or mold, otherwise the end product will develop surface pin holes. Hygroscopy also shortens storage life. Applied in gelcote form, urethanes require a mechanical other than chemical bond. Method of providing a satisfactory mechanical bond is relatively simple. After the urethane gelcote is applied to the form, flock in the shape of ¼-in. steel or aluminum fibers are scattered over the coating before application of epoxides when laminating for strength. There has been no evidence of bond breakdown by following this particular method, even after the production of 5000 cores. This suggests the method has merit, especially when cores are produced by air ramming.

There is also the question of cure time and sand stick. Urethanes require a 24 hr cure plus additional heat curing to completely eliminate sand stick. All these elements point out there is considerable delay in processing a urethane coated pattern or core box, delay which, in most cases, the foundry cannot afford.

These several points have been brought forward to emphasize the importance in weighing the pros and cons of urethanes as a pattern medium. Prior to the inception of epoxy resins in foundry work, great hopes had been held for phenolformaldehyde, then found wanting, simply because of the fact they would not bond to materials other than themselves. This soured and slowed down the use of resins by a number of companies for a considerable time. This lesson also applies to urethanes, they too should be evaluated only for their own particular requirements.

There is, however, a brighter side in the use of urethanes for the foundry. They show up remarkably well under speed slinging conditions. An interesting study and possible solution to extreme wear is seen on a urethane gelcoted pattern under tests. Seven thousand five hundred molds have been produced up to the moment without any indication of wear, yet on the other hand, epoxy resins have failed to produce more than 300 molds under the same harsh conditions.

In brief, the patternmaker has now under his command a highly elastomeric material in which to combat erosion on foundry pattern equipment, erosion which gives a great deal of concern to all parties involved in expediting the equipment to the foundry.

## CONCLUSION

From observation, evidence points out two factors in erosion resistance. There is on the one side, rigid or hard surfaced materials which are highly resistive to wear, whereas on the other side, there are the resilient materials, which in all probability could surpass any other material in the rigid state to retard or eliminate erosion. It is quite evident that the inception of synthetics to the foundry industry has presented the patternmaker with a much larger and wider scope in attempting to combat erosion through the use of resilient materials, and has been directly instrumental in providing ways and means whereby patternmaking follows or keeps abreast of foundry technology.

# CALCIUM CARBIDE USE IN ACID CUPOLAS

by R. Schulze


## ABSTRACT

**A special calcium carbide having a melting point sufficiently low to melt and burn in the combustion zone of the conventional acid cupola is described. This material has found wide use in West German iron foundries as a cupola addition to raise the temperature of the iron, increase melting rates and permit a reduction in the amount of coke used in the melting process. Lowered coke consumption results in a reduction in the sulfur content of the iron. Typical examples of the use of the special carbide and the results obtained are given.**


## INTRODUCTION

This report is a summary of experience in West Germany since 1954 involving the use of a special calcium carbide of low melting point as a cupola charge addition.

Calcium carbide is a dark gray stone-like substance obtained by fusing coke and lime in an electric furnace. In general appearance and weight it resembles a crushed gray limestone and when added to the cupola charge it is used in similar sizes. The principal use of calcium carbide is for the generation of acetylene, a fuel gas used in welding and cutting of metals, and in the chemical industry as a basic material for the production of many chemicals. The carbide which is used for acetylene production contains approximately 80 per cent calcium carbide ($CaC_2$), the remaining 20 per cent consisting mostly of lime (CaO) plus small amounts of minor impurities.

Calcium carbide of this grade has a melting temperature of approximately 3400 F (1870 C), and cannot melt at temperatures existing in the melting zone of the cold blast acid cupola. The special carbide is of a eutectic composition. It contains approximately 72 per cent calcium carbide, with the balance mostly lime and melts at about 2970 F (1632 C).

The first systematic experiments involving the addition of carbide to cupola furnaces were performed by S. F. Carter.[1] The results of his experiments with the use of commercial calcium carbide (welding carbide) in basic hot blast cupolas were reported at the AFS Annual Meeting in 1950. Carter found that addition of carbide to the basic cupola increased metal temperature and produced iron of lower sulfur content. The excellent results from the addition of carbide to basic cupolas reported by Carter caused several West German foundries to attempt to obtain higher temperature iron and lower sulfur in cold blast cupolas operated with acid slags.

## EXPERIENCE WITH CARBIDE IN THE CUPOLA

H. Timmerbeil[2] has reported several experiments performed at the Vollmarstein Steelworks of Knorr Bremse since 1952. The addition of 2 per cent commercial calcium carbide (welding carbide of 80 per cent $CaC_2$) without change in the coke charge yielded a decided temperature increase. This permitted a reduction in the coke charge of about one-third when melting iron for the side blow converter process and for the production of malleable iron. The iron at the spout was of equal temperature to that obtained by the usual practice involving the use of more coke (without carbide). When the coke charge was lowered, melting capacity increased and the sulfur content of the iron was reduced.

From the fact that the carbide became effective slowly and that unconsumed carbide remained in the cupola drop, Timmerbeil concluded that the normal carbide is not sufficiently reactive (Carter also observed this). To overcome these difficulties the author's company developed a special carbide of lower melting point. This special carbide contains about 72 per cent $CaC_2$ and corresponds to a eutectic in the system $CaC_2$—CaO, illustrated in Fig. 1. The eutectic carbide begins to melt at 2966 F (1630 C), a temperature about 360 F (180 C) lower than for normal calcium carbide.

This special carbide has been used successfully in acid cold blast cupolas at the Volmarstein plant since 1954. In Sept. 1956, it was announced at the International Foundry Convention in Dusseldorf.

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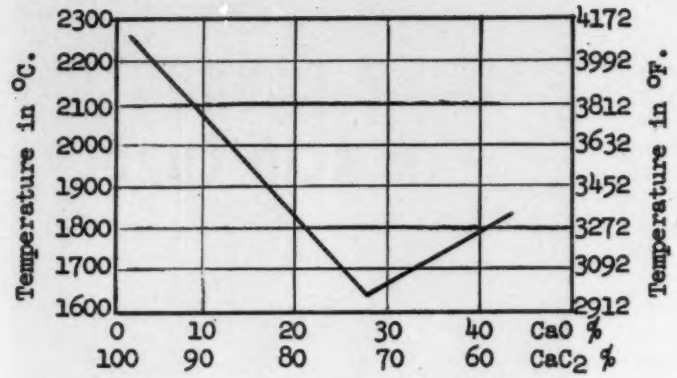



Fig. 1 — Melting temperature in the system $CaC_2$ — CaO (Ruff and Foerster).

Since then calcium carbide has found use in iron foundries of all types. R. Schulze[4] has reported concerning the practical application of calcium carbide and the nature of its reaction in the cupola. W. Patterson, F. Neumann and H. Trump has systematically investigated the influence of calcium carbide in the melting process in an experimental cupola at the Aachen Institute of Technology.

The results have been presented in diagrams showing the relation between coke charge, carbide charge, blast volume, melting capacity and iron temperature. The results reported by F. Neumann[5] confirmed experience gained in actual furnace practice. Figures 2 and 3 illustrate results obtained from the use of calcium carbide in melting gray iron.

It will be noted the consumption of coke for steel-free charges is reduced from 13.5 to 10 per cent when 2 per cent carbide is used. This is sufficient coke to maintain the correct height of the coke bed and the addition of coke booster charges is not necessary. The temperature of the iron at the spout increased 70 to 90 F (21 to 32 C) compared to the practice involving heavier coke consumption but without calcium carbide. The increased melting rate from 4.3 to 5.3 tons/hr should also be noted.

The reactiveness of calcium carbide is evident when a cupola heat is started. When 2 per cent carbide is used the first tap is hotter; 4 per cent carbide on the bed coke and on the first charge results in a rapid temperature rise. The first iron is 140 to 180 F (60 to 82 C) hotter and can be easily cast.

The rapid rise in metal temperature from use of calcium carbide is used to insure a hot start by some foundries who do not employ carbide during the

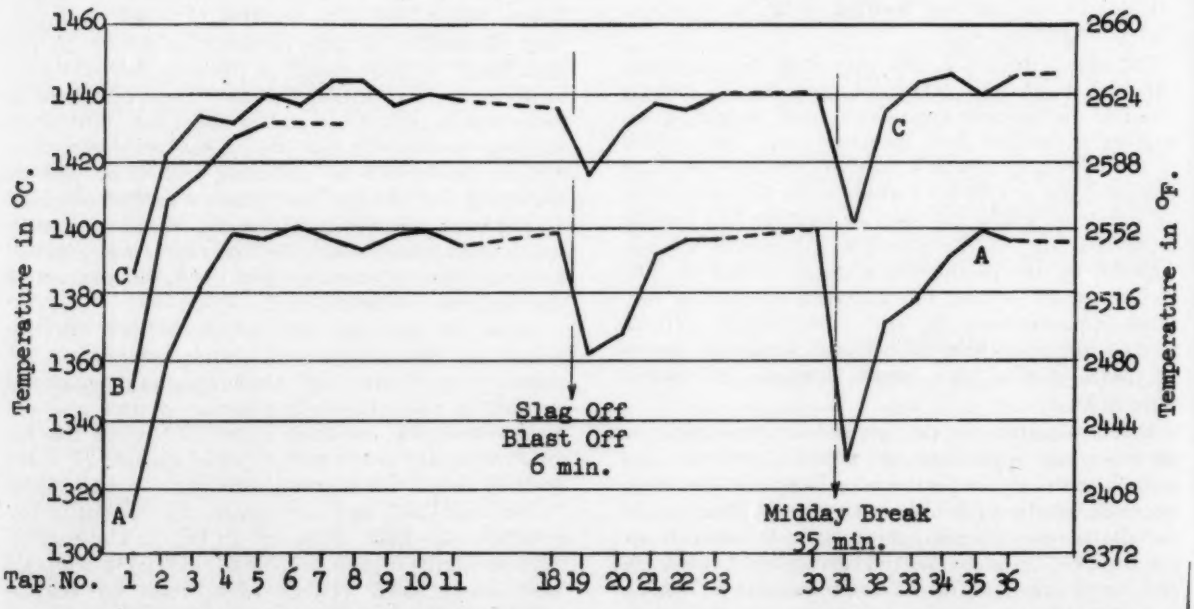



Fig. 2 — A comparison of heats melted with and without carbide using an acid cold blast cupola of 32 in. diameter in continuous production of soft iron. Charge weight 880 lb. A) With 13.5% charge coke, 4% limestone (Taps one and 31 were pigged). B) With 10% coke, 1% stone, 2% calcium carbide. C) With 10% coke, 1% stone, 4% calcium carbide on the bed coke and first iron charge, 2% calcium carbide on the following charges. Melting Capacity: A = 4.3 tons/hr — C = 5.3 tons/hr. Temperature measurements were obtained by optical pyrometer adjusted with an immersion thermocouple. Two readings were taken at the spout per tap.

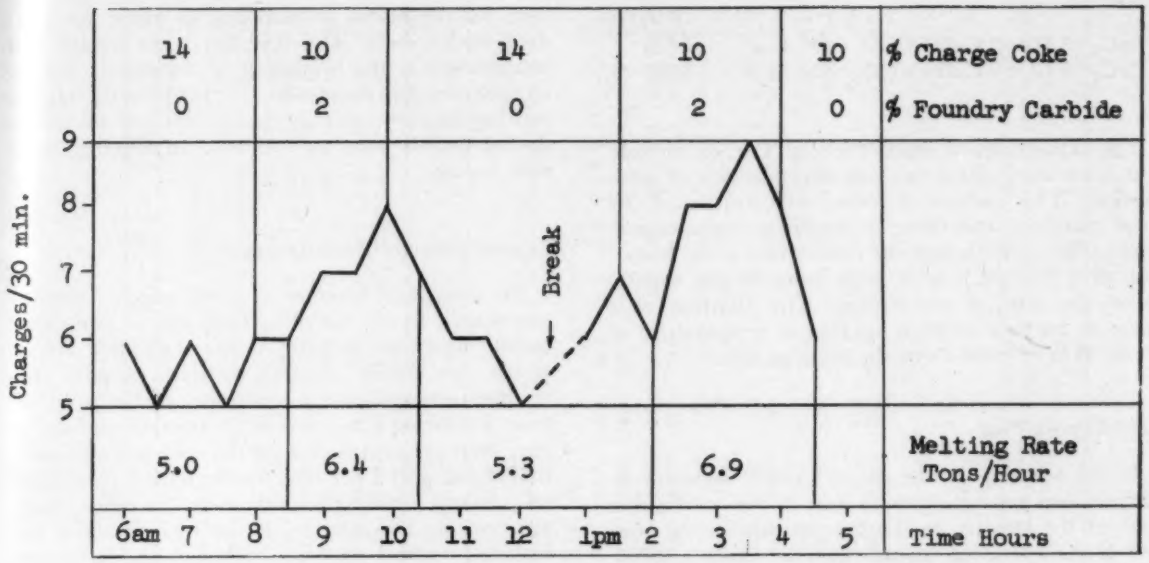



Fig. 3 — Adaptation of melting capacity to the need for iron — cold blast cupola, 35 in. diameter, charge weight 880 lb (A uniform charge with regulation of the silicon content by ladle addition according to the specification of the casting).

balance of the heat. It is also used to compensate for planned shutdowns such as the lunch period. The last charges before the shutdown receive a carbide addition timed so that the carbide just reaches the coke bed when the blast is taken off. Careful handling results in a negligible temperature drop.

It has been shown that an increased melting rate results when carbide is used to permit a reduction in the coke charge. Thus, use of calcium carbide makes it possible to alter the melting capacity of a furnace to meet changing needs of the plant, for instance in plants needing extra iron temporarily for the pouring of small castings during a regular run of heavier work. Iron production can be increased by a simultaneous reduction in the coke charge and a planned addition of carbide. Such flexibility of melting capacity, while retaining satisfactory iron temperature cannot be obtained by any other practice. The same process is also used to shorten the time needed to collect the iron for large castings.

Figure 3 shows the intentional alteration of cupola output by a reduction of the coke charge from 14 to 10 per cent with simultaneous addition of 2 per cent carbide. Average production per hour in the first and third periods was 5.1 tons compared to 6.7 tons per hour during the periods when carbide was used, an increase of 31 per cent in output.

## ADVANTAGES OF EUTECTIC CARBIDE

Foundry experience and laboratory investigations have amply demonstrated that the eutectic calcium carbide of low melting point has the following advantages for use in acid cupolas when compared to regular carbide:

1. A rapid rise of metal temperature as soon as the carbide reaches the combustion zone.
2. 2 per cent carbide produces iron 70 to 90 degrees higher in temperature with the use of less coke.
3. Complete utilization of the carbide with no residue left in the coke bed.

The question immediately comes up as to why the high melting regular carbide is useful in basic cupolas while only low melting carbide is fully efficient in the acid furnace. The explanation lies in the ready solubility of calcium carbide in basic slags. Calcium carbide dissolves easily in basic slags, the solubility improving as the calcium oxide content of the slag is increased. It appears that carbide starts to dissolve as soon as slag is formed in the melting zone of the basic cupola. When the carbide reaches the coke bed it is dissolved by the slag running over it.

### *Metal Oxide Reduction*

The carbide taken up by the slag reduces any metal oxides present and aids in desulfurization of the iron. Carbide which is not needed for this reaction burns in the tuyere zone and contributes the greater part of the temperature rise.

Calcium carbide is not readily soluble in acid slags, and as the slag in an acid cupola cannot dissolve the carbide, material which is in itself sufficiently reactive must be used in these furnaces. The special carbide with low melting point fulfills these requirements as the temperature at which the carbide eutectic melts is in the range reached in the combustion zone of the cold blast acid cupola.

As soon as the particles of carbide reach the coke bed and start to melt, combustion of the carbide begins. The heat generated by this combustion causes the carbide particles to melt rapidly, and the liquid carbide burns vigorously as it descends into the tuyere zone.

The reactions involved are:

| | Btu/Lb $CaC_2$ |
|---|---|
| $2\ CaC_2+3\ O_2=2\ CaO+4\ CO$ −382 kcal. | −5370 |
| $2\ CaC_2+5\ O_2=2\ CaO+4\ CO_2$ −650 kcal. | −9130 |

The calcium oxide enters the slag. Carbon dioxide and monoxide, of course, join the products of combustion. The carbide is completely consumed by these reactions and there is no direct metallurgical effect. The only effect of the combustion is the formation of a shallow zone of high temperature, slightly above the level of the tuyeres. The addition of 2 per cent carbide yields a maximum temperature of about 3275 F (1246 C) in the melting zone.

### *Acid Cupola Fuel*

In the acid cupola, the special carbide serves as an efficient fuel generating extra heat in the combustion zone of the cupola. As the temperature in the coke bed of the operating cupola normally exceeds 3000 F (1649 C), the special low melting carbide melts as it reaches the upper part of the coke bed. The droplets of molten carbide then burn as they pass through the combustion zone, generating extra heat.

As previously mentioned, the main effect of the combustion of carbide is the formation of a shallow zone of elevated temperature slightly above the cross-section at the tuyeres. The increased temperature in the coke bed yields higher iron and slag temperature and an increase in the carbon and silicon contents of the iron. Use of carbide permits a reduction in the coke charge introducing less sulfur and resulting in higher melting rates. Molten calcium oxide formed by combustion of carbide is a highly reactive flux promoting slag fluidity and minimizing difficulties from bridging or other operating troubles.

The experienced cupola operator will recognize at once that many of these results are similar to the benefits obtained by use of preheated air for the cupola blast. Indeed, the use of carbide does permit the operator of a cold blast cupola to obtain most of the advantages inherent in hot blast operation.

Use of carbide in the acid cupola does not result in appreciable desulfurization of the iron as the acid lining does not permit a desulfurizing basic slag. The calcium oxide formed by combustion of the carbide must be allowed for by a reduction in limestone to yield a normal acid slag. As the calcium content is about 50 per cent greater than that of limestone, the usual 2 per cent addition will replace about 3 per cent of stone. A reduction in the coke charge will of course yield iron somewhat lower in sulfur, since the sulfur in the coke is responsible for the increase resulting from remelting in the acid cupola.

## PRACTICAL APPLICATIONS

Following the experience of the previously mentioned plant in the melting of malleable iron and steel castings, the carbide melting process, was rapidly accepted for the melting of gray iron. In West Germany, calcium carbide is today regularly used in more than 150 foundries. About half of these use it in their entire melt. The rest use it to obtain high temperature at the beginning of the cupola heat, to compensate for shutdowns, to temporarily increase melting capacity and in the production of iron of special quality such as that used in making ductile iron castings.

### *Cupola Benefits Determination*

To determine whether a given cupola operation can benefit by the use of carbide, one or two experimental heats are melted. At the start data are collected concerning normal melting capacity, iron temperature and the composition of the iron and slag. Following this, tests with calcium carbide are run. Four per cent is used on the coke bed and on the first charge and 2 per cent for the second through the sixth charge omitting the limestone. No other changes are made in the mixture. If the furnace reacts well, carbide is added without further delay. If the coke charge is to be lowered, regulation of the stone additions and blast are necessary. The coke input is reduced the next melting day, and the blast volume adjusted to obtain the best temperature and melting rate.

### *Carbon and Silicon Adjustment*

In order to adjust carbon and silicon to normal values further corrections are made to the mixtures as indicated by the iron analysis. Generally within two melting days, and without unusual troubles or production loss, sufficient data to indicate the technical advantages as well as preliminary data for economic evaluation of the material are obtained. If a foundry decides to adopt the carbide process, the amount necessary will be determined during the following melting days. The amount of carbide charged depends on the conditions. The average is about 2 per cent, but in some cases this can be reduced to one per cent or increased to 3 per cent. As soon as the optimum conditions for the new melting process are determined, and the charging crew has become accustomed to the changes, melting will proceed as smoothly as before the change.

Table 1 shows the results obtained in nine different plants producing different foundry products. Line 1 in each case indicates the results determined by observation of the original practice. In the case of *A*, lines 2 and 3 list transition values, while line 4 indicates the final results obtained. For *C* an intermediary step is shown. For the rest of the products intermediate values were omitted, line 2 indicating the results obtained after adopting the use of carbide.

### *Benefits From Calcium Carbide*

Table 2 provides an evaluation of the data in Table 1, indicating the various types of benefits obtained from the use of carbide. A check mark appearing under any particular category such as "Higher

TABLE 1 — CHANGE FROM COLD BLAST CUPOLA TO MELTING WITH CALCIUM CARBIDE

| | MIXTURES | | | | | | | | | IRON | | | | | | | | SLAGS | | | | | | | |
|---|---|---|---|---|---|---|---|---|---|---|---|---|---|---|---|---|---|---|---|---|---|---|---|---|---|
| | | | | | | | | Calcium[a] | | | Temp.,[b] | | Analysis, % | | | | | Analysis, % | | | | | | | |
| A-G: Gray Iron For | Cupola Dia., in. | Chg. Wt., Lb | Pig Iron, % | Scrap Iron, % | Steel Scrap, % | Revert Scrap, % | Chg.[a] Coke, % | Carbide, % | Limestone,[a] % | Melting Rate, tons/hr | F | C | C | Si | Mn | P | S | $SiO_2$ | CaO | $Al_2O_3$ | $Fe_2O_3$ | FeO | MnO[c] | S[d] | CaO / $SiO_2$ |
| A. Enameled Ware | 32 | 880 | 40 | 28 | — | 32 | 14 | — | 4 | 4.3 | 2516 | 1380 | 3.32 | 2.75 | 0.52 | 0.80 | 0.132 | 51.9 | 27.3 | 8.1 | 2.2 | 5.9 | 3.4 | 0.13 | 0.52 |
| | 32 | 880 | 40 | 28 | — | 32 | 10 | 2 | 1 | 5.4 | 2606 | 1430 | 3.50 | 2.88 | 0.60 | 0.81 | 0.097 | 53.4 | 28.9 | 8.8 | 0.5 | 3.8 | 3.1 | 0.14 | 0.54 |
| | 32 | 880 | 36[e] | 34 | — | 30 | 10 | 2 | 1 | 5.5 | 2624 | 1440 | 3.47 | 2.69 | 0.58 | 0.75 | 0.108 | 54.5 | 28.6 | 8.2 | 0.6 | 3.5 | 3.3 | 0.12 | 0.52 |
| | 32 | 880 | 36[e] | 34 | — | 30 | 10 | 1.8 | 2 | 5.5 | 2606 | 1430 | 3.45 | 2.70 | 0.53 | 0.77 | 0.103 | 51.0 | 34.3 | 9.0 | 0.2 | 2.3 | 1.9 | 0.30 | 0.67 |
| B. Radiators | 36 | 1100 | 30 | 42 | — | 28 | 15 | — | 5 | 5.0 | 2552 | 1400 | 3.35 | 2.68 | 0.68 | 0.80 | 0.140 | 49.0 | 31.9 | 9.1 | 0.8 | 3.6 | 4.0 | 0.21 | 0.65 |
| | 36 | 1100 | 30[e] | 45 | — | 25 | 11 | 2 | 3 | 6.3 | 2678 | 1470 | 3.48 | 2.61 | 0.66 | 0.83 | 0.106 | 47.2 | 37.0 | 8.6 | 0.0 | 2.9 | 2.4 | 0.29 | 0.78 |
| C. Machine Tools | 40 | 1540 | 35 | 33 | 5 | 27 | 12.5 | — | 5 | 7.2 | 2570 | 1410 | 3.40 | 1.82 | 0.68 | 0.40 | 0.110 | 46.9 | 34.0 | 12.8 | 0.0 | 3.0 | 2.6 | 0.40 | 0.72 |
| | 40 | 1540 | 35 | 33 | 5 | 27 | 10 | 2 | 2 | 8.6 | 2678 | 1470 | 3.53 | 1.90 | 0.70 | 0.39 | 0.090 | 48.1 | 36.1 | 10.6 | 0.0 | 2.7 | 2.0 | 0.44 | 0.75 |
| | 40 | 1540 | 30 | 34 | 12 | 24 | 11 | 1.5 | 2.5 | 7.8 | 2660 | 1460 | 3.44 | 1.85 | 0.63 | 0.35 | 0.095 | 47.2 | 36.6 | 10.9 | 0.0 | 2.5 | 1.7 | 0.41 | 0.77 |
| | 40 | 1540 | 40 | 29 | 10 | 21 | 13 | — | 5 | 7.2 | 2588 | 1420 | 3.23 | 1.55 | 0.80 | 0.27 | 0.113 | | | | | | | | |
| | 40 | 1540 | 40 | 29 | 10 | 21 | 10.5 | 2 | 2 | 8.3 | 2678 | 1470 | 3.41 | 1.66 | 0.88 | 0.30 | 0.105 | | | | | | | | |
| | 40 | 1540 | 35 | 27 | 18 | 20 | 11 | 1.5 | 2.5 | 7.9 | 2642 | 1450 | 3.21 | 1.52 | 0.85 | 0.25 | 0.100 | | | | | | | | |
| | 40 | 1540 | 40 | 30 | 20 | 10 | 13 | — | 5 | 7.2 | 2588 | 1420 | 3.05 | 1.37 | 0.88 | 0.22 | 0.108 | | | | | | | | |
| | 40 | 1540 | 40 | 30 | 20 | 10 | 11 | 2 | 2 | 7.9 | 2642 | 1450 | 3.18 | 1.40 | 0.90 | 0.22 | 0.098 | | | | | | | | |
| | 40 | 1540 | 32 | 28 | 30 | 10 | 11.5 | 2 | 2 | 7.7 | 2660 | 1460 | 3.10 | 1.29 | 0.91 | 0.18 | 0.101 | | | | | | | | |
| D. Automobile Cylinders | 40 | 1320 | 50 | 10 | 5 | 35 | 13 | — | 6 | 6.8 | 2588 | 1420 | 3.20 | 2.18 | 0.74 | 0.40 | 0.110 | 44.3 | 36.9 | 10.5 | 0.0 | 3.1 | 2.8 | 0.30 | 0.83 |
| | 40 | 1320 | 40 | 12 | 15 | 33 | 11 | 2 | 3 | 7.6 | 2678 | 1470 | 3.25 | 2.15 | 0.72 | 0.25 | 0.089 | 46.6 | 37.5 | 8.2 | 0.0 | 2.9 | 2.3 | 0.32 | 0.80 |
| E. Centrifugal Pipe | 32 | 1100 | 50 | 20 | — | 30 | 13.5 | — | 4[h] | 4.4 | 2552 | 1400 | 3.32 | 2.06 | 0.78 | 0.40 | 0.103 | 46.1 | 31.8 | 14.0 | 0.3 | 2.6 | 3.2 | 0.38 | 0.69 |
| | 28 | 880 | 45[e] | 27 | — | 28 | 10 | 2 | 2[h] | 4.5 | 2642 | 1450 | 3.49 | 2.01 | 0.81 | 0.43 | 0.072 | 44.7 | 35.6 | 13.3 | 0.0 | 2.1 | 1.8 | 0.53 | 0.80 |
| F. Piston Rings | 28 | 660 | 60 | — | — | 40 | 16 | — | 5[h] | 3.1 | 2588 | 1420 | 3.52 | 2.10 | 0.72 | 0.62 | 0.090 | 42.1 | 34.4 | 17.5 | 0.0 | 2.0 | 1.6 | 0.50 | 0.79 |
| | 24 | 550 | 60[e] | — | — | 40 | 11 | 3 | 2[h] | 3.2 | 2714 | 1490 | 3.77 | 2.16 | 0.75 | 0.63 | 0.062 | 42.0 | 37.3 | 15.8 | 0.0 | 1.2 | 1.3 | 0.55 | 0.89 |
| G. Ingot Molds | 52 | 2200 | 40[f] | 60[g] | — | — | 12.5 | — | 3 | 12.7 | 2516 | 1380 | 3.53 | 1.80 | 0.48 | 0.08 | 0.095 | 50.2 | 29.0 | 7.9 | 1.3 | 6.0 | 3.8 | 0.20 | 0.58 |
| | 52 | 2200 | 40[f] | 60[g] | — | — | 10.5 | 1 | 2.5 | 14.3 | 2534 | 1390 | 3.60 | 1.85 | 0.46 | 0.08 | 0.081 | 49.7 | 34.4 | 7.4 | 0.5 | 2.9 | 2.1 | 0.32 | 0.71 |
| H. Malleable Iron Fittings | 36 | 1320 | 30 | — | 25 | 45 | 18 | — | 3[i] | 4.6 | 2570 | 1410 | 2.95 | 0.63 | 0.45 | 0.07 | 0.22 | 52.3 | 26.6 | 9.3 | 1.5 | 5.0 | 2.8 | 0.12 | 0.51 |
| | 36 | 1320 | 20 | — | 35 | 45 | 12.5 | 2.5 | —[i] | 6.6 | 2696 | 1480 | 3.02 | 0.65 | 0.42 | 0.06 | 0.20 | 52.8 | 31.2 | 7.3 | 1.1 | 4.2 | 2.7 | 0.15 | 0.59 |
| J. Steel Castings | 40 | 1760 | 35[f] | — | 25 | 40 | 19 | — | 7 | 5.0 | 2588 | 1420 | 3.10 | 1.08 | 0.40 | 0.05 | 0.14 | 51.1 | 28.5 | 11.0 | 0.3 | 5.3 | 3.5 | 0.14 | 0.52 |
| Premelted Iron For Converter | 40 | 1760 | 20[f] | — | 40 | 40 | 13 | 2 | 5 | 6.9 | 2678 | 1470 | 3.18 | 1.20 | 0.45 | 0.05 | 0.09 | 50.5 | 31.8 | 9.5 | 0.3 | 3.2 | 3.0 | 0.20 | 0.63 |

NOTES:

a) % of iron charge = 100 (West German Foundry Coke with one % sulfur).
b) Temperature: Daily average.
c) The valences of the oxygen were not determined.
d) The content of sulfate in the slags is low (rarely above 0.1% $SO_3$).
e) To correct the increasing silicon content of the iron a portion of the pig iron charged was low silicon.
f) Exclusively Hematite pig iron.
g) Broken ingot molds.
h) A Bauxite addition increases the $Al_2O_3$ content of the slag to improve its fluidity.
i) Calcium sulfate is added to increase the sulfur content of the iron and to improve machinability (treating and tapping).

TABLE 2 — EVALUATION OF CARBIDE EFFECTS FROM TABLE 1

| A-G: Gray Iron For | Higher Temp. | Lower Coke | Higher Tons /Hr | Iron Lower S | Iron Higher C | More Steel |
|---|---|---|---|---|---|---|
| A. Enameled Ware | √ | √ | √ | √ | √ | — |
| B. Radiators | √ | √ | √ | √ | √ | — |
| C. Machine Tools | √ | √ | √ | √ | → | √ |
| D. Automobile Cylinders | √ | √ | √ | √ | → | √ |
| E. Centrifugal Pipe | √ | √ | — | √ | √ | — |
| F. Piston Rings | √ | √ | — | √ | √ | — |
| G. Ingot Molds | — | √ | √ | √ | — | — |
| H. Malleable Iron | √ | √ | √ | — | → | √ |
| J. Steel Castings | √ | √ | √ | √ | → | √ |

Temperature" or "Lower Coke" indicates benefit from the use of carbide.

When difficulties are experienced with the iron, the use of carbide will always be helpful. Higher carbon and increased temperature improve fluidity, while increased melting capacity and lower sulfur result from the lower coke consumption. Experience of several years shows that most cold blast cupolas melting steel-free mixtures produce excellent iron with 10 per cent coke and 2 per cent carbide.

### *Cost Advantage*

The carbide melting process has rapidly become popular because of its cost advantages. It is difficult to set up rules for determining the savings obtained from use of calcium carbide, as the advantages which can be utilized depend somewhat on the conditions in an individual plant. Most calcium carbide is used in producing gray iron castings, a field in which costs are extremely important and it has been found the advantages invariably outweigh the costs involved.

The lower overall costs resulting from use of carbide come from:

a. Improved Iron Quality. Better fluidity results from higher temperature, increased carbon and lower sulfur, therefore less rejects.
b. Increased Melting Capacity. Shorter heats and elimination of overtime work.
c. Reduced Coke Consumption.
d. More Economical Mixtures. Less pig iron and more iron or steel scrap can often be used.

## CONCLUSIONS

Since 1956 a special calcium carbide of eutectic composition containing roughly 70 per cent carbide and 30 per cent lime has found wide use in Germany as an addition to acid cold blast cupolas. This material melts at about 2970 F (1632 C), a temperature existing in the melting zone of the average cupola, and burns in the coke bed causing a shallow zone of high temperature directly above the tuyeres.

Use of the special carbide in the amount of about 2 per cent of the iron charge yields increased metal temperature, higher carbon in the iron and reduced oxidation of silicon. Use of carbide usually permits lowering the coke charge resulting in a decided increase in melting rate and a reduction in the sulfur content of the iron.

Carbide is used by some foundries to increase the temperature of the first iron at the beginning of the cupola heat, to minimize temperature loss resulting from planned shutdowns and to regulate the melting rate of the cupola to meet varying demands for metal.

Depending on local conditions in individual foundries, the use of carbide permits savings from a reduction in coke usage, elimination of delays and overtime wages and use of more economical mixtures containing less pig iron and increased amounts of iron or steel scrap.

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# BRONZE VALVE VERTICAL GATING IN SHELL MOLDS

by F. E. Murphy, G. J. Jackson and R. A. Rosenberg


## ABSTRACT

**A gating system has been devised for casting 85-5-5-5 and *M* Metal pressure containing parts in shell molds. The gating system was developed by experimentation with mold halves covered with glass; the system indicated as best by this test was then used in regular molds under rigorous inspection and constant monitoring of melting and pouring conditions.**

**The gating system thus developed is one of the saxophone type, with a well at the bottom of a tapered sprue and enlarged upgates to reverse the direction of flow. Provisions are made for the use of a strainer core, and storage heads are added to compensate for deformation of the mold. The efficiency of the system has been demonstrated in shop use by a rejection rate less than 3.0 per cent.**


## INTRODUCTION

This paper comprises ten or more years experience in producing bronze castings in shell molds at the authors' company. It is not the purpose of the authors to defend or recommend vertical rather than horizontal gating for shell molds. Each method has its shortcomings and advantages, which are dependent on requirements differing for individual foundry operations.

In the vertical pouring of brass and bronze alloys in shell molds, nearly all the difficulties associated with the production of smooth, pressure tight, and dimensionally accurate castings can be attributed to improper gating. For this discussion, gating includes the arrangement of the patterns on the plate, and the sizes and shapes of the pouring basin, downgates, runners, upgates and ingates.

## LITERATURE REVIEW

### *Early Practical Experiences in Gating Shell Molds*

Initial training of the authors in the shell molding was under the supervision of the inventor of the process, Dr. Johannes Croning. Quite naturally, he exerted great influence on the design of the first shell mold gating systems at the authors' company. Up to that time shell mold gating practices in Europe had not been adopted to the casting of pressure vessels. Experience quickly revealed that casting pressure tight bronze products presented its own peculiar problem. Also, years of practical experience in horizontal gating for 85-5-5-5 and composition *M* Metal in green sand molds influenced us to choose this method for our initial gating systems. However, with the passage of time, vertical gating was developed and adopted as standard practice; this greatly reduced losses to where scrap can now be maintained at less than a 3.0 per cent level.

### *Experimental Work on Vertical Gating Using Regular Molds*

The first castings poured in vertical shell molds were gated directly from the downgate. Figure 1 shows a typical arrangement with simple pouring cup and untapered cylindrical sprue from which the castings filled through reversed ingates. Jets of metal entered the casting cavity at high velocity, penetrating the cores and mold walls, causing poor surface finish and producing castings which, in many cases, were inferior to those made in green sand.

By examining Fig. 1, there are a number of things seen to be obviously wrong. First, the shape of the pouring cup introduces a swirling action to the metal, generating a vortex; the cup is too shallow to prevent splashing, and is not effective in retaining slag. The initial corrective action was to enlarge the basin and make it more cup shaped, and to reduce the outlet. However, these changes made the swirling action worse, while the restricted bottom orifice increased the velocity of the metal and drew slag down into the gate.

In an effort to control or nullify the swirling action introduced by the pouring basin, the downgate was changed from a circular to a rectangular cross-section. However, since the falling metal stream did not immediately fill the basin and so contact the sidewalls, before filling the downgate, the swirling action was not checked soon enough to prevent some slag entering the mold.

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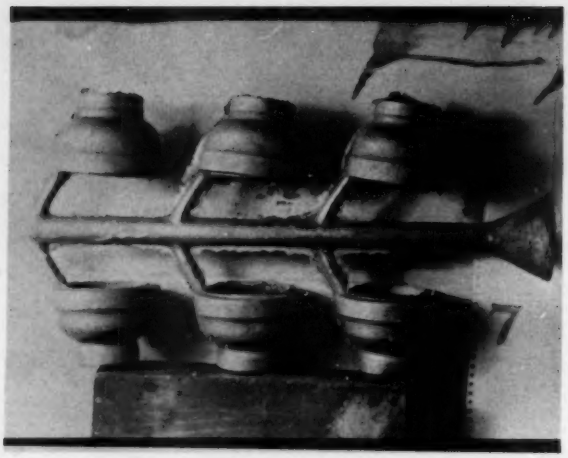


Fig. 1 — One of the early gating systems. The cone shaped basin and untapered cylindrical downgate did not control the metal. Ingates directly from the sprue caused erosion, as noted.

### *Minimizing Velocity of Metal*

To minimize the effect of velocity, the ingates were tilted upwards to form a *Y* shaped arrangement, but when a number of molds were poured from a single ladle, many of the castings contained shrinkage. Next, lateral runners were used with bottom ingates. Since the ingates were at the lowest part of the casting and were usually tapered from a rather heavy runner, there was considerable jetting action with resultant turbulence. This has been fully described in the literature.[1] At one time, it was felt that the jetting action was due to the restriction of the ingate. Therefore, a ball expansion chamber was inserted between the runner and the casting. This tended to lessen but not eliminate the turbulence.

However, by increasing the mass of liquid metal below the casting shrinkage developed at the ingate area. A broad, thin ingate, approaching the dimensions of a kiss gate, was tried on the premise it would solidify before the casting, but this also proved unsatisfactory. It was concluded that gating shell molded castings from the bottom did not ensure a pressure tight casting.

The next step was to gate from the top so that a heavy lateral runner would act as a riser to the casting. However, the turbulence generated as the bronze alloy cascaded around the core was most harmful. The erosive effect on the core was intensified, and the displaced air, seeking to escape through the top ingate, was readily ingested into the stream and remained entrapped in the solidified casting; so this method was rejected.

Research being performed at the time showed the advantage of using a tapered sprue. Swift et al[2] noted that "The sprue cross-section should decrease in proportion to the increase in velocity so that no separation of the liquid from the sprue wall and no suction or gas aspiration tends to occur." Investigations confirmed the efficacy of matching the taper of the downgate to the natural shape of the falling stream. By confining the metal in a form-fitting tube, turbulence was lessened and the aspiration of air into the stream was prevented. The rectangular cross-section was retained in the tapered sprue.

The use of the tapered section demanded a choke at the base to insure that the downgate remained full and to prevent the ingestion of air through the pouring basin. It also dictated the size and shape of the pouring basin and, since we now had a pressurized system, prevented ingating directly from the downgate.

This led to the development of a "downgate-upgate" or reverse flow system ("Saxophone" gates) where the forces generated in the falling metal could be controlled or dissipated by reversing the direction of flow, and where the static head of the upgate would gradually compensate for the metal in the downgate, thus resulting in quiet, unpressurized metal entering the mold cavities. This system has the disadvantage of requiring a larger ratio of gate to product than

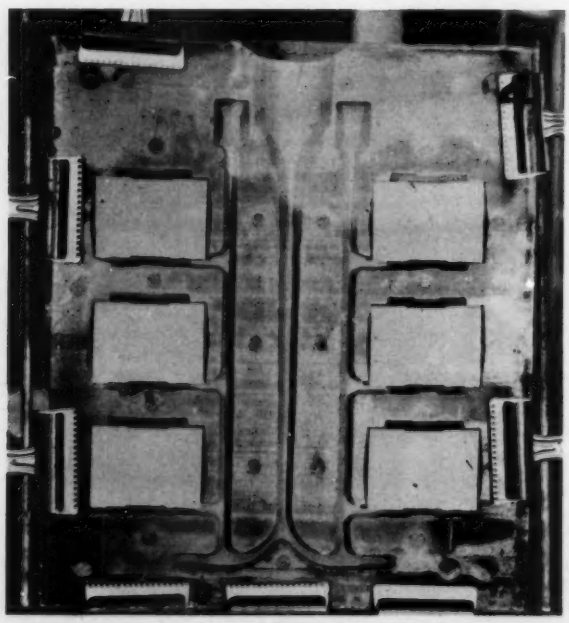


Fig. 2 — Intermediate design of the reverse flow system, shown ready for pouring study. Note the streamlined construction of the gates and the generous radius at all corners.

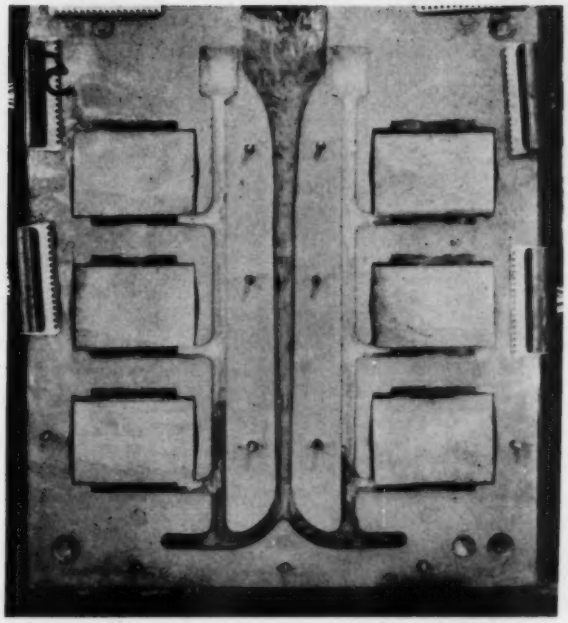


Fig. 3 — The glass faced mold without a strainer core is shown ⅓ full of water. Turbulence; the unequal levels of liquid in the upgates are evident.

the direct flow system. Although the yield was lower, so was the scrap, which meant a net gain. Figure 2 shows one form of this reverse flow system.

### *Experimental Studies Using Glass-Covered Mold Holders and Water*

These gating design studies were facilitated by observing the flow characteristics when a glass plate or plastic sheet was clamped tightly against the face of a half mold. Cores were cut in half along the plane of the mold parting line and coated with a moisture resistant film. The face of the mold was greased to ensure intimate contact with the glass, care being taken so that the mold surface of the ingates, runners and cavities retained their permeability. The mold was placed in a vertical position and water was poured into the basin. The flow was readily observable through the glass and the defects of the system could be examined.

Minor corrective changes could be made immediately by filing away parts of the mold or by building up areas with modeling clay. High speed motion pictures aided in the interpretation and recording of the flow characteristics of the various gating systems.

Although the work of Swift et al[2] was performed on horizontal gating, their basic principles apply equally well to vertical gating. They found that water gave a satisfactory approximation of the flow characteristics of molten bronze. Admittedly, water is not a substitute for molten metal and, in the case of drossy type alloys, this method will not be completely satisfactory. However, dye can be introduced into the water to show the presence of static areas, regions of turbulence or the generation of eddy currents in the stream, and to check the effectiveness of corrective changes. The introduction of solids of varying types, shapes and densities into the water simulated the action of dross, slag or sand in the system and, for this application, indicated the need for a strainer core.

Under the best melting and ladling conditions, nonferrous metal will contain some dross. As the pouring ladle is emptied, the proportion of dross to liquid increases greatly, particularly where 12 to 15 molds are poured from one ladle and the ladle is tipped back and forth a number of times. Even with continual skimming of the metal some dross is usually missed. Besides the efficient skimming action of the strainer core it also minimizes somewhat the variation in velocity head caused by pouring from various heights as the ladle is emptied. The bottom castings in a vertical mold are not as smooth as the upper castings. As the metal head increases, penetration increases. To partially offset this, the lip of the ladle should be kept as close to the pouring basin as possible, which will minimize the variation in pressure head.

### *Strainer Core*

It is easy to perform all of the proper operations in the laboratory, but not so easy to adapt all of them to production. The strainer core was one extremely helpful item that was fully practical and helped considerably in off-setting some of these difficulties. Logically, any strainer should be placed at the base of the sprue, but space limitations on the pattern plate plus the additional labor and care to insert a strainer during mold assembly forced a compromise, and the strainer was located at the bottom of the pouring cup. The cup was deepened and made with a shoulder into which a strainer core could be dropped into place in the assembled mold at the pouring station.

The conventional strainer with round holes divided the stream into a number of jets, which caused considerable turbulence in the downgate. The strainer was redesigned with narrow, concentric slots which delivered sheets of molten metal capable of reblending with a minimum of agitation.

As noted, streamlining the system, particularly at the base of the downgate, had lessened turbulence but had increased the velocity of the metal. The metal was driven into the upgates to a height much above the level in the cavities, often to the extent that upper castings started to fill sooner and at a slower rate than had been anticipated. Note in the photographs of Figs. 3 and 4 the turbulence and entrained air in the gating system and the agitation

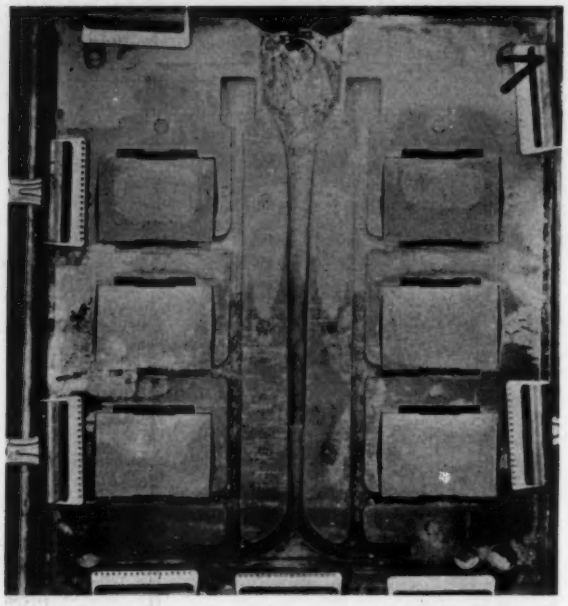


Fig. 4 — Seconds later, with the molded half filled; agitation and entrainment of air has increased and the turbulence within the casting cavity is noticeable.

within the casting cavity. This caused cold shuts or misruns, particularly if the pressure head varied due to erratic or intermittent pouring.

The choke at the base of the downgate was designed to be small enough to control the flow into the bottom castings when there is, at first, differential pressure on the upstream side, and yet be large enough to ensure full flow in the later phase when the weight of metal in the upgate counteracted the downgate pressure. A formula was devised as a guide in the laboratory for choking the stream, but many of our pattern plates have chokes which were altered empirically in the foundry. Figure 5 shows the chart based on the rate of flow in lb/sec for various choke areas. From production records, weights of castings and time studies, a good average pouring time was obtained. Hence it was easy to predict a desirable choke area. Figure 6 is a chart of the rate of rise through the same choke areas, and was used to refine the above determination as the number and size of pieces in a mold varied.

### *Well Development*

Other studies aided in developing a well at the base of the downgate. Unfortunately, there was insufficient layout space on the pattern plate for a well of the dimensions recommended. The glass faced mold was valuable in arriving at a design, which for most patterns completely fulfills requirements. Minor adjustments were necessary on certain atypical layouts, but in each case these were checked out by studying a water flow pattern. The basic form is a 1.0 in. diameter cylinder with the top slightly domed to prevent entrapment of air. The lateral runners are placed slightly below the center of the chamber so there is a definite well below them.

It must be understood there will be areas of turbulence and air bubbles in the water which flows first through the system. No way was found to eliminate this. The measure of the effectiveness of good design in the gates is the speed with which this initial turbulence is purged from the liquid. If the metal following is quiet and free of air, the castings will most likely be sound. At the end of each bottom runner an enlarged pocket is provided to entrap and hold the first metal.

Most jetting action takes place as the metal enters the upgates. To control this condition the upgates are enlarged in cross-sectional areas. The ingates are nearly horizontal to prevent gravity flow of metal from the castings after the pressure surge of pouring has ceased. In order to prevent the ingates from freezing too quickly, they are enlarged as much as possible, in some cases the corner is rounded off the core print at the ingate to increase the mass of metal at this point.

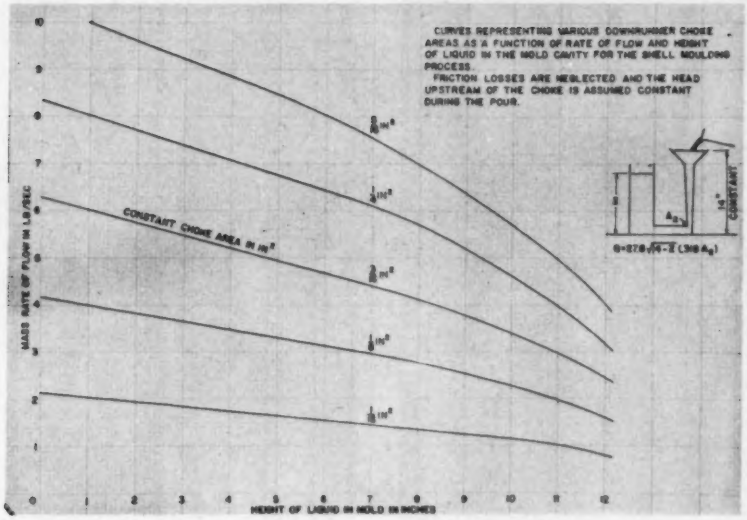
The series of photographs, Figs. 7 through 12, show water flow patterns in this improved system. Figures 7 and 8 illustrate how the system, without a strainer core, is purged of initial turbulence and remains free of agitation or inspiration of air. Figures 9 and 10 show that although the slotted strainer generates some turbulence, the design of the well at the sprue base has quieted the metal. Figures 11 and 12 demonstrate that even a properly designed well cannot purge air entrainment from the system caused by the round holes of conventional strainers.

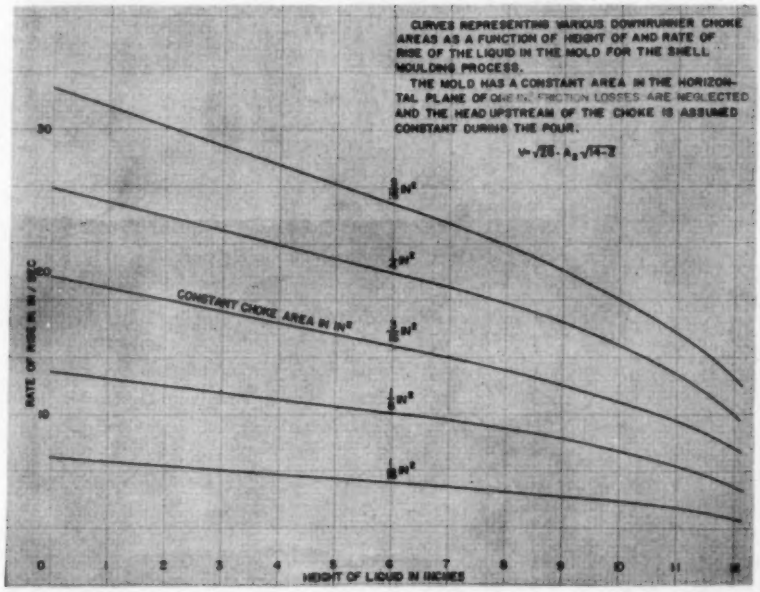
This series of photographs also indicates that if clean metal can be assured, there are flow advantages in pouring without a strainer, but if a strainer core must be used, the slotted type is more efficient than the round hole type.

### *Use of Storage Heads to Overcome Casting Shrinkage Due to Mold Deformation*

Having developed the gating system to this point, despite precautions, a slight deflection in the mold wall, caused by softening of the resin under the heat of the molten metal, resulted in shrinkage of the casting. This hot deformation took place after pouring had stopped, and caused an increase in the size of the casting and volume of the cavity at a time when the gating system had begun to freeze and no liquid was available for feeding.

To overcome this deficiency, a storage head was placed at the top of the upgates and the metal was allowed to continue passing through the gating system after all the cavities had been filled. A "rule of




**Figs. 5 and 6 — Graphs used in predicting desirable choke areas in laying out gating for laboratory evaluation.**

thumb" was used for the amount of metal evacuated into the storage head; this was simply an amount of metal equal to that in the upgates. This accomplished three things:

1) It kept pressure on the ingates.
2) Prevented the upgates from freezing, shrinking and forming shrinkage voids which would tend to attract liquid metal into the vicinity.
3) Made certain that the metal in the gates was hotter than that in the casting, thus promoting directional solidification.

It is to be noted that these are not shrink heads, and although their presence achieves the same results as a shrink head — that is, a sound casting — their action is not wholly that of riser.

### *Adapting Experimental Results and Practice*

While some work was performed on calculating the dimensions of downgates, runners, upgates and ingates, the necessity for getting patterns into production has kept this work to a minimum and no formal study was recorded. Ratio gating was used with the information available in the literature serving as a guide. Fortunately, the gating system developed for one item served as the prototype for many others.

The reverse flow system or "Saxophone Gating" has been in use for many years on production items in the foundry, and has been proved valuable. The 3.0 per cent scrap rate has been consistently lower than with other systems.

Studies are still being performed to improve the design, and any new items introduced into shell

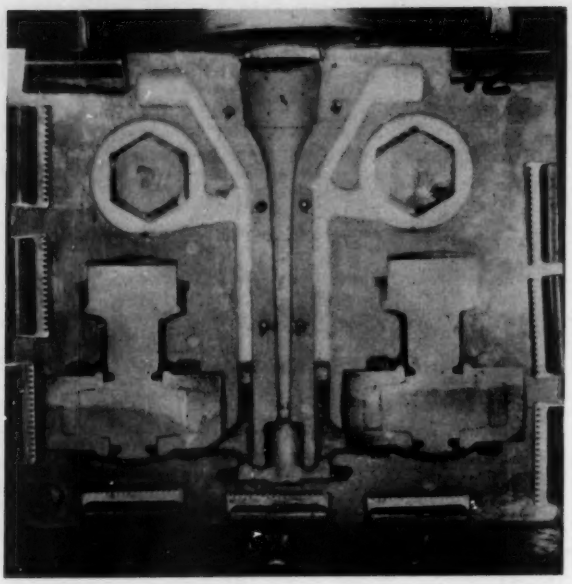


Fig. 7 — Revised gating, incorporating a sprue-base well, shown without a strainer core and ⅓ full of water. Note the absence of air and agitation except in the first liquid in the upgates.

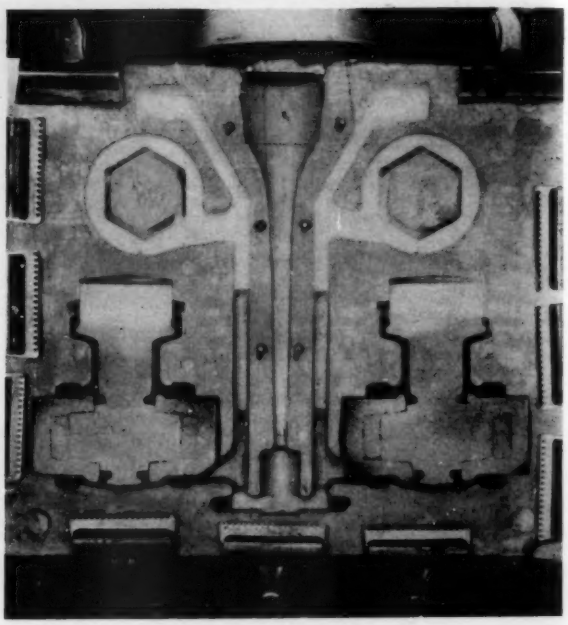


Fig. 8 — Same as Fig. 7 except with the mold half filled. The effective purging action of the well is evident.

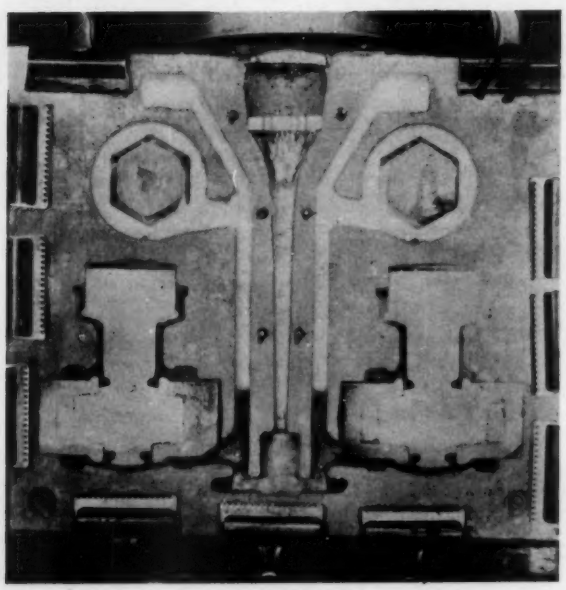


Fig. 9 — Compare with Fig. 7. The addition of a slotted strainer core has slightly increased the initial agitation.

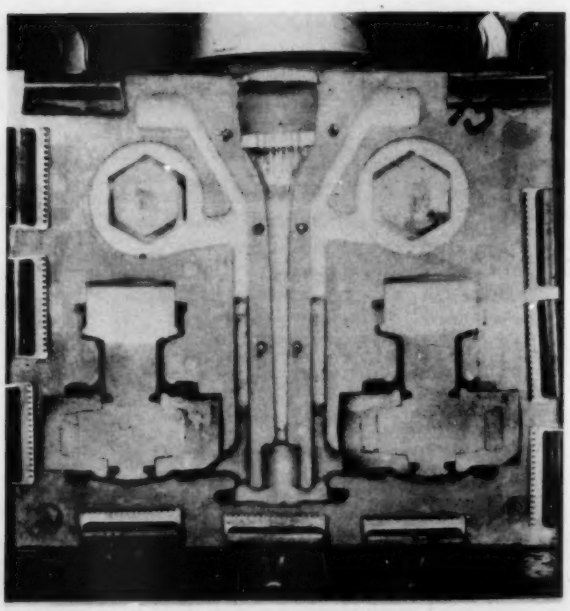
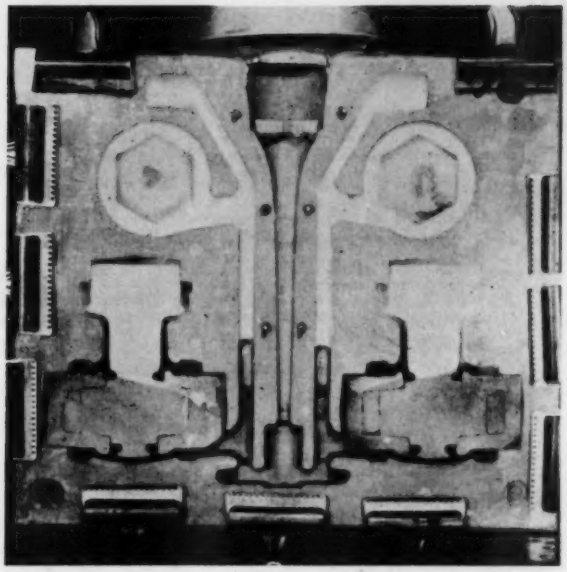


Fig. 10 — As pouring continues the sprue-base well has quieted the agitation introduced by the slotted strainer core.

Fig. 11 — The use of a round hole strainer makes little change in the initial flow of liquid into the mold.



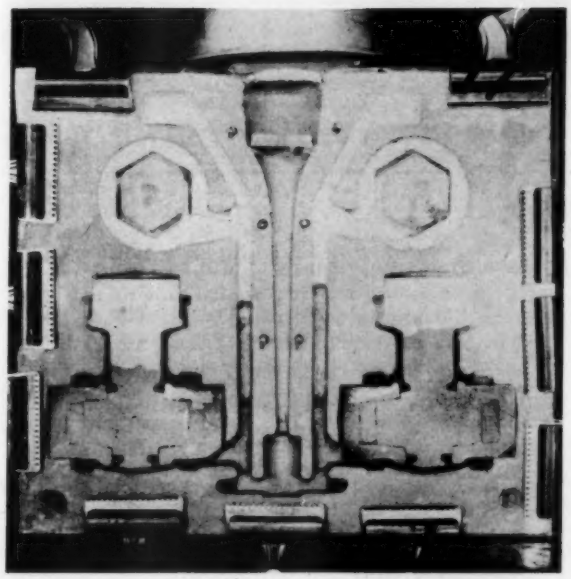


Fig. 12 — In the later stage of pouring the fine bubbles of entrained air generated by the multiple jets of the round hole strainer are still noticeable in the liquid.

molds are studied by glass faced shell molds using water-flow analysis in the laboratory before release to production. Regular spot checks on production castings are made by fracture tests and by radiographic analysis; also close control is maintained on melting and pouring conditions and on pouring temperatures since, without controls in these areas, no gating system can function properly.

## SUMMARY

The results of the experimental work reported herein, and of ten or more years practical experience on gating design for shell mold castings of 85-5-5-5 and *M* Metal at the authors' company are:

1. Vertical gating has proved vastly superior to horizontal gating for making sound, clean, pressure tight castings.
2. The "Saxophone type" gate with proper shape and dimensioning of sprue-cup, downgate, catch basin, upgate, ingates and storage head can reduce the incidence of shop scrap to less than 3.0 per cent.
3. Vertical shell molds undergo an expansion due to softening of the resin under the added pressure of extra metal height, which causes a slight expansion of castings and thus increased shrinkage; this can be overcome by use of a storage head.
4. The use of half molds covered with a glass plate into which water is poured is an excellent means for studying gating systems.
5. Adaptation of the experimental work to practice proves the authenticity of the results obtained in the laboratory.

## ACKNOWLEDGMENTS

The authors wish to express their sincere appreciation to the many people who participated in the development of the shell mold program over the years at the Walworth Co. They wish to thank Professor H. F. Taylor for reviewing this work and for his comments and criticism. Mr. D. Winslow aided in making illustrations and Miss R. Nelson prepared the drafts.

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# CALCINED FLUID COKE AS NEW MOLDING MEDIUM

## *Preliminary evaluation*

by E. G. Gentry and C. L. Lear


### ABSTRACT

**A grain-shaped composition of high carbon content, generally referred to as fluid coke, has been produced by a new petroleum refining process. The author describes a preliminary investigation of the use of calcined fluid coke as a new molding medium,* replacing sand in green and baked molds, baked cores and shell molds and as the principal ingredient of a mold and core wash. Gray iron, bronze and aluminum castings of superior surface smoothness and grain structure have been cast in calcined fluid coke. The inert nature and low thermal expansion of calcined fluid coke are principal properties credited for advantages shown over sand.**


### INTRODUCTION

Foundrymen have long appreciated the desirable characteristics of carbonaceous materials as ingredients of mold and core mixtures and in mold and core washes. The addition of sea coal to molding mixtures is universally practiced. Certain proprietary products of high carbon content are used as coatings for molding sand. Pulverized carbon and graphite may be incorporated into mold and core mixtures or applied as a wash. Solid carbon blocks have been machined to form semi-permanent molds. Some work has been done with crushed foundry coke as a molding material.**

In most of all experiences with materials of high carbon content in casting operations, three features are outstanding:

1. Carbon establishes a reducing atmosphere in the mold.
2. Carbon contributes lower thermal expansion to the mold mixture.
3. Carbon resists wetting by molten metal and tends to resist metal penetration. This paper describes experimental work carried out with a material which could be loosely defined as a carbon sand, and which differs substantially from previously known carbon sands by having higher real density, higher grain hardness and better grain shape.

A petroleum refining process has been developed by a research and engineering company whereby residual petroleum oils are further refined to yield gasoline and coke. This refining process is known as the Fluid Coking Process, the term being descriptive of the use of fluidized solids in the process, and the coke so produced is generally referred to as fluid coke.

Fluid coke, as produced at the refinery, is essentially carbon, consisting of hard, round to subangular grains, similar in grain size and distribution to foundry sands, and is available in large quantities. To remove all volatile hydrocarbons, raw fluid coke was calcined at approximately 2800 F (1538 C). It was noted that the calcining treatment increased the real density of the grains, and it was later determined that calcining improved the thermal stability of the fluid coke. Table 1 shows a typical inspection of the calcined fluid coke evaluated in this report, and a typical screen analysis is given in Table 2. These characteristics of calcined fluid coke indicated it

*U. S. Patents No. 2,830,342 and 2,830,913.

**Described by Harry W. Dietert in "Processing Molding Sands," Charles Edgar Hoyt Annual Lecture, AFS TRANSACTIONS, *vol. 62*, p. 1 (1954).

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should be tried as a molding medium for the casting of metals.

A considerable amount of experimental work was carried out to evaluate calcined fluid coke as the principal molding material, replacing sand in green molds, baked molds, baked cores and shell molds for casting of both ferrous and nonferrous metals. Pulverized calcined fluid coke was also tested in mold and core washes. Due to the pioneering nature of this work, most of these experiments were conducted in actual foundry production and evaluated by observation. Therefore, the examples reported in this paper were chosen for completeness of data, photographs and laboratory support, rather than as examples of preferred technique in the use of calcined fluid coke. All work reported in this paper was conducted with calcined fluid coke, but, to permit a less cumbersome terminology, the material will be referred to hereafter in this report simply as fluid coke.

TABLE 1 — TYPICAL INSPECTION OF FLUID COKE TESTED

| | |
|---|---|
| Carbon, Wt. % | 97.7 |
| Sulfur, Wt. % | 2.0 |
| Volatile, Wt. % | nil |
| Apparent density (gm./cc) | 1.11 |
| Lb/gallon | 9.26 |

NOTE: Carbon, sulfur and volatile contents of fluid coke vary with origin of petroleum stock and processing technique.

TABLE 2 — TYPICAL SCREEN ANALYSIS OF FLUID COKE AND TEST SAND

| | Per Cent Retained on | |
|---|---|---|
| U.S. Sieve No. | Fluid Coke | New Jersey 60 Silica Sand |
| 6 | — | — |
| 12 | — | — |
| 20 | 0.2 | 0.2 |
| 30 | 0.4 | 0.4 |
| 40 | 2.4 | 3.9 |
| 50 | 12.6 | 17.2 |
| 70 | 27.7 | 34.9 |
| 100 | 37.9 | 30.6 |
| 140 | 14.2 | 8.9 |
| 200 | 3.8 | 2.5 |
| 270 | 0.6 | 0.8 |
| Pan | 0.2 | 0.6 |
| Clay | None | 0.5 |
| A.F.S. Fineness No. | 67.5 | 63 |

## GENERAL TEST PROCEDURES

The tests were set up to provide an early appraisal of fluid coke in practical foundry applications. Castings were produced with fluid coke molds and cores, and were compared with similar castings produced by conventional sand practice. To the extent possible with production equipment, these tests were conducted in a manner allowing the mold and/or core composition to be the only variable. The castings intended for comparison were poured side by side, generally from the same ladle, and were then examined for any differences in surface finish and soundness.

### *Tests of Fluid Coke in Green Molds*

To evaluate fluid coke as a green mold material, fluid coke mold mixtures were prepared in a laboratory-size mixer, and thereafter, handled with regular foundry production equipment. Jolt-squeeze molding machines were used to make all molds. Patterns for castings of gray iron, bronze and aluminum were selected from current production work. In each case, molds of the sand mixture regularly used by the foundry were poured alongside the test mold as a control.

Table 3 lists fluid coke molding mixtures and corresponding control sand mixtures, showing their physical properties, and identifying the type of metal cast in each; the metals cast being gray iron, bronze and aluminum. Each pair of molds was rammed, poured and shaken out in a comparable manner. The castings so produced were inspected for surface smooth-

TABLE 3 — GREEN MOLD MIXTURES FOR VARIOUS METALS

| | Gray Iron | | Bronze | | Aluminum | |
|---|---|---|---|---|---|---|
| Mix No. | 1 | 2 | 3 | 4 | 5 | 6 |
| New Jersey 60 Silica sand, Lb | 1440 | — | — | — | — | — |
| New Jersey 120-2 Silica sand, Lb | — | — | 105 | — | — | — |
| New Jersey 160 Silica sand, Lb | — | — | — | — | 1000 | — |
| Providence River sand, Lb | — | — | 105 | — | — | — |
| Fluid coke, Lb | — | 30 | — | 40 | — | 40 |
| Cereal, Wt. % | 1.2 | 1.2 | — | — | — | — |
| Wheat flour, Wt. % | 1.0 | — | — | — | — | — |
| Western bentonite, Wt. % | 2.0 | 4.0 | — | — | — | — |
| Southern bentonite, Wt. % | 5.8 | — | 4.8 | 5.1 | 4.0 | 4.1 |
| Wood flour, Wt. % | — | — | — | — | 0.4 | 0.4 |
| Sea coal, Wt. % | 4.4 | — | — | — | — | — |
| Pitch, Wt. % | — | — | 1.0 | 1.0 | — | — |
| Moisture (determined) Wt. % | 3.6 | 4.8 | 3.3 | 3.8 | 6.0 | 4.9 |
| Lytron, Wt. % | — | — | — | — | 0.025 | 0.025 |
| **Physical Properties** | | | | | | |
| Green permeability | 81 | 49 | 52 | 37 | 27 | 34 |
| Green compression, p.s.i. | 6.3 | 5.4 | 8.2 | 4.7 | 4.4 | 3.4 |
| Green deformation, in./in. | 0.012 | 0.028 | 0.012 | 0.012 | 0.019 | 0.020 |
| Hot strength, 1500 F, p.s.i. | 95 | 165 | 180 | 180 | 442 | 101 |
| Hot strength, 2000 F, p.s.i. | 16 | 4 | 102 | 6 | 55 | 3 |
| Thermal expansion, 1500 F, in./in. | 0.028 | 0.004 | Not Run | Not Run | 0.015 | 0.006 |
| Metal specifications | A.S.T.M. A48 | Class 45 | Navy M Bronze | | Heat treatable Si-Al alloy | |
| Metal pouring temperature, F | 2620 | 2620 | 2250 | 2250 | 1290 | 1290 |

NOTE: Mixes 1, 3 and 5 were regular production batches.

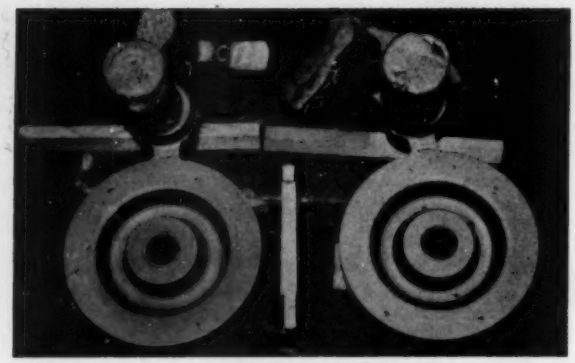


Fig. 1 — Gray iron castings made in fluid coke green mold (left) and green sand mold (right).

ness, chill tendency, sulfur and carbon pickup, shrinkage tendency and grain structure.

It was established by observation that better peel was obtained with fluid coke than with regular molding sand, the castings taken from the fluid coke molds being clean and practically free of any adhering mold material. All the castings, gray iron, (Fig. 1) bronze and aluminum, cast in fluid coke molds showed somewhat smoother surface finish than respective castings produced in the regular sand molds. Brinell hardness readings (not recorded) established there was no appreciable difference in hardness between castings made in fluid coke and those made in sand. Radiographs taken of the gray iron castings showed some improvement in internal soundness of the castings produced in fluid coke molds.

Of particular significance, photomicrographs of etched cross-sections of the bronze and aluminum castings revealed a finer grain structure was obtained with both bronze and aluminum cast in fluid coke molds than was obtained in the sand-cast castings (Fig. 12).

### *Tests of Fluid Coke in Baked Cores*

To determine the characteristics of fluid coke as a replacement for zircon and silica sands in baked cores, a mold and core assembly was utilized in a fashion to accommodate six 2 in. diameter standard permeability based core specimens prepared from the core mixtures given in Table 4. Thus, fluid coke was compared to zircon sand, silica sand and their mixture, as indicated in Table 4 (Figure 2 is a drawing of the test mold, similar designs of which are familiar to most foundry researchers. Figure 3 illustrates the mold and test cores prior to closing). Navy *M* bronze was poured at 2075 F (1079 C), and the resultant casting (Fig. 4) was inspected for any differences in the cored area.

For the purpose of discussing the results observed, the core location numbers also identify the composition of the cores as given in Table 4. The cored areas of the test casting (Fig. 4) at locations 1, 2, 3 and 5 showed no penetration. Considerable veining was present at location 6, and location 4 was completely penetrated. The conclusion drawn was that fluid coke was equal to zircon sand and much supe-

TABLE 4 — CORE MIXTURES FOR PENETRATION TEST WITH NAVY M BRONZE

| Core Location in Test Casting | 1 | 2 | 3 | 4 | 5 | 6 |
|---|---|---|---|---|---|---|
| Fluid coke, Lb | 25 | 25 | 12.5 | — | — | — |
| Zircon sand, Lb | — | — | 12.5 | 1020 | 50 | — |
| New Jersey 60 Silica sand, Lb | — | — | — | 930 | — | 1440 |
| Cereal, Wt. % | 2.0 | 2.0 | 2.0 | 0.4 | — | 0.5 |
| Highly dextrinized cereal, Wt. % | — | — | — | 0.1 | 0.1 | 0.2 |
| Southern bentonite, Wt. % | — | — | — | — | 0.5 | — |
| Core oil A, Wt. % | 2.0 | — | 2.0 | — | — | — |
| Core oil B, Wt. % | — | 2.5 | — | 0.7 | 1.5 | 0.8 |
| Moisture (determined) Wt. % | 6.6 | 6.0 | 6.0 | 5.5 | 1.4 | 5.8 |
| **Physical Properties** | | | | | | |
| Dry permeability No. | 27 | 27 | 21 | 65 | 37 | 140 |
| Scratch hardness | 91 | 87 | 100 | 97 | 88 | 88 |

Test cores.....2 in. diameter standard permeability specimen.
Baking equipment....Air-circulating laboratory oven.
Baking temperature...425 F.
Baking time.....2 hr
Wash treatment.....All baked specimens coated with 40 Baume graphite wash.
Weight of test casting..178 lb (including riser)

NOTE: Mixtures 4 and 6 were regular production batches. All other mixtures were mixed in a laboratory mixer — 3 min dry, 3 min water, 5 min oil.

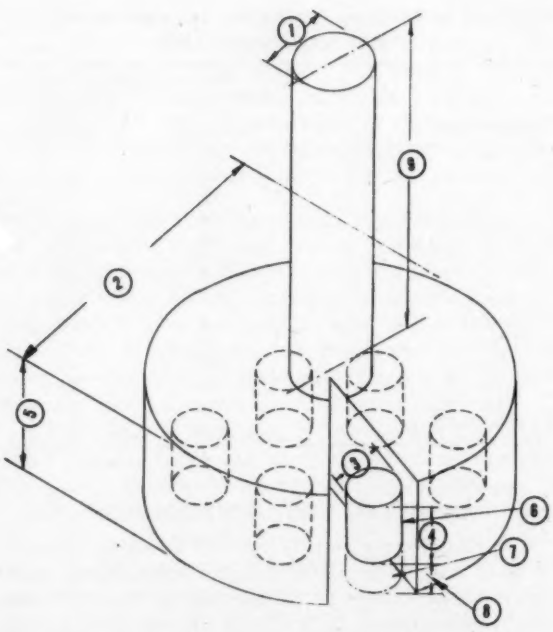



Fig. 2 — Drawing of mold and core assembly used to show comparative penetration of bronze into baked test cores. Weight of casting and riser — 178 lb. *1* — 3 dia; *2* — 13 dia; *3* — 2 dia; *4* — 2 in.; *5* — 4 in.; *6* — core 2 in. dia, 3 in. long; *7* — core print one in. deep; *8* — one in.; *9* — 11 in.

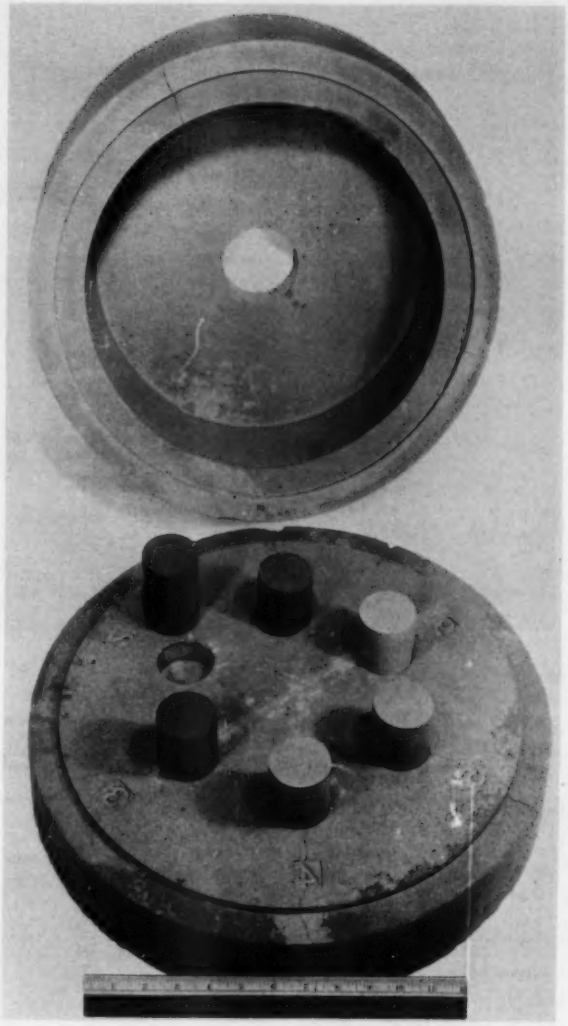


Fig. 3 — Penetration test mold showing location of test cores.

rior to silica sand in resistance to penetration of Navy *M* bronze under the conditions of this test.

To further compare fluid coke to zircon sand as a coremaking material, centrifucal pump impeller cores were prepared from each according to the core mixtures and data given in Table 5. The test cores were baked, coated with 40 degrees Baume commercial graphite base core wash, set in the molds and poured with Navy *M* bronze. The resultant castings (Fig. 5) were examined for condition of the cored areas. Both castings were smooth and free of veining in the cored areas. It can be said that fluid coke proved equal to zircon sand under the conditions of this test.

To evaluate fluid coke as a coremaking material for use with gray iron, cores were made for commercial gray iron bearing housings, according to the core mixtures and preparation data shown in Table 6, thus comparing a fluid coke core with the silica sand core regularly used for these castings. The molds for these castings were baked molds of the same composition as the respective cores. These molds were poured consecutively with gray iron at 2480 F (1360 C). Castings were radiographed, then cross-sectioned for examination (the cored areas are shown in Figs. 6 and 7).

The radiographs revealed two small areas of shrinkage porosity in the casting produced with the sand mold and cores, whereas, only one small area of shrinkage porosity was evident in the casting made with the fluid coke mold and cores. The surface finish on both mold and core areas of the casting made with fluid coke was superior to that obtained with the sand mold and cores.

### *Test of Fluid Coke for Shell Molds*

To investigate the use of fluid coke as a replacement for silica sand in shell molds, a mixture of fluid coke and 6 Wt. per cent powdered shell mold resin was prepared in the same fashion as a mixture of No. 155 Superior Dividing Creek sand and 6 Wt. per cent of the same resin. Shell molds for aluminum air hammer housings were made with each mixture. The shells were invested on heated patterns by the dump box method and cured under electrical resistance heaters. Regular oil sand cores were set in

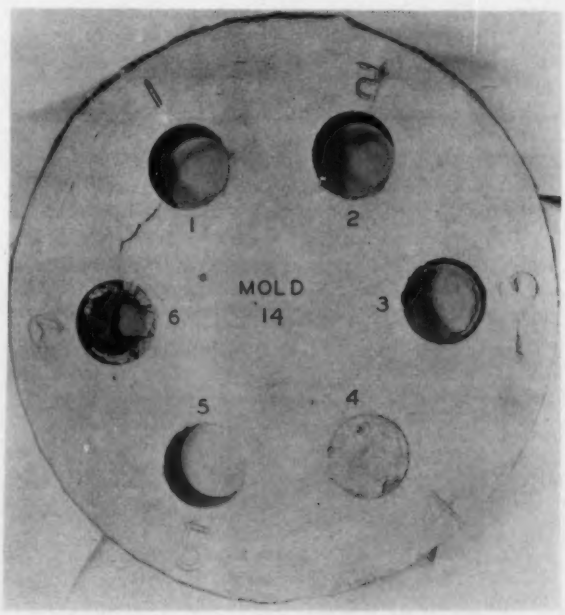



Fig. 4 — Penetration test bronze casting showing degree of penetration into each test core (location numbers identify core composition as given in Table 4).

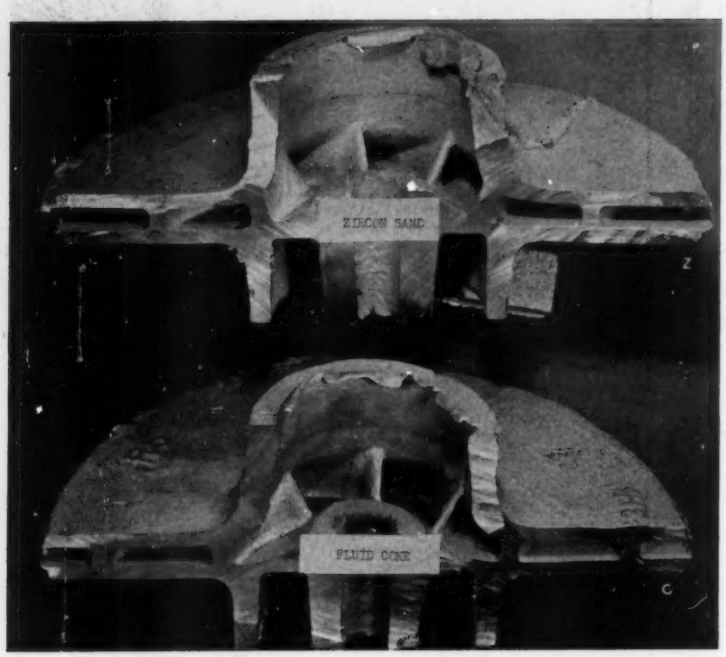



Fig. 5 — Cross-sections of bronze pump impellers showing interiors cored with zircon sand core (top) and with fluid coke core (bottom).

TABLE 5 — CORE MIXTURES — FLUID COKE VS. ZIRCON SAND FOR NAVY M BRONZE PUMP IMPELLERS

| Mix no. | 1 | 2 |
|---|---|---|
| Zircon sand, (AFS no. 95) Lb | 500 | — |
| Fluid coke, (AFS no. 95) Lb | — | 25 |
| Core oil, Wt. % | 0.5 | 2.0 |
| Cereal, Wt. % | — | 2.0 |
| Moisture, Wt. % | 0.8 | 7.0 |
| Physical Properties | | |
| Dry permeability | 39 | 39 |
| Scratch hardness | 98 | 100 |

Baking equipment ..... Production Batch-type Oven.
Baking temperature .... 400 F.
Baking time ..... 3½-hr.

NOTES: 1. The fluid coke was sieved and blended to match the AFS Fineness number of the zircon sand.

2. Mix 1 was regular batch from foundry core room. Mix 2 was mixed in a laboratory mixer — 3 min dry, 3 min water, 5 min oil.

these molds. The molds were clamped and poured horizontally with heat treatable silicon-aluminum alloy at a pouring temperature of approximately 1290 F (699 C). The aluminum castings so produced were compared for surface finish and soundness (castings and molds are shown in Fig. 8).

The casting produced in the fluid coke shell mold was of somewhat smoother surface finish then the casting made in the silica sand shell mold. Photomicrographs (Fig. 13) taken on the cross-sections of both aluminum castings showed a finer grain structure was obtained with the fluid coke shell mold.

### *Test of Pulverized Fluid Coke in Core Wash*

Fluid coke was pulverized to a particle size of 96 per cent through U.S. 200 mesh for use in a test to determine its characteristics as the principal ingredient of core wash. A fluid coke core wash having a 39 degrees Baume gravity was prepared according to the formula:

| Material | Wt. % |
|---|---|
| Fluid coke (200 mesh) | 42 |
| Western bentonite | 3 |
| Dextrin | 2 |
| Water | 53 |

A test was set up to compare this wash with a commercial graphite base core wash having a gravity of 40 degree Baume. Two identically baked oil-sand cores were dipped, one core being coated with fluid coke and the other with commercial graphite wash. Both cores were dried in a production drying oven and then placed in a double casting mold to produce, in a single mold, a gray iron casting with each core (Fig. 9). Gray iron was poured at 2600 F (1427 C). The castings were cross-sectioned and examined for surface finish on the cored areas (Fig. 10). The casting produced with the core coated with the fluid coke core wash showed somewhat smoother surface finish than was obtained with the commercial graphite base core wash.

## FURTHER OBSERVATIONS AND DISCUSSION

The first trail of fluid coke as a molding medium established one basic characteristic subject to earlier apprehension. It was immediately determined that fluid coke would not ignite and be consumed by the heat of the molten metal. Fluid coke grains in direct contact with molten iron may burn to a white ash, however, the grains immediately adjacent to this ash

appear to remain unaffected. The amount of ash formed is not sufficient to prevent continuous recycling of shakeout fluid coke.

Fluid coke, used either as mold and core material or as a wash is considered beneficial for peel, since castings shake out clean and practically free of any adhering mold and core material. Furthermore, fluid coke assures a reducing or inert blanket of gas at the mold-metal interface during pouring, without addition of sea coal or other organic materials ordinarily used for that purpose. In no case, during the many tests, did any casting show a blow or other defect due to gas evolution from fluid coke.

**Bulk Density.** The low bulk density of fluid coke as compared to sand requires special consideration. As shown in Table 1, the fluid coke used in these tests weighed 9.26 lb/gallon, whereas most foundry sands weigh near to 12 lb/gallon. Since practically all additives and binders function on a volume relationship to the base material, it follows that a larger weight percentage of water and binders is generally needed to produce fluid coke mixtures having green and baked strengths equivalent to sand mixtures. While this bulk density ratio of fluid coke to sand was not generally followed in setting up the fluid coke mixtures described in this paper, it should be kept in mind in interpreting the physical properties of the mixtures used.

From a material handling standpoint, the fluid coke used in these tests weighed approximately three-fourths the weight of an equal volume of silica sand, and less than 40 per cent the weight of an equal volume of zircon sand.

**Thermal Expansion.** The low thermal expansion of fluid coke as compared to silica sand is, perhaps, its most unique characteristic. In all thermal expansion tests, which have been run on fluid coke mixtures, using the $1\frac{1}{8} \times 2$ in. specimen in a dilatometer at 1500 F (816 C), the highest expansion value was 0.006 in./in. and the lowest was 0.002 in./in. In most cases, thermal expansion of fluid coke mixtures was about one-seventh that of a corresponding silica sand mixture.

TABLE 6 — CORE MIXTURES FOR GRAY IRON TEST CASTING

| Mix no. | 1 | 2 |
|---|---|---|
| New Jersey 60 silica sand, Lb | 1440 | — |
| Fluid coke, Lb | — | 40 |
| Corn flour 1, Wt. % | 0.5 | 2.0 |
| Corn flour 2, Wt. % | 0.2 | — |
| Core oil, Wt. % | 0.83 | 2.0 |
| Moisture (determined), Wt. % | 6.8 | 7.0 |

Baking equipment ..... Production Vertical Oven.
Baking temperature ..... 550 F.
Baking time ..... 1¾-hr, load to unload.
Wash treatment ..... Both baked cores coated with 40 Baume graphite wash.

NOTE: Mix 1 was regular production batch from foundry core room.
Mix 2 was mixed in a laboratory mixer — 3 min dry, 3 min water, 3 min oil.

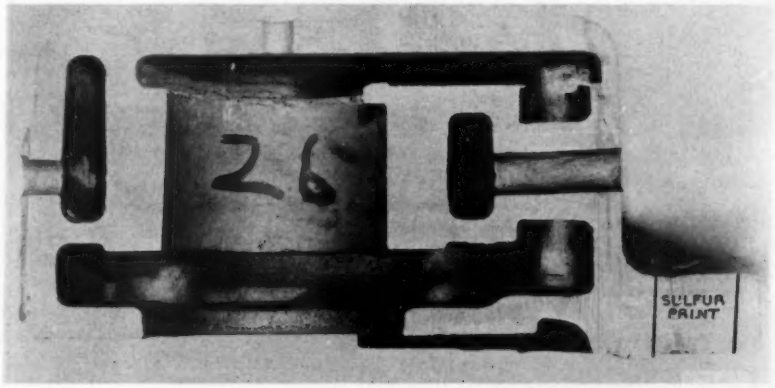



Fig. 6 — Cross-section showing cored area of gray iron casting produced with baked oil-bonded fluid coke mold and cores.

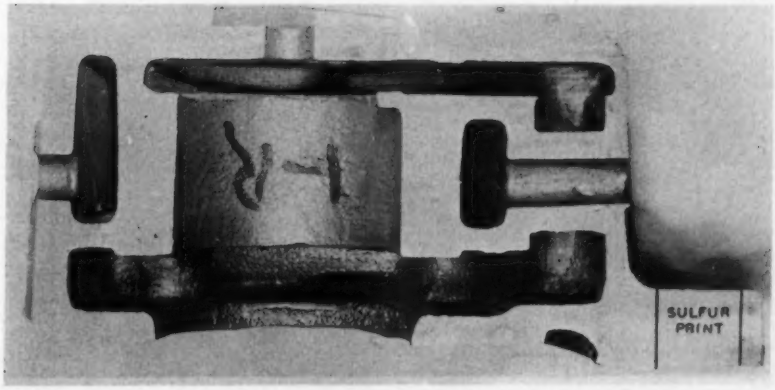



Fig. 7 — Cross-section showing cored area of gray iron casting produced with baked oil-bonded silica sand mold and cores.

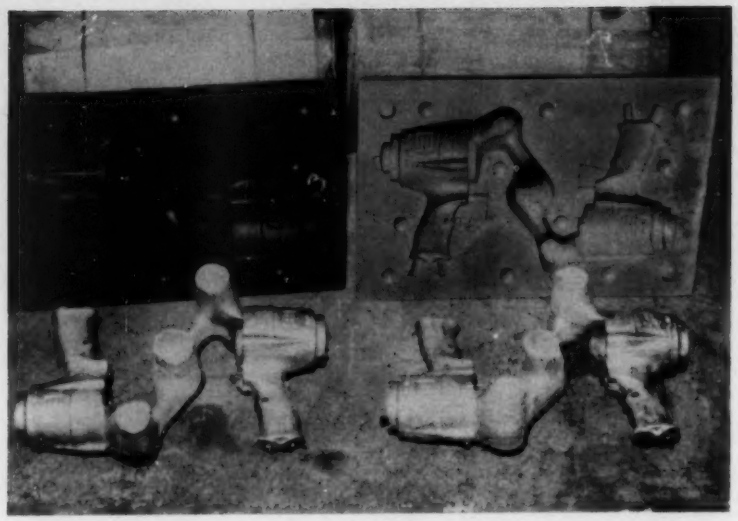


Fig. 8 — Aluminum air hammer housings cast respectively in fluid coke shell mold (left) and silica sand shell mold (right).



Fig. 9 — Gray iron castings cored with oil-sand cores coated with fluid coke core wash (right) and commercial graphite-base core wash (left).

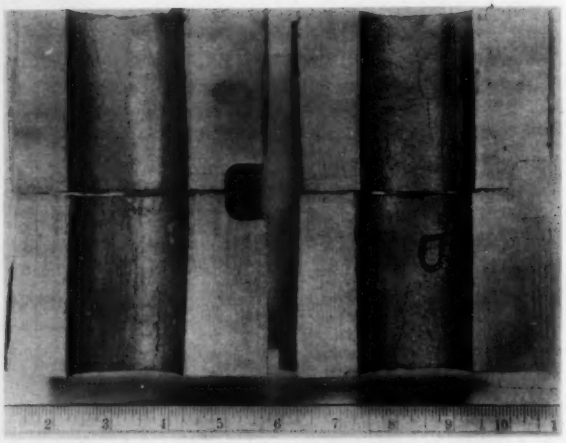


Fig. 10 — Cross-sections of gray iron castings showing areas cored with fluid coke core wash (left) and commercial graphite-base core wash (right).



Fig. 11 — Gray iron castings shaken out 30 min after pouring showing faster cooling with fluid coke green mold (left) compared to green sand mold (right).

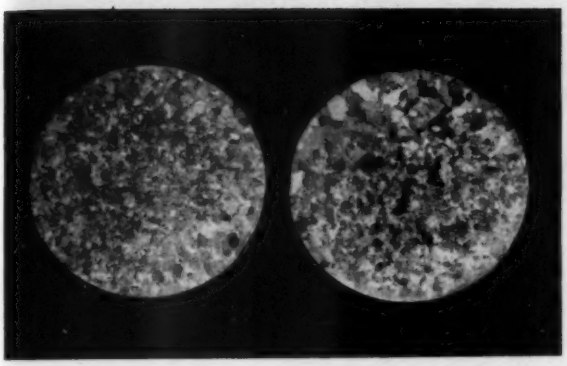


Fig. 12 — Photomicrographs showing grain structure of bronze cast in green mold of fluid coke (left), and sand mold (right).

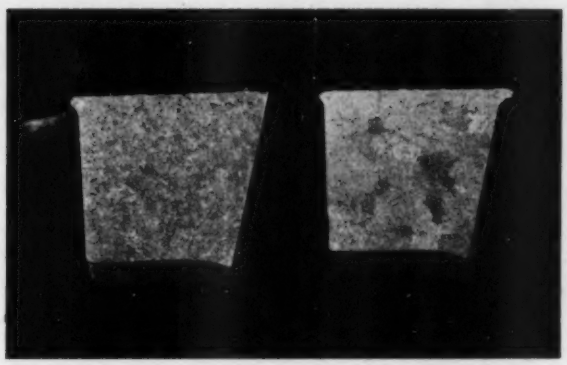


Fig. 13 — Photomicrographs showing grain structure of aluminum cast in shell molds of fluid coke (left) and silica sand (right).

**Thermal Conductivity.** As could be expected, fluid coke conducts heat rather rapidly. Core baking time and investment and cure times for shell molds can be shortened somewhat.

The effect of heat transfer on the progressive solidification and cooling rate of castings poured in fluid coke is a more important consideration. Identical gray iron castings shaken out 30 min after pouring showed the casting poured in a sand mold to be a cherry red heat, while the casting made in a fluid coke mold was at black heat (Fig. 11). Contrary to expectations, no chilling effect was evident on the casting surface. Due to the faster cooling of the castings made in fluid coke, there was some evidence of less shrinkage within the castings, which would indicate that fewer or smaller risers would suffice to feed the casting, and in some cases, risers might be eliminated. Less shrinkage in the casting is believed attributable to both the faster cooling of the castings and to the thermal stability of the fluid coke molds.

Photomicrographs revealed that a finer and tighter grain structure was obtained in bronze and aluminum castings made in fluid coke. The advantages realized in strength and pressure tightness are obvious. This improved grain structure is considered due to the faster cooling rate provided by fluid coke (Figs. 12 and 13).

**Surface Finish.** The most obvious advantage obtained with fluid coke was the superior surface finish of gray iron, bronze and aluminum castings produced with fluid coke green molds, baked cores, shell molds and core wash. Since molten iron does not wet carbon, and considering the low thermal expansion of fluid coke, a decrease should be expected in casting defects such as penetration, burn-on, veining, scabbing and rattailing through the use of fluid coke.

## CONCLUSION

The advantages of fluid coke appear to be these:

1. The grain structure, grain size and grain distribution of fluid coke is similar to that of foundry sands, and therefore, it lends itself to handling with ordinary foundry equipment.
2. The high carbon content of fluid coke provides low thermal expansion and resistance to penetration by molten metal.
3. The high rate of heat conductivity with fluid coke provides faster cooling of castings before shakeout, with consequent improvement in soundness and grain structure of the castings, particularly bronze and aluminum.
4. Fluid coke imparts superior surface finish to castings by any casting method in which it has been so far tried.
5. The low weight per unit volume of fluid coke reduces the effort and cost of handling.

This preliminary evaluation of fluid coke as a molding medium covers some of the initial work done to determine its relative usefulness, as compared to more conventional materials. The tests indicated that, while fluid coke does not cure all foundry molding problems, it does merit consideration as an aid toward meeting the ever increasing demand for higher quality castings. Fluid coke is unique in that its most interesting characteristics serve to improve the resultant castings. Work has continued to determine the compatibility of fluid with the newer casting processes such as $CO_2$ process, cold-setting binder process, resin-coated shell mold practice and hot-box process. Suffice to say, the later work has further confirmed the reliability of the observations reported in this paper.

## ACKNOWLEDGMENT

The author wishes to acknowledge and sincerely thank Ingersoll-Rand Co., Phillipsburg, N.J., for their cooperation and assistance in conducting the work described in this paper.

# MAXIMIZING SOUNDNESS IN ALUMINUM DIE CASTINGS

by L. M. Elijah


## ABSTRACT

**The increasing usage of aluminum die castings, the insistent demands for higher mechanical requirements and the quickening tempo of production toward automation has made highest quality requirements mandatory. With this in view, a complete analysis is made of the known shortcomings, indicating causes and suggesting actual remedies which stress often overlooked and little known metallurgical features.**


## INTRODUCTION

With the rapid growth of aluminum die castings into a major industry, the headaches and pains which are part of the growth process have correspondingly increased. The fact that quality cannot be inspected into a part but must be cast into it, makes it vital that all precautions be taken before the molten metal is frozen in the die cavity. Procedures must be rigorously controlled and checked for discrepancies. Correct casting and die cavity design have been clearly indicated,[1] hence for the purpose of this paper, it will be generally assumed that they are followed. Other investigators have elucidated the theoretical and practical aspects of die casting thermo- and hydro-dynamics.[2,3,4] However, little has been mentioned as to the metallurgical factors involved.

## MODE OF ANALYSIS

It should be recognized that the sources of defects are:

a. The die.
b. The machine.
c. The metal.

This can be further subdivided:

a. 1. Die cavity design.
   2. Operating conditions of the die.
b. Operational techniques of the casting equipment.
c. 1. The intrinsic nature of the alloy itself.
   2. Exact conditions for melting and pouring the alloy.

However, since most defects originate from two or more of the above factors, they will be generally classified as:

A. Observed prior to injection.
B. Caused during injection.
C. Observed subsequent to injection.

## TROUBLE SHOOTING THE PROBLEMS

### DEFECTS OBSERVED PRIOR TO INJECTION

*Table 1 — Corrosion Resistance (Poor)*

*Classification:*

Rapid surface deterioration (caused by oxidation, etc.) which mars the bright, shiny appearance.

*Causes:*

1. Improper surrounding atmospheres.
2. Exposure to the elements (storage outdoors).
3. Absence of grain refiners.
4. Alloy not suitable for particular application.

*Corrections:*

1. Presence of chlorine and hot furnace gases in atmosphere, hence store away from such fumes.
2. Store in dry, enclosed area.
3. Grain refiners like Ti help improve surface appearance.
4. Effects of elements are:

   Increasing amounts of Mg do not decrease the excellent corrosion resistance of Al. Si additions up to 13 per cent result in a slight deterioration (negligible for practical purposes). Up to 2.5 per cent Zn has no harmful effect. Increasing Cu content lowers corrosion resistance specially up to one per cent. Over this amount the alloy improves (hence Cu should be maintained below 0.3 per cent. Increasing Fe (over 0.5 per cent) generally lowers this property.

*Table 2 — Drossing (High)*

*Classification:*

Formation of excessive surface layer of impurities on molten metal.

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*(Table 2 — continued)*

*Cause:*

Oxidized metal ($Al_2O_3$) combined with spent flux, together with other impurities (usually nonmetallic). $Al_2O_3$ occurs eventually in castings either as smaller, well dispersed particles (seen unetched, under a low power microscope as tiny, dark specks on a polished surface) or as large, conglomerated masses (appearing as black spots under a low power microscope). The former is caused by insufficient cleaning of returned scrap or turbulent gating design. The latter is caused by careless ladling or neglectful metal cleaning and housekeeping operations.

*Corrections:*

1. Prevent undue turbulence under solid fluxing. For this reason, $N_2$ or $Cl_2$ gas bubbled quietly is more efficient ($Cl_2$ is better, but in alloys requiring Mg, loss of Mg must be compensated by having a slight initial excess).
2. In the cold chamber process, keep the surface clean. Ensure that the operator does not agitate the surface in order to obtain clean metal.
3. When transferring molten metal from one furnace to another or from a furnace to the injection chamber, pour in a thick, steady stream with a minimum of metal drop.
4. Oxidized, finely divided or coated scrap causes high dross. Make sure that the returns as well as fresh charge are bulky, clean and free from oxides, oil, dirt, paint, etc. Do not store metal outdoors or in contact with chlorine or furnace fumes.
5. Prevent overheating of the molten metal.
6. High magnesium content in an alloy causes excessive drossing. This can be reduced considerably by adding 0.01 per cent of each Be and Li. If possible, use an alloy low in Mg.
7. If a high magnesium alloy is required, make controlled use of Be.

### *Table 3 — Fluidity (Poor)*

*Classification:*

Inadequate flowability of the molten alloy to fill intricate designs and sharp corners.

*Causes:*

1. Presence of sludge.
2. Presence of $Al_2O_3$ in metal.
3. Incorrect alloy composition.
4. Low temperatures decrease fluidity.
5. Alloy used has high shrinkage.
6. Inadequate feeding.

*Corrections:*

1. Eliminate conditions causing sludging.
2. Chlorinate metal to reduce $Al_2O_3$ inclusions and hence increase flowability.
3. Higher silicon content (up to 12 per cent) increases fluidity. High Mg causes drossing which reduces fluidity. Mn over 0.30 per cent reduces castability.

*(Table 3 — continued)*

4. Higher temperatures increase fluidity, enabling thin intricate sections to be poured.
5. If after using chlorine, poor fluidity still persists, add approximately 0.15 per cent Ti which tends to reduce shrinkage. Higher Fe content, up to one per cent, also reduces the shrinkage characteristics.
6. Increase ingate and feeding system.
7. The fluidity of alloy increases with an increase of Cu up to 12 per cent.

### *Table 4 — Hot Shortness*

*Classification:*

At higher temperatures, alloys possessing this defect tend to embrittle.

*Cause:*

Basic characteristic of some alloys.

*Corrections:*

1. Presence of Sn, Pb increases hot shortness. Higher Mg alloys also have this tendency. Both Si and Ni tend to decrease this characteristic slightly. The tendency increases up to 4.5 per cent Cu and then decreases, this is reduced in the presence of one per cent Fe.
2. For design defects, refer to "Shrinkage Cracks."

### *Table 5 — Metal Loss (High)*

*Classification:*

Low metal recovery on melting.

*Cause:*

Oxidation, sludging and carry out of metal during melting and pouring.

*Corrections:*

1. As any of the conditions that tend to cause excessive dross also causes high metal loss, observe the factors mentioned therein.
2. Similarly, all of the factors minimizing sludging (refer) must be followed.
3. In magnesium bearing alloys, maintain chlorination at a minimum as, along with the removal of magnesium, eventually aluminum is also removed.
4. The use of solid fluxes causes a carry out of molten metal when residues of the former are removed. Hence, during melting, casting and pouring, reduce the use of solid fluxes and dross removal to a minimum.
5. Employ indirect heating (e.g., electric induction and crucible type furnaces) in preference to direct-fired reverberatory furnaces.

### Table 6 — Porosity — Gas Absorption

*Classification:*

Smooth round highly reflective holes, mostly internal and evenly distributed, caused by gas dissolved in the molten metal due to exposure to humid atmospheres, furnace combustion gases, wet or dirty charge, etc., and the eventual release of these gases on cooling and solidification. They decrease mechanical properties and if they appear at or near the surface, cause galling and seizing of tight contacting surfaces in motion.



**Fig. 1 — Porosity which could be caused by absorbed gas, turbulence or excessive die coating.**

*Cause:*

Improper treatment of the molten alloy.

*Corrections:*

1. Use clean, dry scrap.
2. Melt rapidly. Do not overheat.
3. Degas thoroughly immediately prior to pouring ($Cl_2$ gas is most effective, however it also reduces Mg. Should be used in alloys where Mg is an impurity. In alloys containing Mg use $N_2$ gas).
4. Make sure that solid fluxes employed are thoroughly dry. If exposed to humid atmospheres, they tend to absorb moisture.
5. Ascertain that only dry gases are employed for degassing, e.g., dry, oil pumped $N_2$, etc.
6. Furnace design should be such that spent gases do not come in contact with the molten metal surface.
7. Make sure that excessive Ca is not present in the metal, as this tends to cause porosity. Ca can be minimized by fluxing with Cl.
8. Avoid violent stirring when adding solid flux to molten metal.
9. Possibly the die coating is excessive or defective. Check if oils, greases are the cause.
10. Vacuum melted castings will improve quality only if molten metal is initially degassed.
11. Use may be made of high frequency vibrations before solidification to minimize porosity.

### Table 7 — Sludging (High)

*Classification:*

Segregation of various constituents of the alloy.

*Cause:*

Precipitation of higher melting point intermetallic compounds generally caused by a lowering of temperature, presumed to be nucleated by formation of $MnAl_3$ at cold spots. The presence of $Al_2O_3$ particles (finely divided or chunk) in the melt also act as nuclei to accelerate precipitation.

*Corrections:*

1. Preheat metal charge being added (to approximately 600 F (316 C) away from open flame). Add gradually to prevent localized or general chilling of the melt.
2. Raise the temperature of the melt to approximately 350 F (177 C) above the liquidus, and mix gently to bring all of the more refractory constituents into solution at regular intervals. Use dry $N_2$ for degassing and mild mixing. $Cl_2$ gas is preferable as it also cleanses the melt of $Al_2O_3$, however its effect on Mg must be considered. Maintain an operating temperature of at least 70 F above liquidus in the electric and 100 F in the crucible type furnaces.
3. Use an electric induction furnace in preference to gas or oil fired furnaces, owing to the constant automatic metal circulation.
4. If sludge formation is excessive, empty the furnace completely before introducing a new heat, because after most of a heat is used and it is then corrected by a high temperature reheat, as in step 2, the composition of the contained liquid would go out of tolerance as also the metal used.
5. Keep Mn as much below 0.30 per cent as possible. Also use alloys in which Cr, Ti and Fe are as low as possible. At normal pouring temperatures, just above the liquidus, if the total of ($\%$ Fe + 2 × $\%$ Ti + 2 × $\%$ Mn + 3 × $\%$ Cr) exceeds 1.9, sludge will form. 356 (7.5 per cent Si and 0.85 per cent Fe) has less tendency to dross, oxidize or sludge than 360 or 380. 43 forms a crust and clogs channels of induction furnace at 1200 F (649 C). This may be corrected by maintaining at 1225 F (657 C) but it reduces production by 25 per cent due to higher freezing time. By changing to an alloy with 9.0 per cent Si, crust disappears at 1200 F (649 C).
6. a) If an iron crucible is used, coat it internally at frequent, regular intervals.
   b) The burners should not heat the contents unevenly, causing hot and cold spots.
7. It is preferable to have separate melting and holding furnaces, so that there is no chilling effect of a continuously added cold charge and the correct composition of metal is maintained.
8. If separate melting and holding furnaces are not available, after addition of each batch of fresh ingots or a cold charge, the whole contents must be reheated as in step 2.
9. Do not add more metal than that which the furnace burners can handle effectively.

*(Table 7 — continued)*

10. Sometimes excessive amounts of an element like Ni increases general liquidification temperatures, causing a premature freezing at a given operating range.

### *Table 8 — Venting of Dies (Poor)*

*Classification:*

This defect causes misruns, laps, seams, etc.

*Cause:*

Back pressure of air in the die-mold which prevents complete filling of the cavity.

*Corrections:*

1. For the efficient working of a hot dry die, vents should be at least half the gate area.
2. With increased lubrication and colder dies, the venting area should be increased.
3. It is advisable to vent from the overflow cavities.
4. Ensure that the metal does not plug the venting orifices prematurely.



**Fig. 2 — Folds observed in the metal as a result of cold die or metal, excessive die coating or inadequate venting.**

5. Use die release spray that dissipates below operating die temperatures to give least back pressure. If dies operate at 350 F, the die release spray should evaporate below this temperature.

## CASTING DEFECTS

### *Table 9 — Cavities — Reciprocating Plunger Action*

*Classification:*

Ragged cavities occurring both internally and externally at various locations. They sometimes appear as approximately parallel surface cracks.

*Cause:*

When a plunger is at the bottom of its stroke, should the injected metal not be completely solidified and the plunger not allowed to dwell long enough in this position, by its quick withdrawal, it will suck back on the mushy, semi-liquid metal in the mold, causing this defect.

*Corrections:*

1. Hold the plunger at the bottom of its stroke a fraction longer.
2. Decrease the gate section so that the metal chills off before the sucking action of the plunger can deform it.
3. Water cool ¾ - 1¼-in. below the die cavity to expedite solidification.
4. Prevent overheating of the molten metal and die.

### *Table 10 — Cavities — Shrinkage*

*Classification:*

Ragged, irregularly shaped cavities caused by internal shrinkage. They possess dull cavities as opposed to gas entrapment or porosity which are rounded and highly reflective.

*(Table 10 — continued)*

*Cause:*

Inadequate feeding of shrinkage in a casting because of the premature bottleneck freezing action of the molten metal nearer to the gate or the adjacency of a thin and thick section (shrinkage resulting from a difference in specific volumes of the molten metal at the liquidus and the solidified metal at its solidus temperature).

*Corrections:*

1. Check if die temperatures are too low.
2. Even out die temperatures, by increased feeding, using thicker gates and more overflow wells adjacent to the coolest portion.
3. Minimize drastic changes in section thickness.
4. Increase injection pressure.
5. Use a grain refiner for the molten metal (refiners are not normally necessary due to the chill of the metal die, however presence of 0.15-0.20 residual Ti in the melt sometimes helps to decrease defect. Use should be made of a 5 per cent Ti, 95 per cent Al master alloy).
6. Use an alloy with better feedability characteristics.
7. Check if lower metal temperatures would decrease this defect without causing cold shut.

### *Table 11 — Cold Shut*

*Classification:*

Marks caused by metal streams failing to merge together and completely fill the die cavity. Linear in appearance, they are noted by the absence of non-

*(Table 11 — continued)*

metallic materials (if they look like flakes of unmelted metal, it is due to metal folds containing die lubricant caused by turbulence because of too large a gate and excessive lubricant).



**Fig. 3 — Depressed edge caused by incomplete filling.**

*Causes:*

Advancing metal front divides into two streams which lose heat, develop thin oxide surface films (possibly also contain oxidized metal) and upon meeting, fail to interdiffuse partially or completely into one another, because of sluggishness, insufficient velocity or pressure to break the oxide film surface tension or due to the back pressure of air or gas, in the cavity.

*Corrections:*

1. If the die is cold, increase the temperature.
2. Check if the cooling channels are approximately 1 to 2 in. from the die cavity.
3. If poured metal is cold, increase temperature but do not heat excessively.
4. Reduce mold back pressure by adequately venting the cavity.
5. Increase injection pressure to help complete the merging of metal surface fronts.
6. If high plunger speeds tend to trap air, check if a lower speed would reduce this defect.
7. Adjust runner and ingate size (increase if too small). Correct other design deficiencies.
8. Improve fluidity of the alloy by minimizing oxide formation through adequate fluxing, prevention of overheating, eliminating turbulence during melting and pouring, controlling other impurities which would cause sluggishness.
9. Employ another alloy with better fluidity as a last resort.
10. As may be evident, most of the above factors can be minimized by vacuum die casting.

### *Table 12 — Cracks — Shrinkage*

*Classification:*

Cracks caused by shrinkage upon cooling.

*(Table 12 — continued)*

*Cause:*

Internal shrinkage stresses exceeding the mechanical properties of the alloy.

*Corrections:*

1. If the metal solidifies around a core, increase the taper or remove the casting promptly off the core, or employ an alloy that is not hot short (refer).
2. Even out drastic changes to improve casting design in section and eliminating stressed areas.
3. Check if the designated alloy has adequate mechanical properties (refer).
4. If lower mechanical properties are obtained due to low injection pressures, chemical composition of the alloy or other reasons listed in Table 26, follow the recommendations listed therein.

### *Table 13 — Depressions — Shrinkage*

*Classification:*

Shrinkage depressions on the surface of casting which are caused by inadequate feeding, slow cooling or heavy section, usually appearing as dished, depressed areas.

*Causes:*

1. Pure Al, higher Zn alloys, alloys containing over 5 per cent Ni or below 0.5 per cent Fe, etc., having a high shrinkage, accentuate this problem, sometimes resulting in cracks or cavities at the vertex of inside corners and changes of sections.
2. Inadequate feeding, e.g., low injection pressures or small injection ingates.
3. Heavy section adjacent to a thin section.
4. At areas of slow cooling, hot spots, where the casting is fed by too small an ingate, or where the metal impinges against a core.

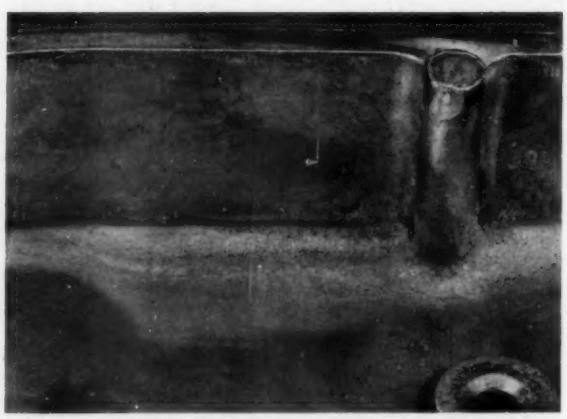


**Fig. 4 — Frosted appearance and shrinkage depressions at the corners due to hot spot caused by ingate.**

*Corrections:*

1. Use alloy with higher Fe content (0.9-1.3 per cent) and Mn up to 0.30 per cent.

*(Table 13 — continued)*

2. Feed correctly.
3. Even out section thicknesses.
4. Eliminate causes of hot spots.

### *Table 14 — Flow Lines*

*Classification:*

Marks confined to the surface having no appreciable influence on the strength of the casting. Difficult to remove for plating.

*Causes:*

1. Excess of die lubricant (entrapped die lubricant areas appear black at low magnifications and highly reflective at higher magnifications).
2. Improper gate location.
3. Back pressure in the die cavity.



**Fig. 5 — Flow lines near ingate.**

*Corrections:*

1. Excess of lubricants or dressings (especially in the hot sleeve) can cause this defect. Eliminate any excess lubricant, enlarge die vents and incorporate more overflow pads.
2. Ensure adequate venting.
3. Increase injection pressure.
4. If the die temperature is low, increase gradually and observe effect (if die is too hot, increased gases will cause worse flow marks).
5. Increase runner channels or enlarge gate entry.
6. Vacuum die casting can be useful.

### *Table 15 — Leakers (Poor Pressure Tightness)*

*Classification:*

Leakages obtained under pressures which the given wall thickness is normally expected to withstand.

*Causes:*

Shrinkage, oxide inclusions, cold shut, laps, misruns and porosity caused by improper melting procedure, can result in this defect.

*Corrections:*

1. If possible, modify the die design to minimize shrinkage. This usually occurs in the same location, being caused by hot spots, inadequate feeding, etc. Also follow the recommendations outlined previously.
2. Decrease shrinkage by employing more suitable alloy, e.g., pure Al has a high shrinkage value (6.3 per cent), on the other hand alloy 13 has low shrinkage (3.3 per cent), yet because both have a narrow freezing range, cavities will be observed.
3. Minimize the occurrence of $Al_2O_3$ inclusions by not overheating the molten metal and fluxing thoroughly. These oxide inclusions are of two types, skin or chunk, both vary in density of population and location. It is industrially impractical to allow the oxides to settle as considerable delay is involved.
4. Cold shuts, laps, misruns, etc., vary their location and are caused by low pouring temperatures and cold dies. Try to eliminate them.
5. Check porosity due to injection turbulence (refer) or gas absorption (refer), by making adequate provision for vents and overflow pads.

### *Table 16 — Porosity — Turbulence*

*Classification:*

These holes are rounded in nature, being trapped air caused by the turbulence of the injection process. They show a high degree of reflection in the cavities (as opposed to shrinkages which have ragged edges and dull cavities). They decrease mechanical properties and if they appear at or near surface, they cause galling and seizing of tight contacting surfaces in motion.

*Cause:*

Agitation during injection and pouring.

*Corrections:*

1. Pour molten metal into the cold chamber as quietly and with as low a drop as possible.
2. Decrease plunger speed (to minimize squirting and churning) without reducing pressure.
3. Use more or larger overflow wells and increase venting to help lessen the effect to a limited extent.
4. Increase injection pressure.
5. Reduce sudden changes of direction and their frequency of occurrence to a minimum.

*(Table 16 — continued)*

6. Increase depth of ingates gradually. Check results at each stage. Excessively thick ingate causes porosity at break-off point.
7. Change the ingate position to relocate porosity in a less critical area (as a final resort, if it cannot be eliminated).

### *Table 17 — Seams*

*Classification:*

Surface lines caused by the outer layers of two streams failing to interdiffuse completely.

*Causes:*

Even though the metal when poured may be hot enough to fuse, the extreme outer layers only of the two streams (chilled by the die) will not fuse when the two streams meet, causing fine seams. The depth increases at lower die temperatures. Sometimes hair cracks in the die, due to thermal attack, also impart this appearance.



**Fig. 6 — Seam clearly visible at edge of hole.**

*Corrections:*

1. Raise the die temperature (if cold).
2. Check if the die has hair cracks.
3. Check if the die cavity is adequately vented.
4. Increase injection pressure to help metal surfaces to merge.
5. Vacuum die casting can help this defect.

### *Table 18 — Soldering to Die*

*Classification:*

Casting adhering to die mold resulting in pimples, torn skin on certain areas of the casting.

*Cause:*

Molten aluminum being almost a universal solvent,

*(Table 18 — continued)*

first removes the oxide skin, then tends to alloy with the hot unoxidized iron die surface that it impinges against.



**Fig. 7 — Torn skin as a result of die soldering.**

*Corrections:*

1. Avoid high metal temperatures.
2. Avoid high die temperatures.
3. Prevent metal impingement on the critical area of die cavity by slightly altering die design by:
   a) Increasing ingate area.
   b) Altering angle of impingement.
   c) Slowing down injection speed.
4. Polish critical area of the die cavity to a high finish, after modifying the metal impingement direction and velocity.
5. Check die coating.
6. Verify if the alloy composition is correct (0.8-1.2 per cent Fe reduces die soldering).
7. If metal has been chlorinated, remove dissolved $Cl_2$ gas, $AlCl_3$ or $MgCl_2$ by cleaning with $N_2$ gas before usage.
8. Allow impurities like $MgCl_2$ to rise to top as a crust and remove (if production is not slowed down).
9. Examine die cooling system to ascertain if the critical area is being adequately cooled.
10. Check if correct die steel is employed.
11. Sludge particles in the metal promote die soldering. Ensure that causes of sludge formation are eliminated.

### *Table 19 — Stickers (To Die)*

*Classification:*

Castings that are either difficult to eject or stock to the wrong die half.

*Causes:*

Improper die design with regard to casting balance, ejection, taper, surface finish. Incorrect die coat.

*Corrections:*

1. Minimize surface roughness at the undesired "area of adhesion," e.g., normally the casting is retained on the sprue pin side, from where it is ejected. For this reason, the sprue hole should be well polished and the sprue plug relatively rough to retain the casting on the plug side from where it may be ejected.
2. Incorporate sufficient ejection pins, evenly and properly located.
3. Balance the castings evenly about the sprue (if possible).
4. Check if the die spray is effective.

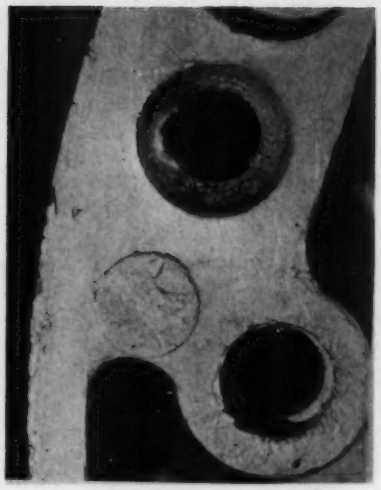


**Fig. 8 — Part of metal around hole scraped out due to contraction of metal around core.**

5. Ensure that taper and draft requirements are adequate (per A.D.C.I. standards[1]).
6. Determine if inserts or cores hinder ejection.

## DEFECTS OBSERVED SUBSEQUENT TO INJECTION

### *Table 20 — Anodizability (Poor)*

*Classification:*

Poor surface appearance caused by nonuniform anodizability.

*Causes:*

Incorrect alloy, unclean surface, nonuniform basic surface characteristics such as variation in composition or condition of alloy constituents, streaks, stains, foliations, scratches, etc. Lubricant and oxide defects appear as dark clouded areas upon anodization. Where a surface has been buffed, it should be examined for oxides under low magnification as color dyeing after anodization accentuates this defect.

*Corrections:*

1. Pure Al has the best anodized appearance but is difficult to die cast. Alloys high in Si and Cu yield a poor appearance upon anodization. Use alloys low in Si (max. 0.75 per cent) and Cu (max. 2.5 per cent). Al-Mg alloys on the other hand do not produce a clear finish. Small variations in composition cause a change in coloring.
2. Inadequate cleaning of die coat, lubricant, fingerprints, etc., can cause a patchy surface. Ensure correct die lubrication during injection and thorough cleaning immediately prior to anodizing.
3. Eliminate any of the factors that could cause flow lines (refer).
4. Scratches probably marring the surface (refer).

*(Table 20 — continued)*

5. Check if staining is the basic cause (refer), or streaking (refer).
6. Determine if foliation is responsible (refer) or poor polishability (refer).
7. Inhomogenous treatment, improper production practice, e.g., nonuniform die temperature or heterogenities in the metal itself (to which a casting is more susceptible than sheet or extrusion, etc.) tend to show up on anodizing. A homogenizing anneal in the range of 900-1000 F (482-538C) for 24 hr, depending on the alloy, sometimes can be of benefit (if anodizability is more important than any other property).

### *Table 21 — Extruded Surface Defects*

*Classification:*

Whiskers or mushroom shaped droplets on casting surface.

*Causes:*

Pressure in metal forcing the higher melting point liquid complex from the interior through the solidified surface dendrites due to:

1. Usage of certain types of alloys (e.g., 195 contains 4.5 per cent Cu, however, extruded droplets sometimes contain 30 per cent Cu).
2. Gas absorption by the molten metal.
3. Heavy section thickness.
4. Chilling of surface.
5. Premature removal of casting from die mold.

*(Table 21 — continued)*

*Corrections:*

1. Use standard recommended die casting alloys.
2. Correct the conditions that cause gas absorption during melting, holding and pouring – degas thoroughly.
3. Avoid heavy sections that stay liquid for extended periods of time.



**Fig. 9 — Tiny Surface Defects.**

4. Do not cause excessive contraction by chilling too fast. Keep the die relatively warm.
5. Reduce rate of casting ejection. However, this reduces productivity and should only be done as the last resort.

### *Table 22 — Foliations*

*Classification:*

Small leaflike metallic layers partially fused to the casting surface.

*Causes:*

Formed when a warmer metal stream meets a chilled metal deposit on the die walls, penetrates between the deposit and the wall, forming a thin layer of unfused metal over it. Also occurs when the thin fore-runners flowing along the die walls freeze and are covered a little later by streams of molten metal.

*(Table 22 — continued)*

*Corrections:*

1. Ensure that the die cavity is thoroughly clean before each injection.
2. Raise the die temperature to its correct level.
3. Increase ingate size to minimize the fore-runner effect.

### *Table 23 — Hard Spots*

*Classification:*

Tiny areas on the casting that cause excessive tool wear.

*Cause:*

Presence of harder-than-normal locations in the base metal.

*Corrections:*

1. Use correct melting procedure, minimizing any turbulence, agitation or overheating.
2. Keep the molten metal surface thoroughly clean. The $Al_2O_3$ formed on the surface by oxidation and agitation coagulates and hardens into a crust upon continued exposure. This gets incorporated into the molten metal and forms hard spots.
3. Ensure that sludging is not responsible (due to the presence of intermetallic compounds of Cr, Ti, Mn or Fe).
4. Check if inclusions are non-metallic in nature. If so, freshly rammed furnace lining may be responsible.
5. Conglomerated masses of $Al_2O_3$ severely impair machining. Clean alloy with dry Cl gas (in alloys requiring magnesium, loss should be compensated for by a slight initial excess). This treatment is usually effective.

### *Table 24 — Machinability (Poor)*

*Classification:*

Alloy causing excessive tool wear during machining.

*Causes:*

1. Use of incorrect alloy, e.g.,
   a) Increasing percentages of Si and Fe decrease machinability.
   b) Increasing percentages of Cu (up to 12 per cent), Mg (up to 10 per cent), Ni, Sn and Pb, within limits, improve machinability.
2. Presence of $Al_2O_3$ in metal.
3. Porosity due to absorbed gas or turbulence.
4. Presence of hard spots.

*Corrections:*

1. Employ a more suitable alloy, if other conditions are favorable.
2. Eliminate drossing and formation of $Al_2O_3$ (refer).
3. Eliminate causes of porosity (refer).
4. Correct hard spots (refer).

### *Table 25 — Machined Appearance (Poor)*

*Classification:*

Usually irregularities, often not due to oxide hard spots.

*Causes:*

Presence of sludge particles and other inhomogenities in metal.

*Corrections:*

1. A machined surface is not unduly disrupted by unseen particles of sludge. However, the surface displays varying degrees of brightness due to sludge precipitation. Basic causes of sludging must be eliminated (refer).
2. Remove causes of gas absorption and porosity (refer).
3. Check if hard spots (refer) cause drag lines.

### *Table 26 — Mechanical Properties (Low)*

*Classification:*

An inherent loss of strength.

*Causes:*

Weakness of the casting due to:

1. Wrong composition.
2. Unsoundness due to porosity or inclusions.
3. Other incorrect treatment.

*Corrections:*

1. Analyze the molten alloy regularly for impurities in excess of the maximum specified, e.g., Si in high Mg or Mg-Zn alloys, or Mg in high Si alloys or excessive Fe, etc., in alloys where highest elongation is desired. Bring any impurity over the designated limit to within tolerance by dilution or usage of high grade ingots.
2. Based on the above, make sure that there is no chance of the re-usable returns of different alloys getting mixed.
3. In high Fe alloys (over one per cent addition of Mn improves ductility and strength. However, over 0.3 per cent Mn, castability decreases, sludging increases.
4. From the above analysis, also check regularly whether the main constituents are within the desired range. If not, sludging might be present (refer).
5. Avoid overheating of the die or metal that will result in a coarse grain structure. Grain refining is usually not necessary because of the die chilling effect. Coarse grains are sometimes caused by excessive Fe (over 1.4 per cent in die castings) specially in high Si alloys.
6. Eliminate the factors that would form oxides and result in oxide inclusions by fluxing molten metal with $Cl_2$ and avoiding storage of ingot outdoors or in contact with chlorine or other furnace fumes. In developing "oxide lines" or areas where oxide conglomeration forms "microstress raisers,"

*(Table 26 — continued)*

the casting can lose a considerable percentage (up to 40 per cent) of its mechanical properties.

7. Correct the causes of porosity.
8. Over modification with Na lowers properties. Modification is not normally required for die castings.
9. Die castings are not usually heat treated, but if so designated, do not overage above the recommended temperatures and times.
10. In higher Mg alloys, avoid excessive degassing with $Cl_2$ as this tends to remove Mg.
11. If the die injection pressure is low or nonexistent, e.g., if a thick (approx. 0.020 in.) flash is obtained, the metal is not at its designated high pressure and instead of obtaining a high strength die casting a lower strength permanent mold casting is obtained. Therefore, make sure that normal die casting pressures are employed and that excessive flash is not formed.
12. Sound die castings usually have superior mechanical properties. However, if low mechanical properties are observed and they are not caused by porosity, a low temperature heat treatment of parts may increase mechanical properties to their normal desired ranges. Generally stress relieving a part at about 300-350 F (148-177 C) for 5-10 hr (approx.) improves properties.

### *Table 27 — Platability (Poor)*

*Classification:*

Poor surface appearance upon plating, due to inferior adherence.

*Causes:*

Presence of nonmetallics in casting surface, e.g., tiny particles or agglomerated masses of $Al_2O_3$. Poor basic surface or improper surface treatment (which also results in poor polishability).

*Corrections:*

1. Use $Cl_2$ gas for adequate fluxing of oxide particles during melting.
2. As unrefined solvent in the die lubricant causes oxide-slag lines due to unvolatilized hydrocarbons, these act as hard spots during polishing and nodes during plating. Hence the solvent employed should be highly refined and must completely volatilize below the regular die temperatures, 450-600 F (232-316 C).
3. Though the segregation of high melting point constituents (sludging) does not always disrupt a machined surface, an uneven sheen and surface often results, which leaves a surface with ridges. This effect is magnified upon plating. Factors which cause sludging must be eliminated.
4. Furthermore, as the plating bath is set for adherence to an Al base, any Fe-Mn-Si-Al complex will cause a bubble effect in the casting. Again the factors which cause sludging must be eliminated.
5. Degrease and clean surface well before plating.

### *Table 28 — Polishability (Poor)*

*Classification:*

Casting surface having a rough, dull or nonuniform appearance.

*Causes:*

Rough die surface due to wear, resulting from molten alloy impact and/or constant wear caused by mode of injection. Failure to polish and maintain the dies at frequent, regular intervals. Excessive die or metal temperatures. Incorrect alloy. Factors causing defects like roughness, die soldering, hard spots flow lines, etc.



**Fig. 10 — Defects in casting surface caused by hair cracks in die.**

*Corrections:*

1. If necessary, correct die design to minimize die surface wear by molten alloy injection.
2. Obtain the highest die hardness commercially feasible, by heat treating the whole die without taking undue chances of cracking the same.
3. Nitrided and surface hardened cavity areas tend to dissolve or flake off. They also have to be smooth. Usually employed only on cores, etc.
4. Maintain consistently the degree of die surface finish necessary for obtaining a good casting surface.
5. Investigate a better die cavity coat that could possibly reduce die wear.

*(Table 28 — continued)*

6. Ensure that excessive alloy and die temperatures are not causative.
7. Employ an alloy with better polishability, e.g., generally speaking, alloys high in magnesium have good polishability, followed closely by those high in zinc and to a lesser extent nickel and copper. With an increase in Si content, this property deteriorates.
8. Check if die soldering, flow lines, seams, roughness and poor anodizability, etc., are responsible.
9. Tiny particles or conglomerated masses of $Al_2O_3$ cause tearing and pile up of miniature "pinnacles." Hence cleanse molten metal of oxides by $Cl_2$ gas flux.
10. Unvolatilized hydrocarbon resulting from the unrefined solvent in a lubricant can cause oxide-slag lines or hard spots. Use correct solvents that are highly refined which volatilize below normal die temperatures, 450-600 F (232-600 C).
11. Sludging in an alloy gives a buffed surface with an uneven sheen and ridges. Eliminate factors causing sludging.

### *Table 29 — Scratches*

*Classification:*

Casting surface marred by obstruction to casting ejection.



**Fig. 11 — Scratches observed on the side of a casting due to contraction on and withdrawal from a core.**

*(Table 29 — continued)*

*Causes:*

Incorrect die design or injection procedure.

*Corrections:*

1. If the scratches are always in the same location, check and correct the die.
2. If they are not always in the same location but are oriented in the direction of ejection from the die, usually deep at the starting end, short in length and tapering off until they disappear, they are caused by metal splash during injection. Alter the shot velocity and/or open the gate so that no splash can precede the main metal shot.

### *Table 30 — Stains*

*Classification:*

Discoloration of the surface.

*Causes:*

Presence of $CaCO_3$, $CaSO_4$, $CaCl_2$ or free Fe in water soluble mold coating or coolant or casting quenching mediums. Graphitic suspensoid type of coatings, lubricant contamination and high temperatures also cause this effect.

*Corrections:*

1. If water soluble coolant is used, sequester the offending material by adding 8 oz of sodium hexametaphosphate to 50 gallons of coolant.
2. Use a mineral seal or kerosene for dipping or spraying the casting.
3. If observed directly after ejection from the die, check whether caused by lubricant contamination or die coat. Eliminate the cause.
4. If excessive metal and die temperatures are responsible, check and correct.

### *Table 31 — Streaks*

*Classification:*

Streaks observed usually in front of a gate, more grayish in color than the base metal, both of them possessing the same chemical composition.

*(Table 31 — continued)*

*Cause:*

First spurt of metal into the die solidifies immediately and is covered by rest of the metal. This causes a difference in the dispersion size of the Si particles.

*Corrections:*

1. Change size and/or location of ingates in a way to prevent presquirting of the metal.
2. Change direction of injection into cavity.
3. Raise temperature of the die and make it more uniform in every part of the die cavity.
4. Check if lower injecting piston speeds (keeping pressure the same) will decrease this defect.

### *Table 32 — Warpage*

*Classification:*

Castings distorted over their permissible tolerances.

*Causes:*

1. Uneven contraction during solidification or eventual cooling.
2. Unbalancing of internal stresses by machining.
3. Uneven physical support when cooling.

*Corrections:*

1. Improve faulty mold design (if possible) such as abrupt changes in section, unbalanced casting design, etc., shown by uneven location of ribs, bosses, etc.
2. Check for uneven ejection of the casting from the die.
3. Eliminate improper (uneven or sudden) cooling of the casting after ejection.
4. Ejected castings should not be stacked with uneven support or piled haphazardly, especially if they are large and bulky.
5. If warping is accentuated upon heat treatment (when specified) place castings in or on an adequate fixture.
6. When heat treatment is required, use a less drastic operation (if possible).
7. To correct warpage, heat to 600-650 F (316-343 C), clamp in a correcting vise or fixture and allow to slow cool. Keep in mind any prior or subsequent heat treatment.
8. As far as possible, avoid machining the outer skin causing unbalancing of the internal stresses.

## CONCLUSION

As with any other program of product improvement, quality control or scrap reduction, a system for betterment necessitates a well organized plan. It must be thorough, precise, yet simple. Since operating conditions are dynamic, each problem has its own peculiarities and will merit individual consideration based on the varied recommendations offered above.

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# SHOOTER, CEROXIDE, CERMET OR SLAG

by C. A. Sanders, H. J. Heine and R. F. Marande

## ABSTRACT

**It has been found that porosity, pin holes and gas holes are often associated with a fluid and gassy slag, which can usually be seen floating on the iron along the walls of delivery and pouring ladles. The nature and origin of the fluid slag is discussed at some length. Effective and practical means of controlling this potential threat to good castings are suggested. Good ladle practice including lining, care and skimming are the most important control methods.**

## INTRODUCTION

In the past few years malleable foundries have given many different reasons for porosity, pin holes and gas holes. A previous investigator[1] has divided pin holes into two different causes—"evolution" and "reaction." It has been felt that reaction pin holes warrant a closer investigation. Recently, several observers have begun to associate these defects with the presence of certain types of slag, especially in bull ladles and pouring ladles. Slag inclusions as such take their toll in scrap castings, but often inclusions from many sources are lumped together under the heading of "dirt."

This paper will attempt to isolate one type of slag that is particularly annoying to foundrymen because of its extreme fluidity at ordinary pouring temperatures. This type of slag is always a potential threat, even though it appears only in sprues and runners but not in the castings. An effective and practical means of controlling this fluid slag before it reaches the mold will be suggested. This thin, watery slag has appeared on distribution and pouring ladles, and has been given different names by different observers. Some of these names appear in the title, and all are indicative of the observed behavior or property of this type of slag.

The name "shooter" derives from the fact that this slag is usually associated with gas, and can sometimes be seen in the ladle or pouring basin actually evolving gas and jumping about in a wild state of agitation. The names "cermet" and "ceroxide" refer to the composition found by some chemists. The former suggests a combination of ceramic and metallic origin, while the latter suggests a combination of ceramic oxides. However, the writers prefer the more prosaic name of "gas-slag" as describing more closely the observed effects.




**Fig. 1 — Slag defect removed from the top of a riser.**

One type of slag which was found on top of a riser after solidification is shown in Fig. 1.

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Fig. 2 — Ceroxide defect in a steel casting.

## CEROXIDES IN STEEL FOUNDRIES

Many steel foundrymen claim that the "ceroxide" inclusions found in steel castings are peculiar to their industry. Sanders was among the first to recognize and describe the "snotter" defect.[2] This unpopular defect has been shown to be a slag inclusion,[2,3,4] which appears principally on the upper surfaces of steel castings and is often more serious in small than in large castings. This is because the inclusion may be found below the surface on the smaller or thinner section castings.

It has now been confirmed by petrographic analysis that the so called "snotter" is a slag inclusion, and its origin is likely in the deoxidation process. This is particularly true where aluminum is the deoxidant. An example of a "ceroxide" defect in a steel casting is shown in Fig. 2.

## CERMETS OR GAS-SLAG IN MALLEABLE PRACTICE

In recent years the malleable industry has begun to realize that it too has a slag problem. By slag we do not mean the ordinary viscous slag that can be easily removed from the ladle with a skimming rod. We refer to the thin, watery slag that is sometimes difficult to see and defies most conventional methods of removal. Several foundries have volunteered their observed information and remedies for this troublesome problem. It has been found that if this thin slag is not controlled it may result in a high incidence of scrap due to gas and pin hole porosity and blows.

Numerous malleable castings, sprues and runners have been studied, which clearly show the results of the gas-slag condition. If the risers and runners are examined before cleaning, the defect, if present, will be found covered with a scale of varying thickness usually on the upper or cope side. When the scale is broken with a suitable scraper, a smooth surface is usually found covered with a blue oxide film. There is always a space between the scale that can be broken off and the smooth surface underneath. It is presumed that this space was formed by gas. The brittle scale portion of this defect seems to adhere and merge completely with the sand or core surface with which it has come in contact.

## SPRUE AND RUNNER STUDIES

It has been observed that a thin slag sometimes forms at the edges and lips of pouring ladles and flows into the mold. This slag has been described as having the appearance of molten wax. Sprues and runners have been checked on molds that this slag entered during pouring. Some observers claim that during slag inclusion troubles, feeder heads may also show evidence of these unusually fluid slags. These slags will remain liquid even after the metal has solidified. An example of this condition has been found in Y, T and L sections where the slag, if present, will exude as the metal solidifies, leaving the slag deposit on the outside. A typical appearance of this type is shown in Fig. 3.

The analyses of a number of slags, which have been taken from runners and sprues and associated with blows, gas and pin holes, were found to vary within the limits shown in Table 1.

TABLE 1 — GAS-SLAG FOUND IN RUNNERS AND SPRUES

| | |
|---|---|
| $SiO_2$, % | 40 to 54 |
| $Al_2O_3$, % | 6 to 21 |
| FeO, % | 20 to 26 |
| MnO, % | 6 to 17 |
| CaO, % | 0.1 (approx.) |
| MgO, % | 1 to 5 |

During the same period another sample was taken of the mushy accumulation formed in the bottom of a pouring ladle while being preheated. This sample was analyzed for iron oxide only and was found to contain 34.6 per cent.

## FURTHER GAS-SLAG STUDIES

Many foundrymen have observed the buildup and accumulation near the top of pouring and distribution ladles. This buildup can get so severe that it will actually cut down the capacity of the ladles. Samples of this buildup material have been taken from the pouring and distribution ladles with the results shown in Table 2.

It is important to note the high proportion of manganese oxide and iron oxide in these samples. Both of these oxides greatly increase the fluidity of slags in which they are found. Furthermore, the oxide can

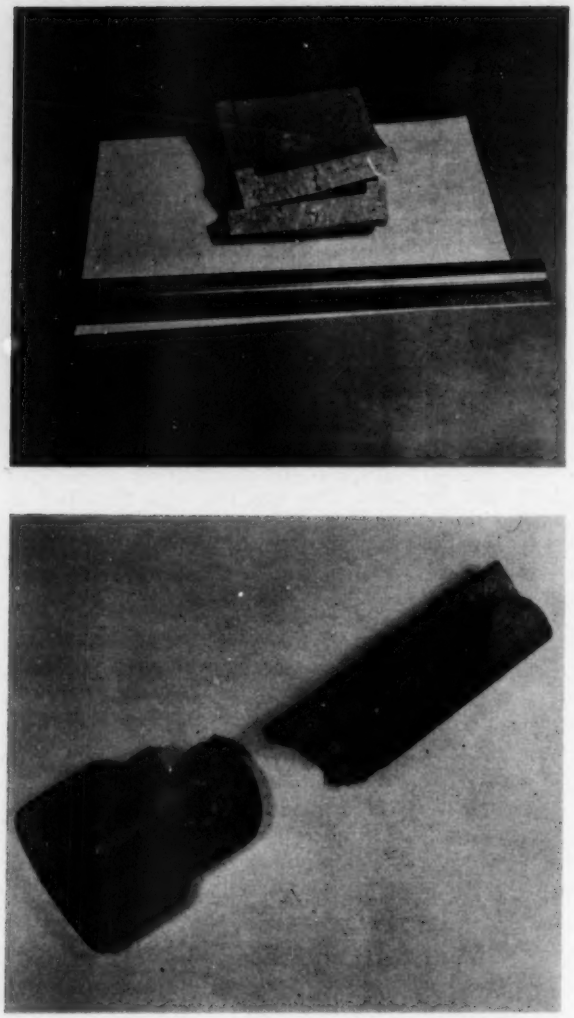


Fig. 3 — Typical defects caused by ceroxide or gas-slag in malleable iron.

TABLE 2 — LADLE BUILDUP ACCUMULATION

| | Pouring Ladle, % | Distribution Ladle, % |
|---|---|---|
| $SiO_2$ | 63.3 | 72.4 |
| $Al_2O_3$ | 10.2 | 10.3 |
| FeO | 20.6 | 14.3 |
| MnO | 2.4 | 1.8 |

react with carbon present in the iron and in the mold to generate gas as follows:

$$FeO + C \rightarrow CO + Fe$$

The relationship between the observed facts and the form of the gas-slag defect seem more than coincidental. Furthermore, the high iron oxide content of the buildup in the ladles fits the observation of a fluid slag and gas in the gas-slag defect as well as the analysis of the defect scale. Foundrymen have had a tendency to ignore the reactions between metals and refractories until quite recently. However, it was shown by Colligan[4] that a reaction may occur between iron and silica through the process of metal oxidation resulting in FeO.

## AIR FURNACE SLAGS

The possibility that this fluid slag could come from the furnace should not be overlooked. Furnace linings are open to attack from many sources, such as the ash from the coal used as fuel or from soda ash used as a desulfurizer in the furnace. If soda ash is used, it should not be left on the bath more than 15 or 20 min. It should then be thoroughly removed by skimming. If a fluid slag persists, a ground brick grog or other siliceous material added to the bath is used by some operators to cut down erosion and facilitate removal.

In general, however, it is believed that good malleable air furnace practice will effectively prevent appreciable amounts of slag from being carried through the tap hole with the metal. The slag which collects on the furnace bath should be removed as it forms and not allowed to accumulate. The slag should be removed as completely as possible toward the end of the heat, so that a considerable amount will not be left in the furnace after it is drained. The analysis of the iron should be controlled at the end of the heat in other ways than by a slag cover.

All the iron should be removed from the furnace at the end of each day's run to minimize oxidation on reheating. A typical furnace slag skimmed from an air furnace used for duplexing malleable iron is:

| | Analysis, % |
|---|---|
| $SiO_2$ | 45.50 |
| $Al_2O_3$ | 32.46 |
| FeO | 12.24 |
| MnO | 4.18 |
| CaO | 0.42 |
| MgO | 3.04 |

It will be noted that this slag also has a high amount of iron oxide and manganese oxide similar to that found on the ladles. Any fluid slag forming along the walls of the furnace, however, should be swept toward the back due to the action of the flame and combustion gases. Also the iron is tapped from the hole several inches below the surface of the furnace bath, which should preclude the possibility of an appreciable amount of the fluid slag being carried out of the tap hole.

Furthermore, if much of this fluid slag were leaving the furnace through the tap holes, all experience with slags indicates that this would soon enlarge the holes. This is not the case, since the tap out blocks are used for at least five days without any sign of enlargement. Oxidized metal can also collect along the sides of the furnace spout. This should be knocked off as often as possible before it has a chance to penetrate the refractory and contribute to the slag problem in the ladles.

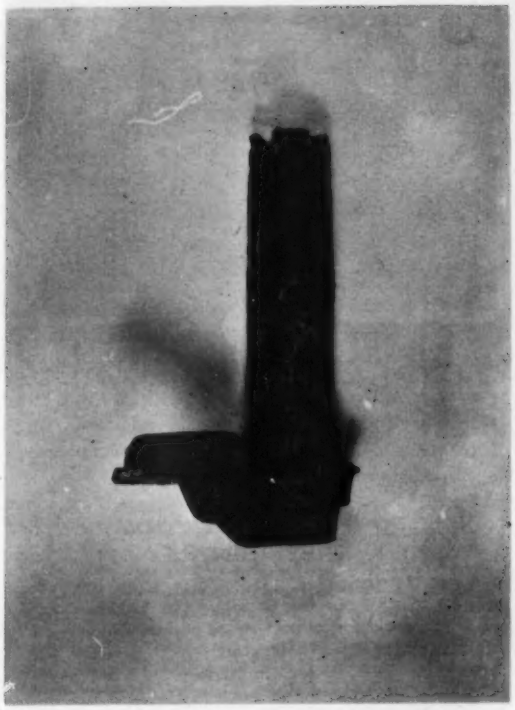


Fig. 4 — Downsprue, after cleaning removed the slag.

## FERRO-ALLOYS, OXIDATION AND DEOXIDATION

Some foundries have reported that excessive additions of ferromanganese and/or ferrosilicon to liquid white iron can be responsible for creating slag problems. In this case, the cause and the remedy lie in the conditions that make it necessary to add large amounts of ferromanganese and ferrosilicon to the air furnace. The cause is usually excessive oxidation in the cupola in duplexing processes or faulty flame control in the cold melt air furnace process. Two of the most common causes of excessive oxidation in the cupola are bridging, and failure to keep the stack full.

Bridging can usually be corrected by an analysis of the problem. This may involve proper sizing of materials, fluxing practice, and lining practice. Failure to keep the cupola stack full is inexcusable and should not be tolerated, since it will invariably result in excessive loss of manganese and silicon. Therefore, it should be clear that any ill effects caused by large additions of silicon and manganese are the result of an oxidizing condition that should not have been allowed to occur.

Another factor that should be taken into account when excessive amounts of ferrosilicon are used is the aluminum content of the latter. This aluminum content can contribute to the slag inclusion problem in the same way that aluminum deoxidation does in steel practice. It is generally believed that the aluminum content of molten iron should be kept below 0.001 per cent to avoid certain pin hole defects. This can usually be accomplished by specifying the maximum percentage of aluminum that will be acceptable in ferrosilicon.

## GATING SYSTEMS

It is certainly important to have good gating practice. However, a good gating system which will trap most ordinary slags will not be too effective in stopping the fluid, watery, slag that has been discussed in this paper. The observations which follow will show that this fluid slag must be kept under control ahead of the mold to be effective.

### *Observations on Ladles and Pouring*

At one point during the many tests and observations on the gas-slag problem a pouring ladle in which an unusually large amount of foamy, fluid slag had formed was noticed. A mold was deliberately poured without trying to remove the slag. The sprue, runner and casting showed not only the common refractory slag but also a severe gas-slag condition.

Shortly after the above observation four molds consisting of a 10 in. by 10 in. by ½-in. plate were poured. The first two were poured with a ladle that had accumulated a large amount of fluid slag. The third mold was poured from a clean ladle, and with what appeared to be clean iron. The fourth was also poured with clean iron, except that scale from the gas-slag defect collected from the sprues and runners of previous tests was added to the ladle. This latter immediately developed a thin foamy slag before pouring.

Of these plates the first two showed both common slag and gas-slag. The third was clean. The fourth showed the gas-slag condition. The occurrence of the two types of slags in the test castings indicated two types of slag in the ladle. Close observation of the pouring and distribution ladles revealed the two types. The more common is the viscous slag, which will usually segregate and collect in the center of the ladle. The other type is that which can be seen actually forming against the ladle walls. It is sometimes foamy but is always fluid.

It is this second type which causes the gas-slag defect. This was confirmed by observing the pouring of 20 molds with ladles on which the fluid slag was present in considerable quantity. Without exception, the sprue and runners contained the gas-slag defect when the thin, fluid slag was seen entering the mold. An example of the appearance of a downsprue, after cleaning removed the slag, is shown in Fig. 4.

Another observation that was significant concerned the buildup on the walls of ladles mentioned earlier. If the distribution ladles were not skimmed, this buildup would soon reduce the capacity of the ladle appreciably. One morning a ladle with considerable buildup was filled with iron at the furnace after a suitable preheat. Immediately afterwards, two ladles newly lined with a dense silica material were filled with iron.

A large amount of fluid, foamy slag formed on the older ladle with the buildup. The two new ladles were perfectly clean. The contrast was striking. The ladle with the large amount of slag was pigged and then refilled. It still contained too much slag to be



Fig. 5 — Distribution ladle being skimmed.



Fig. 6 — Pouring ladle being skimmed.

usable and the contents were again pigged. It was then noted that the ladle capacity had considerably increased. In other words, a large amount of the buildup material had floated up as slag. The newly lined distribution ladles continued to show little slag. These observations were strong evidence showing that the origin of the fluid slag was in the ladle, and that the gas-slag defect was a product of the oxide buildup in the ladles.

### *Remedies for Control of the Gas-Slag Condition*

After reviewing the above tests and observations, it was concluded that the only practical place to control the gas-slag problem and minimize the threat to good castings, is in the ladles. This means that strict and systematic ladle practice must be set up. These items have proved helpful and important in maintaining clean ladles:

1. Number all ladles, preferably by welding on the yoke or bowl.
2. Select a refractory material that can be rammed to a dense lining.
3. A lining material that will gradually erode away is preferred to one that is too refractory.
4. Keep a record of the number of days each ladle is used.
5. Skim all distribution and pouring ladles every time they are filled with iron. This should be done as thoroughly as practical starting with the first time the ladle is used.
6. Establish a maximum number of heats after which ladles will be relined.
7. Make sure that all iron is poured out of the ladles after using. Even small amounts of iron will be oxidized readily on cooling and reheating and contribute to fluid slag formation.

Illustrations of distribution and pouring ladles being skimmed are shown in Figs. 5 and 6.

## CONCLUSIONS

1. *Cause.* The gas-slag defect is caused by a fluid slag which clings to mold walls and generates gas through chemical reaction with carbon in the iron or in the mold.
2. *Source.* The slag which causes the gas-slag defect is produced by continual self fluxing of oxide accumulations in the furnace spout, distribution ladles and pouring ladles.
3. *Control.* This fluid slag can be prevented from ever becoming a threat to good casting production by good ladle practice outlined, and by thoroughly skimming ladles to prevent any accumulation of slag.

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# GREEN SAND PROPERTIES

## *Median grain size effect*

by A. B. Draper and H. A. Knappenberger


### ABSTRACT

**Foundrymen are becoming more aware that variations in sand grain size and distribution affect both the properties of molds and the quality of castings. This paper reports the effect of median grain size on the green compression strength and permeability of molding sand.**

**Previous research failed to isolate the effect of median grain size from other distribution parameters. The present investigation overcomes this deficiency by synthesizing symmetrical sand distributions of varying median grain sizes but with equal dispersion.**

**The results support the generally accepted statement that permeability increases as median grain size increases, but they refute the theory that green compression strength increases as median grain size decreases. It follows that the green compression strength of rammed molding sand does not depend upon the total sand grain surface area or the number of sand grain points of contact.**


### INTRODUCTION

Foundrymen have long known that consistently acceptable castings require close control of molding sand properties. Such control can be achieved by good laboratory techniques if they are properly correlated with both the properties of the mold and casting defects. The foundry engineer could achieve greater control of casting quality if he better understood the effect and interaction of sand grain distribution parameters on mold properties; intuition is not enough.

At least four areas must be investigated before the sand technician will have sufficient understanding to use his molding materials to best advantage, or before a meaningful sand classification system can be developed. These are the effect upon mold properties of variations in:

1. Grain size.
2. Grain size distribution.
3. Grain shape.
4. Electrochemical forces at the sand-clay-water interface.

Schubert[1] studied several commercially available sands and reported that increases in the AFS grain fineness number decreased the AFS permeability values but had no effect on the green compression strength. Schubert's work is inconclusive, however, because the AFS grain fineness number is merely a measure of average grain size (Morey and Taylor,[2] Leaman and Ekey[3]), and he failed to control other distribution parameters such as skewness and dispersion.

This research investigated the effect of average grain size on the green compression strength and permeability of synthetically bonded green sand. The information obtained may eventually lead to the development of a meaningful sand classification system.

### PROCEDURE

Sieve fractions of Ottawa rounded sand grains were prepared in sufficient quantity to construct five different sand grain size distributions. Each had constant dispersion, no skewness, and a different median grain size. All distributions were bonded to successively higher levels of southern bentonite, mulled for a standard time and then tempered to a moisture level somewhat higher than that required for maximum green compression strength.

After test specimens were evaluated for green sand properties with standard laboratory equipment, the sand was exposed to the atmosphere and retested at

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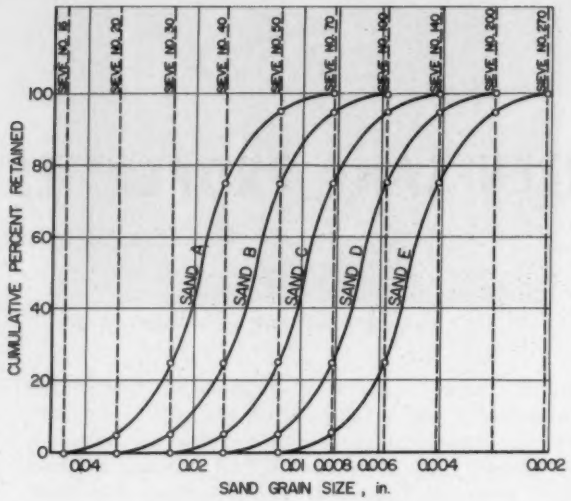



**Fig. 1 — Cumulative frequency distribution for five test sands.**

the lower moisture content. Curves for green compression strength versus moisture were established. Then the test sands were dehydrated, rebonded to a higher binder level and tested as before.

### *Building Test Distributions*

To eliminate disparities in dispersion and skewness, known sieve fractions were combined to yield symmetrical sand grain distributions with equal dispersion but varying average grain size (Table 1). The effect of grain shape was minimized by using only Ottawa silica sand, which has rounded grains. The cumulative frequency distribution curves (Fig. 1) show the similarity of the distributions and yield the median grain size directly. This is convenient because in a symmetrical distribution both the median and the arithmetic average are equal, but the latter is more difficult to determine.

### *Bonding and Tempering Test Sands*

Since several investigators had reported that southern bentonite reaches peak green compression strength in a shorter mulling time than any other binder, it was selected for this work. To include binder levels both typical of foundry practice and within the region of clay saturation (Zrimsek and Heine[4]), four bentonite levels were chosen in an approximate geometric progression, 5, 8, 12 and 20 per cent.

Contrary to customary practice, these percentages were based on the total sand and binder content rather than on the sand alone (e.g., 52.4 gm of binder and 1000 gm of sand for the first test run). After mulling the dry mixture for 5 min, water slightly in excess of that required for peak green compression strength was added and the mulling was continued for 5 min.[5]

Recent research[6] indicated that the moisture required to develop maximum green compression strength could be determined analytically. However, that method proved to be inadequate (Appendix),

**TABLE 1 — TEST SAND DISTRIBUTIONS**

| U.S. Stand. Sieve No. | Mesh Size Openings, in. | Sand A | | Sand B | | Sand C | | Sand D | | Sand E | |
|---|---|---|---|---|---|---|---|---|---|---|---|
| | | Weight | | Weight | | Weight | | Weight | | Weight | |
| | | Gm. | % | Gm. | % | Gm. | % | Gm. | % | Gm. | % |
| 20 | 0.0331 | 50 | 5 | | | | | | | | |
| 30 | 0.0232 | 200 | 20 | 50 | 5 | | | | | | |
| 40 | 0.0165 | 500 | 50 | 200 | 20 | 50 | 5 | | | | |
| 50 | 0.0117 | 200 | 20 | 500 | 50 | 200 | 20 | 50 | 5 | | |
| 70 | 0.0083 | 50 | 5 | 200 | 20 | 500 | 50 | 200 | 20 | 50 | 5 |
| 100 | 0.0059 | | | 50 | 5 | 200 | 20 | 500 | 50 | 200 | 20 |
| 140 | 0.0041 | | | | | 50 | 5 | 200 | 20 | 500 | 50 |
| 200 | 0.0029 | | | | | | | 50 | 5 | 200 | 20 |
| 270 | 0.0021 | | | | | | | | | 50 | 5 |
| Median grain size, in. | | 0.0195 | | 0.0135 | | 0.0098 | | 0.0069 | | 0.0051 | |
| A.F.S. Grain Fineness number | | 30 | | 41 | | 54 | | 74 | | 105 | |

and the proper moisture levels for the materials used in this research were established by qualitative tests. The tests showed that a southern bentonite bonded sand tempered to maximum green compression strength appeared slightly damp but felt distinctly dry. By tempering a little beyond this point and decreasing the water content, the desired curves for green compression strength versus moisture could readily be obtained. The tempered mix was aged for at least 12 hr in sealed glass jars before being riddled through a 20 mesh sieve and tested.

### *Testing Procedure*

A standard moisture teller was used in conjunction with a desiccator for cooling the specimens before reweighing. If two tests were inconsistent, a third test was made.[5]

Four standard AFS specimens from each distribution were tested for permeability, hardness and green compression strength (Table 2), according to AFS procedures.[5] Two exceptions should be noted:

1) If the permeability was in excess of 500, the stop watch method was used.
2) If the compression strength exceeded 18.5 psi, the next specimen was tested in the upper position on the testing machine.

**TABLE 2 — TEST DATA FOR RIDDLED AND UNRIDDLED OTTAWA SAND, AFS No. 130, 2.2% MOISTURE, 6% SOUTHERN BENTONITE***

| Specimen Property | Riddled Sand | Unriddled Sand |
|---|---|---|
| AFS Permeability | 79 | 64 |
| AFS Hardness | 91 | 91 |
| Green compression strength, psi | 10.5 | 9.5 |

*All data is an average of 3 tests.

Because of the labor involved in obtaining the original sand distributions, as much sand as possible was reclaimed. To implement this policy, the sand was exposed to the atmosphere to allow some moisture (approximately 0.25 per cent) to evaporate, then

mulled for 3 min, riddled into glass jars, sealed and retested after proper aging. Thus, the green sand property variations as a function of moisture could be determined in the vicinity of the maximum green compression strength by using only three or four tests.

TABLE 3 — MAXIMUM DEVELOPMENT OF GREEN COMPRESSION STRENGTHS (psi) FOR THE FIVE TEST SANDS AT THE FOUR BINDER LEVELS

| Test Sands | Median Grain Size, in. | Green Compression Strength as a Function of S.B. Content | | | |
|---|---|---|---|---|---|
| | | 5 % | 8 % | 12 % | 20 % |
| A | 0.0195 | 14.4 | 25.8 | 29.1 | 30.4 |
| B | 0.0135 | 15.4 | 29.3 | 37.2 | 31.2 |
| C | 0.0098 | 16.3 | 29.8 | 34.9 | 29.4 |
| D | 0.0069 | 15.9 | 29.1 | 38.9 | 31.7 |
| E | 0.0051 | 17.4 | 30.5 | 36.4 | 30.1 |

The repeated use of the sand grains could be criticized on the following grounds:

1) Excessive mulling might crush the sand grains, changing the grain size distribution and the average activity of the grain surface.
2) The green compression strength of the sand might increase with increased mulling.

It was unlikely that significant crushing of the sand grains could occur, since the muller wheels were rubber. In any case, the permeability did not decrease with increased mulling. The second objection must be invalid, for this research established that there was no significant difference in green compression strength, regardless of cause, within the limits of the experiment. It should also be remembered here that southern bentonite was chosen as a binder for the precise reason that it reaches its peak compression strength with minimum mulling.

TABLE 4 — ANALYSIS OF VARIANCE FOR THE DATA IN TABLE 3

| Source of Variation | Sum of Squares of Deviations | Degrees of Freedom | Mean Square Deviation | F Ratio | $F_{95}$ | $F_{99}$ | Significant? 95% | Significant? 99% |
|---|---|---|---|---|---|---|---|---|
| Median grain size | 41.133 | 4 | 10.283 | 0.107 | 3.26 | 5.41 | No | No |
| Bentonite level | 1035.428 | 3 | 345.143 | 3.7 | 3.49 | 5.95 | Yes | No |
| Residual | 1113.188 | 12 | 92.766 | | | | | |
| Total | 2189.749 | 19 | | | | | | |

There were two principal advantages in reusing the sand. Variations in sand distribution from test to test were constant, and variations in mesh size openings caused by sieving large quantities of sand were minimized.

### *Increasing the Bond Content*

Before increasing the bond content, the sand was completely dehydrated in an electric oven at 215 F. Calcium chloride desiccant was placed in the oven during cooling to prevent the readsorption of moisture. Assuming that 5 per cent of the binder remained, sufficient new southern bentonite was added to give an 8 per cent binder content. A similar procedure was followed for the sands with 12 and 20 per cent binder (Tables 5 to 9).

## EXPERIMENTAL RESULTS

Maximum green compression strength occurred at higher moisture levels as the amount of binder increased (Fig. 2). At 12 per cent binder, only three points were tested because this seemed adequate. It proved to be unwise, and the four-point procedure was resumed.

The maximum green compression strength increased with increasing bond, up to a southern bentonite content of approximately 12 per cent (Fig. 3). At 20 per cent binder, the maximum green compression strength decreased. This substantiates the existence of a binder saturation point in the region from 12 to 15 per cent.

Further analysis suggested that median grain size has no effect on the maximum development of green compression strength. The proposition was tested by an analysis of variance[7] (Tables 3 and 4). This test showed that at both the 95 and 99 per cent confidence levels the median grain size had no effect on the green compression strength.

The effect of southern bentonite level on the green compression strength was also tested. The test showed that at the 95 per cent confidence level bentonite content had a detectable effect, but at the 99 per cent confidence level it had none. This conclusion is reasonable, considering that three of the four bentonite levels tested were so close to the binder saturation range of southern bentonite that changes in binder content had little effect upon green compression strength. However, it is common experience that sands with, let us say, 3 to 8 per cent southern bentonite binder (definitely below the saturation range) do exhibit significant increases in green compression strength with increasing binder content.

### *Permeability-Moisture Relationship*

Peak permeability was not achieved in any of the tests. However, the data obtained support the generally accepted statements that maximum permeability increases as the grain size increases, and that greatest permeability occurs at a higher moisture level than does peak green compression strength.

## SUMMARY AND CONCLUSIONS

An analysis of variance indicates that at the 99 per cent confidence level median grain size has no effect on the maximum development of green compression strength in molding sands.

The data also support three commonly accepted statements relative to permeability:

1) Large sand grains produce higher permeability than smaller sand grains when bonded under identical conditions.
2) Peak permeability occurs at a moisture content higher than that required for peak green compression strength.

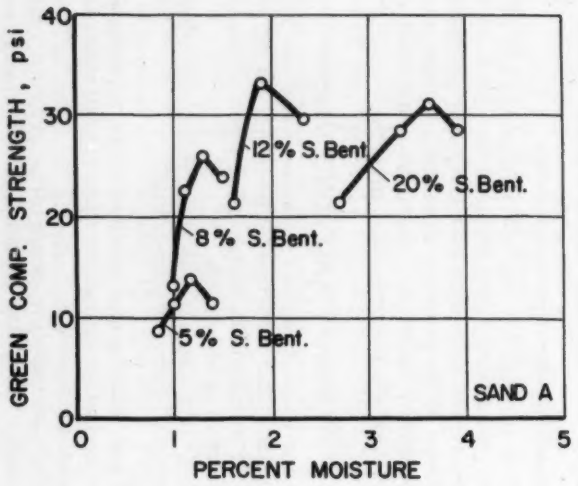



Fig. 2a and 2b — Typical green compression strength vs. moisture content for the test distributions.

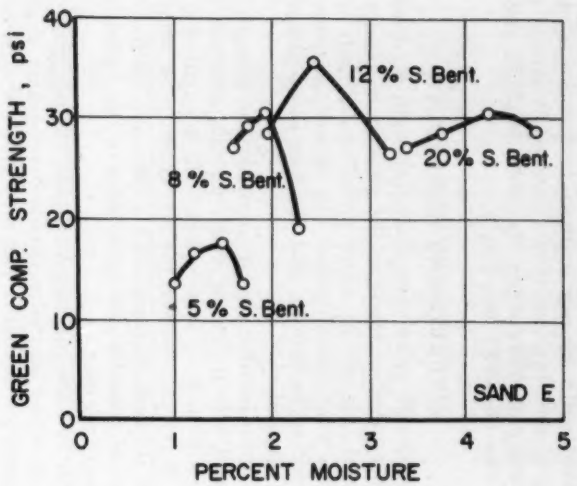



TABLE 5 — EXPERIMENTAL DATA FOR SAND A

| Sand No. 1 | S. Bent., % | Moisture, % | AFS Permeability | | | | | AFS Hardness | | | | | Compression Strength, psi | | | | | Specimen Weight, gm |
|---|---|---|---|---|---|---|---|---|---|---|---|---|---|---|---|---|---|---|
| | | | Spec. 1 | Spec. 2 | Spec. 3 | Spec. 4 | Avg. | Spec. 1 | Spec. 2 | Spec. 3 | Spec. 4 | Avg. | Spec. 1 | Spec. 2 | Spec. 3 | Spec. 4 | Avg. | |
| A-1.1 | 5 | 1.4 | 650 | 650 | 650 | 650 | 650 | 93 | 94 | 93 | 93 | 93 | 11.3 | 11.6 | 10.5 | 10.6 | 11.0 | 163 |
| A-1.2 | 5 | 1.1 | 650 | 650 | 650 | 650 | 650 | 96 | 96 | 96 | 95 | 96 | 15.1 | — | 13.3 | 14.8 | 14.4 | 164 |
| A-1.3 | 5 | 0.90 | 650 | 575 | 575 | 575 | 594 | 96 | 97 | 97 | 96 | 96 | — | 12.3 | 12.6 | 10.7 | 11.9 | 169 |
| A-1.4 | 5 | 0.70 | 550 | 575 | 500 | 500 | 531 | 99 | 98 | 99 | 98 | 98 | 7.8 | 8.0 | 9.5 | 7.0 | 8.1 | 170 |
| A-2.1 | 8 | 1.5 | 600 | 600 | 650 | 600 | 613 | 95 | 94 | 95 | 95 | 95 | 24.2 | 24.0 | 24.7 | 24.5 | 24.3 | 164 |
| A-2.2 | 8 | 1.25 | 525 | 525 | 500 | 500 | 513 | 97 | 97 | 97 | 96 | 97 | 25.8 | 26.2 | 25.5 | 25.5 | 25.8 | 169 |
| A-2.3 | 8 | 1.1 | 500 | 500 | 490 | 500 | 498 | 94 | 95 | 95 | 95 | 95 | 25.5 | 23.0 | 20.5 | 22.1 | 22.8 | 169 |
| A-2.4 | 8 | 0.9 | 450 | 450 | 500 | 575 | 494 | 93 | 94 | 93 | 93 | 93 | — | 15.2 | — | 10.1 | 12.6 | 171 |
| A-3.1 | 12 | 2.35 | 550 | 550 | 525 | 550 | 544 | 94 | 94 | 94 | 95 | 94 | 28.0 | 29.1 | 29.4 | 29.8 | 29.1 | 166 |
| A-3.2 | 12 | 1.8 | 525 | 475 | 500 | 500 | 500 | 96 | 98 | 97 | 97 | 97 | 37.0 | 38.2 | 29.8 | 28.7 | 33.4 | 169 |
| A-3.3 | 12 | 1.55 | 400 | 425 | 450 | 450 | 431 | 97 | 96 | 96 | 96 | 96 | 16.7 | 20.2 | 19.8 | 24.2 | 20.2 | 170 |
| A-4.1 | 20 | 3.9 | 575 | 600 | 600 | 650 | 606 | 93 | 93 | 93 | 93 | 93 | 28.8 | 28.2 | 29.1 | 28.9 | 28.8 | 159 |
| A-4.2 | 20 | 3.3 | 435 | 450 | 400 | — | 428 | 95 | 95 | 96 | — | 96 | 28.3 | 28.4 | 28.2 | — | 28.6 | 165 |
| A-4.3 | 20 | 2.7 | 250 | 325 | 325 | 425 | 331 | 96 | 95 | 96 | 96 | 96 | — | 20.7 | 19.5 | 21.7 | 20.6 | 168 |
| A-4.4 | 20 | 3.6 | 400 | 375 | 400 | 410 | 396 | 95 | 96 | 95 | 96 | 96 | — | 29.1 | 32.5 | 29.7 | 30.4 | 166 |

TABLE 6 — EXPERIMENTAL DATA FOR SAND B

| Sand Specimen No. | S. Bent., % | Moisture, % | AFS Permeability | | | | | AFS Hardness | | | | | Compression Strength, psi | | | | | Specimen Weight, gm |
|---|---|---|---|---|---|---|---|---|---|---|---|---|---|---|---|---|---|---|
| | | | Spec. 1 | Spec. 2 | Spec. 3 | Spec. 4 | Avg. | Spec. 1 | Spec. 2 | Spec. 3 | Spec. 4 | Avg. | Spec. 1 | Spec. 2 | Spec. 3 | Spec. 4 | Avg. | |
| B-1.1 | 5 | 1.4 | 360 | 395 | 390 | 350 | 374 | 95 | 96 | 94 | 94 | 95 | 13.4 | 13.6 | 13.0 | 13.2 | 13.3 | 160 |
| B-1.2 | 5 | 1.3 | 350 | 350 | 325 | 330 | 339 | 97 | 97 | 97 | 97 | 97 | 14.6 | 16.1 | 15.2 | 15.7 | 15.4 | 163 |
| B-1.3 | 5 | 1.1 | 325 | 330 | 300 | 325 | 320 | 97 | 97 | 98 | 97 | 97 | 12.1 | 12.0 | 12.3 | 12.3 | 12.2 | 164 |
| B-1.4 | 5 | 1.0 | 375 | 360 | 330 | 330 | 344 | 95 | 95 | 95 | 95 | 95 | 12.1 | 12.4 | 10.6 | 10.1 | 11.3 | 161 |
| B-2.1 | 8 | 1.65 | 380 | 385 | 360 | 365 | 373 | 94 | 94 | 94 | 94 | 94 | 24.0 | 24.7 | 25.5 | 24.6 | 24.7 | 161 |
| B-2.2 | 8 | 1.4 | 265 | 300 | 280 | 275 | 280 | 97 | 97 | 97 | 97 | 97 | 27.5 | 29.5 | 29.7 | 30.5 | 29.1 | 167 |
| B-2.3 | 8 | 1.3 | 250 | 250 | 240 | 255 | 254 | 96 | 97 | 95 | 96 | 96 | 31.5 | 29.5 | 27.0 | 28.5 | 29.3 | 169 |
| B-2.4 | 8 | 1.1 | 260 | 240 | 245 | 250 | 249 | 96 | 96 | 96 | 96 | 96 | 16.2 | 22.6 | 17.4 | 16.7 | 18.2 | 169 |
| B-3.1 | 12 | 2.7 | 375 | 325 | 350 | 375 | 356 | 93 | 93 | 93 | 94 | 93 | 29.7 | 29.4 | 29.0 | 29.2 | 29.3 | 160 |
| B-3.2 | 12 | 2.1 | 280 | 240 | 260 | 265 | 261 | 97 | 97 | 98 | 97 | 97 | 36.0 | 38.0 | 37.1 | 37.7 | 37.2 | 167 |
| B-3.3 | 12 | 1.65 | 195 | 198 | 215 | 220 | 207 | 98 | 97 | 98 | 97 | 98 | 26.1 | 35.5 | 23.7 | 29.7 | 28.8 | 168 |
| B-4.1 | 20 | 4.4 | 425 | 450 | 425 | — | 433 | 90 | 90 | 90 | — | 90 | 27.0 | 29.2 | 27.8 | — | 28.0 | 152 |
| B-4.2 | 20 | 3.8 | 230 | 225 | 250 | 245 | 238 | 95 | 96 | 95 | 95 | 95 | 30.2 | 29.8 | 31.0 | 29.0 | 30.0 | 161 |
| B-4.3 | 20 | 3.4 | 160 | 145 | 170 | 180 | 164 | 97 | 96 | 96 | 96 | 96 | 27.8 | 26.7 | 25.7 | 26.6 | 26.7 | 167 |
| B-4.4 | 20 | 4.05 | 320 | 325 | 350 | 330 | 331 | 94 | 94 | 95 | 93 | 94 | 30.3 | 29.8 | 32.7 | 32.1 | 31.2 | 159 |

TABLE 7 — EXPERIMENTAL DATA FOR SAND C

| Sand Specimen No. | S. Bent., % | Moisture, % | AFS Permeability | | | | | AFS Hardness | | | | | Compression Strength, psi | | | | | Specimen Weight, gm |
|---|---|---|---|---|---|---|---|---|---|---|---|---|---|---|---|---|---|---|
| | | | Spec. 1 | Spec. 2 | Spec. 3 | Spec. 4 | Avg. | Spec. 1 | Spec. 2 | Spec. 3 | Spec. 4 | Avg. | Spec. 1 | Spec. 2 | Spec. 3 | Spec. 4 | Avg. | |
| C-1.1 | 5 | 1.6 | 180 | 190 | 195 | 195 | 190 | 95 | 94 | 95 | 95 | 95 | 14.5 | 13.4 | 14.4 | 13.7 | 14.0 | 157 |
| C-1.2 | 5 | 1.4 | 175 | 185 | 180 | 175 | 179 | 95 | 97 | 96 | 97 | 96 | 16.1 | 16.6 | 16.4 | 16.3 | 16.3 | 160 |
| C-1.3 | 5 | 1.1 | 150 | 150 | 150 | 155 | 151 | 96 | 97 | 95 | 96 | 96 | 12.6 | 14.8 | 12.9 | 12.1 | 13.1 | 163 |
| C-1.4 | 5 | 0.9 | 125 | 130 | 125 | 120 | 125 | 96 | 97 | 96 | 98 | 97 | 6.4 | 7.6 | 6.7 | 7.1 | 6.9 | 165 |
| C-2.1 | 8 | 1.7 | 165 | 155 | 150 | 150 | 155 | 95 | 94 | 95 | 95 | 95 | 25.2 | 25.8 | 24.5 | 26.1 | 25.4 | 159 |
| C-2.2 | 8 | 1.6 | 130 | 127 | 125 | 130 | 128 | 96 | 96 | 96 | 96 | 96 | 30.5 | 30.5 | 29.5 | 28.7 | 29.8 | 167 |
| C-2.3 | 8 | 1.4 | 130 | 127 | 120 | 125 | 126 | 95 | 96 | 95 | 97 | 96 | 28.4 | 29.8 | 28.7 | 28.2 | 28.8 | 166 |
| C-2.4 | 8 | 1.1 | 127 | 110 | 125 | 120 | 121 | 96 | 96 | 97 | 97 | 97 | 17.8 | 20.6 | 20.4 | 18.2 | 19.2 | 167 |
| C-3.1 | 12 | 2.6 | 190 | 180 | 185 | 180 | 184 | 94 | 92 | 93 | 93 | 93 | 29.5 | 30.5 | 30.2 | 30.6 | 30.2 | 158 |
| C-3.2 | 12 | 2.2 | 115 | 118 | 115 | 120 | 117 | 98 | 98 | 97 | 97 | 98 | — | 39.5 | 32.3 | 32.8 | 34.9 | 167 |
| C-3.3 | 12 | 1.4 | 95 | 95 | 95 | 100 | 96 | 97 | 98 | 97 | 97 | 97 | 29.8 | 29.7 | 27.7 | 26.0 | 28.3 | 170 |
| C-4.1 | 20 | 4.5 | 185 | 187 | 182 | 180 | 184 | 92 | 93 | 93 | 93 | 93 | 27.7 | 29.5 | 28.4 | 29.1 | 28.7 | 153 |
| C-4.2 | 20 | 3.9 | 105 | 100 | 105 | 102 | 103 | 96 | 95 | 95 | 96 | 96 | — | 29.4 | 29.2 | 29.7 | 29.4 | 163 |
| C-4.3 | 20 | 3.5 | 70 | 68 | 75 | 70 | 74 | 96 | 97 | 97 | 97 | 97 | 30.7 | 29.5 | 29.0 | 27.1 | 29.1 | 166 |
| C-4.4 | 20 | 3.1 | 62 | 65 | 63 | 65 | 61 | 96 | 97 | 97 | 96 | 96 | 26.2 | 25.3 | 25.8 | 26.0 | 25.8 | 168 |

TABLE 8 — EXPERIMENTAL DATA FOR SAND D

| Sand Specimen No. | S. Bent., % | Moisture, % | AFS Permeability | | | | | AFS Hardness | | | | | Compression Strength, psi | | | | | Specimen Weight, gm |
|---|---|---|---|---|---|---|---|---|---|---|---|---|---|---|---|---|---|---|
| | | | Spec. 1 | Spec. 2 | Spec. 3 | Spec. 4 | Avg. | Spec. 1 | Spec. 2 | Spec. 3 | Spec. 4 | Avg. | Spec. 1 | Spec. 2 | Spec. 3 | Spec. 4 | Avg. | |
| D-1.1 | 5 | 1.7 | 110 | 105 | 102 | 102 | 105 | 94 | 94 | 94 | 93 | 94 | 12.8 | 12.4 | 12.3 | 12.2 | 12.4 | 155 |
| D-1.2 | 5 | 1.5 | 97 | 96 | 97 | 97 | 97 | 97 | 96 | 95 | 96 | 96 | 16.2 | 15.5 | 15.9 | 16.1 | 15.9 | 158 |
| D-1.3 | 5 | 1.3 | 95 | 94 | 92 | 93 | 94 | 95 | 96 | 95 | 96 | 96 | 13.1 | 12.6 | 12.6 | 13.9 | 13.1 | 157 |
| D-1.4 | 5 | 1.2 | 94 | 93 | 95 | 95 | 94 | 95 | 95 | 96 | 95 | 95 | 11.7 | 11.0 | 11.9 | 11.9 | 11.6 | 156 |
| D-2.1 | 8 | 2.3 | 105 | 106 | 101 | 102 | 104 | 90 | 90 | 90 | 90 | 90 | 20.5 | 19.5 | 19.5 | 19.5 | 19.8 | 154 |
| D-2.2 | 8 | 1.9 | 83 | 84 | 82 | 82 | 83 | 94 | 95 | 95 | 95 | 95 | 27.7 | 29.8 | 29.6 | 28.0 | 28.5 | 160 |
| D-2.3 | 8 | 1.8 | 75 | 76 | 75 | 76 | 76 | 94 | 95 | 95 | 95 | 95 | 30.8 | 28.5 | 29.0 | 28.2 | 29.1 | 161 |
| D-2.4 | 8 | 1.6 | 60 | 62 | 61 | 60 | 61 | 96 | 96 | 97 | 96 | 96 | 26.8 | 24.6 | 27.6 | 26.2 | 26.3 | 165 |
| D-3.1 | 12 | 2.9 | 120 | 112 | 115 | 112 | 115 | 91 | 92 | 91 | 91 | 91 | 26.0 | 26.5 | 26.2 | 26.7 | 26.3 | 153 |
| D-3.2 | 12 | 2.4 | 67 | 66 | 65 | 68 | 67 | 97 | 96 | 97 | 97 | 97 | 38.7 | 40.9 | 37.6 | 38.3 | 38.9 | 162 |
| D-3.3 | 12 | 2.0 | 54 | 52 | 54 | 56 | 54 | 96 | 97 | 97 | 97 | 97 | 30.5 | 35.1 | 30.4 | 33.7 | 32.4 | 167 |
| D-4.1 | 20 | 4.9 | 145 | 150 | 140 | 145 | 145 | 90 | 89 | 90 | 89 | 90 | 28.6 | 28.4 | 29.5 | 29.1 | 28.9 | 147 |
| D-4.2 | 20 | 4.05 | 72 | 72 | 74 | 74 | 73 | 94 | 94 | 94 | 95 | 94 | — | 31.5 | 30.0 | 33.5 | 31.7 | 156 |
| D-4.3 | 20 | 3.7 | 47 | 48 | 51 | 50 | 49 | 96 | 96 | 96 | 96 | 96 | 28.2 | 28.7 | 30.2 | 28.7 | 28.9 | 162 |
| D-4.4 | 20 | 4.6 | 110 | 105 | 112 | 112 | 110 | 93 | 93 | 93 | 94 | 93 | 30.3 | 29.8 | 29.3 | 30.8 | 30.1 | 151 |

TABLE 9 — EXPERIMENTAL DATA FOR SAND E

| Sand Specimen No. | S. Bent., % | Moisture, % | AFS Permeability | | | | | AFS Hardness | | | | | Compression Strength, psi | | | | | Specimen Weight, gm |
|---|---|---|---|---|---|---|---|---|---|---|---|---|---|---|---|---|---|---|
| | | | Spec. 1 | Spec. 2 | Spec. 3 | Spec. 4 | Avg. | Spec. 1 | Spec. 2 | Spec. 3 | Spec. 4 | Avg. | Spec. 1 | Spec. 2 | Spec. 3 | Spec. 4 | Avg. | |
| E-1.1 | 5 | 1.7 | 61 | 61 | 60 | 59 | 60 | 94 | 94 | 93 | 94 | 94 | 13.9 | 13.5 | 13.5 | 13.0 | 13.5 | 153 |
| E-1.2 | 5 | 1.5 | 54 | 55 | 55 | 55 | 55 | 96 | 96 | 96 | 96 | 96 | 17.3 | 17.2 | 17.5 | 17.5 | 17.4 | 156 |
| E-1.3 | 5 | 1.2 | 48 | 48 | 49 | 48 | 48 | 96 | 96 | 96 | 97 | 96 | 16.5 | 15.5 | 14.9 | 16.6 | 15.9 | 157 |
| E-1.4 | 5 | 1.0 | 45 | 45 | 45 | 45 | 45 | 97 | 97 | 97 | 97 | 97 | 13.3 | 13.1 | 13.5 | 13.1 | 13.2 | 158 |
| E-2.1 | 8 | 2.25 | 58 | 59 | 58 | 59 | 59 | 89 | 89 | 89 | 89 | 89 | 19.5 | 19.5 | 19.5 | 19.0 | 19.4 | 153 |
| E-2.2 | 8 | 1.9 | 47 | 48 | 47 | 47 | 47 | 95 | 95 | 95 | 95 | 95 | 32.8 | 32.4 | 28.2 | 28.7 | 30.5 | 158 |
| E-2.3 | 8 | 1.7 | 43 | 43 | 43 | 44 | 43 | 94 | 94 | 95 | 95 | 94 | 29.0 | 31.2 | 28.5 | 28.7 | 29.3 | 160 |
| E-2.4 | 8 | 1.6 | 36 | 35 | 35 | 36 | 36 | 96 | 97 | 96 | 96 | 96 | 29.7 | 28.1 | 27.5 | 26.2 | 27.9 | 162 |
| E-3.1 | 12 | 3.2 | 68 | 67 | 67 | 66 | 67 | 90 | 90 | 90 | 90 | 90 | 26.5 | 26.8 | 25.5 | 26.8 | 26.4 | 149 |
| E-3.2 | 12 | 2.4 | 39 | 38 | 38 | 39 | 39 | 96 | 96 | 96 | 96 | 96 | 39.2 | 36.6 | 33.2 | 36.7 | 36.4 | 160 |
| E-3.3 | 12 | 1.9 | 31 | 30 | 32 | 31 | 31 | 96 | 98 | 97 | 97 | 97 | 29.2 | 29.8 | 28.2 | 31.4 | 29.6 | 164 |
| E-4.1 | 20 | 4.7 | 60 | 58 | 58 | 59 | 59 | 93 | 91 | 91 | 91 | 91 | 28.4 | 28.7 | 28.7 | 28.2 | 28.5 | 150 |
| E-4.2 | 20 | 3.7 | 33 | 32 | 34 | 34 | 33 | 96 | 95 | 95 | 95 | 95 | — | 28.7 | 28.4 | 28.8 | 28.6 | 158 |
| E-4.3 | 20 | 3.4 | 26 | 25 | 26 | 26 | 26 | 97 | 97 | 97 | 96 | 97 | — | 28.3 | 26.8 | 26.2 | 26.2 | 162 |
| E-4.4 | 20 | 4.2 | 37 | 36 | 35 | 36 | 36 | 96 | 96 | 95 | 96 | 96 | — | 30.7 | 29.2 | 30.4 | 30.1 | 160 |

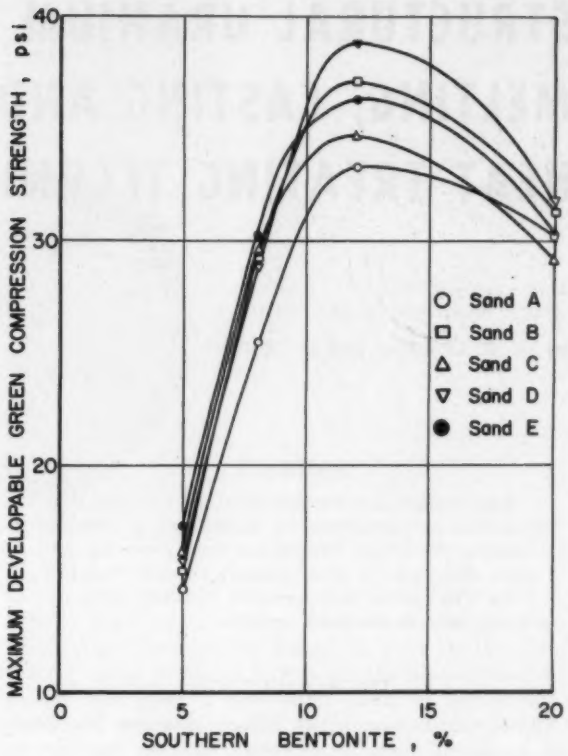



Fig. 3 — Maximum development of green compression strength vs. southern bentonite content for each of the five test distributions.

3) Higher binder levels require larger amounts of water in order to develop maximum permeability in any particular sand distribution.

As a result of this research, it must be concluded that whereas variations in median grain size have a considerable effect on permeability, they have little or no effect on green compression strength. Furthermore, the sand grain surface area and the number of sand grain points of contact have little or no effect upon the strength of the sand-clay bond.

Before the full impact of this information is realized, and before a more meaningful sand classification system can be devised, a better understanding of the factors that affect the sand-clay bond must be developed.

## ACKNOWLEDGMENT

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## APPENDIX

### *Comparison of Theoretical and Experimental Determinations of Moisture Required for Maximum Strength*

**THEORETICAL DETERMINATION**

| U.S. Standard Sieve Number | a Weight Retained (gm) | b Weight Retained (%) | c Moisture Factor[6] (gm water / gm sand) | d Moisture Required (column b x c / %) |
|---|---|---|---|---|
| 40 | 0.4 | 0.7 | 0.00587 | 0.004109 |
| 50 | 0.5 | 0.9 | 0.00834 | 0.007506 |
| 70 | 4.2 | 7.9 | 0.01177 | 0.092983 |
| 100 | 18.9 | 35.4 | 0.01662 | 0.588348 |
| 140 | 21.2 | 39.7 | 0.02380 | 0.944860 |
| 200 | 6.9 | 12.9 | 0.03320 | 0.428280 |
| 270 | 1.0 | 1.9 | 0.47020 | 0.139680 |
| Pan | 0.3 | 0.6 | 0.09100 | 0.054600 |
| Total moisture required for sand (wt.-pct. basis) — | | | | 2.260360 |

[Total moisture required for maximum green compression strength] = [Moisture required for sand (94% of sand clay mixture)] + [Moisture required for 6 per cent southern bentonite]

Total moisture required = (2.26) (0.94) + 6 (0.15) = 3.0%

**EXPERIMENTAL DETERMINATION***

| Moisture, % | 1.1 | 1.4 | 1.8 | 2.2 | 2.5 | 3.0 |
|---|---|---|---|---|---|---|
| Green compression strength, psi | 8.2 | 14.3 | 13.3 | 10.6 | 9.1 | 7.7 |

* All data is an average of three tests.

Experimentally determined moisture requirement for maximum green compression strength = 1.4%

# STRUCTURAL URANIUM ALLOY MELTING, CASTING AND HEAT TREATING TECHNIQUES

by G. D. Chandley and D. G. Fleck


## ABSTRACT

**High strength uranium-base alloys are becoming more important in construction of nuclear and non-nuclear weapons. Watertown Arsenal has been producing uranium alloy parts for Army weapons for more than four years. The various techniques for handling these uranium alloys are described in detail.**


## INTRODUCTION

The most conventional use of uranium has been as a nuclear fuel. It is widely used in reactors to produce heat and various types of radiation. This is because the uranium 235 in natural uranium has an excellent ability to sustain a fission chain reaction. However, the percentage of uranium 235 in natural uranium is only 0.7 per cent, the remainder of the natural uranium being uranium 238. Therefore, in refining the uranium metal in order to obtain a higher percentage of uranium 235, most of the natural uranium becomes a by-product available for uses other than a nuclear fuel. Other general properties of uranium which warrant its use for non-fuel applications are:

1. Uranium has a remarkable ability among metals of reasonable cost to absorb large amounts of gamma radiation. For example, uranium will absorb 12 to 15 times as much gamma radiation as an equivalent thickness of steel. This encourages its use for shielding sources of intense radiation.
2. Structural uranium alloys have another distinct property—they are the heaviest structural metals developed to date. Their density of approximately 0.67 lb/cu in. is more than 50 per cent greater than the density of lead. Therefore, for applications requiring a high density combined with exceptional strength, uranium base alloys are outstanding.
3. The mechanical properties of uranium base alloys equals those normally expected of high strength steel. A 0.1 per cent offset yield strength of 140,000 psi can be obtained, and elongation of 15 per cent (reduction of area of 45 per cent) would be found at this yield strength level. Impact resistance is sensitive to minor impurities, but a charpy V notch impact energy of 5-8 ft-lb at –40 F can be obtained if the stock is vacuum heat treated.

Many classified applications have been developed for uranium alloys based on the above mentioned properties, and wider use of uranium is expected in the future as nuclear power is extended to more applications. Indeed it is quite conceivable that cast uranium alloys could some day play as important a part in space travel as cast aluminum and cast iron have played in ground travel.

## FOUNDRY PRACTICE

### *General Chemical Behavior*

Pure uranium is reactive chemically. It will decompose when exposed to air, forming uranium oxides and hydrides. Atmospheric corrosion is rapid, and highly polished pure uranium will turn completely black overnight if unprotected. The rate of reaction increases with temperature, and at elevated temperatures uranium forms a powdery black oxide which falls off the surface. If uranium is struck sharply with another object, or is sand blasted, a shower of sparks will be produced. As the temperature is raised to the melting temperature of 2090 F (1143 C), one finds that molten uranium will rapidly react in air forming uranium oxide, uranium nitride and uranium hydride.

Therefore, uranium, when molten, must be held under an inert atmosphere or good vacuum. Other important reactions which can occur in uranium processing are due to uranium's ability to form a low melting eutectic with iron and to react with carbon to form uranium carbides. Contact between uranium and these materials above 1340 F (727 C) must be avoided or contamination of the uranium will be found. Many refractory oxides are stable in contact with molten uranium and refactories as stable as zircon, magnesium oxide and mullite can be used for molds and crucibles.

When heated, uranium will dissolve hydrogen readily. A pickup of several parts per million of hydrogen will be encountered in heating uranium

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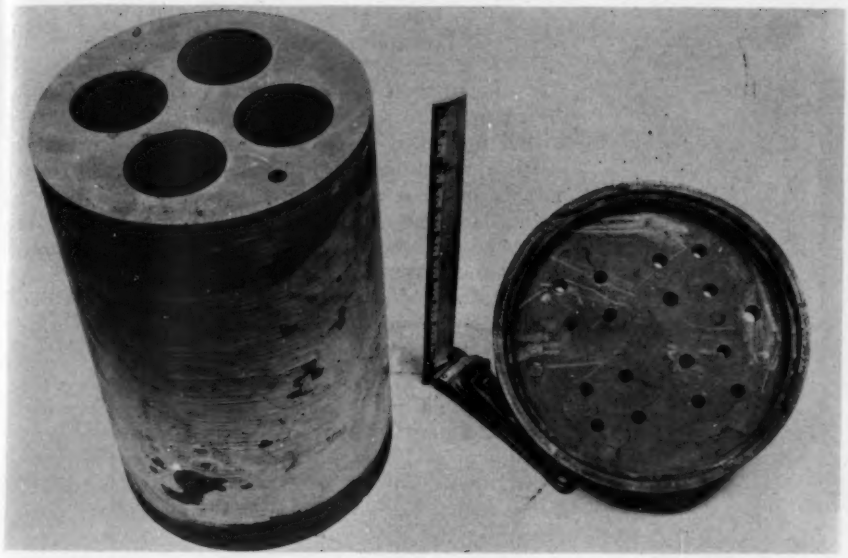


Fig. 1 — Typical graphite billet (left) with pouring basin (right).

in neutral salts conventionally used for scale free heat treating of steel.

*Molding*

As noted, molding materials must be inert to molten uranium. Since there is a number of refractories which are more stable than uranium oxide, the considerations imposed by vacuum melting become of prime importance in selecting a suitable refractory. For example, refractories with high vapor pressures cannot be used. It has been conventional throughout atomic energy facilities to use machined graphite molds coated with various refractories to prevent interaction between the uranium and graphite.

Also, the graphite molding material can be induction heated, in most cases in the same coil which does the melting. This introduces temperature gradients in the mold, and promotes directional solidification to insure a completely sound casting. When graphite molds are used, extreme care must be taken to remove all traces of gas within the graphite. It has been found in practice that graphite molds must be heated to 1300 F (704 C) under vacuum in order to make absolutely certain that no gas porosity will occur within the uranium casting. A typical graphite mold for making bar type castings is shown in Fig. 1.

This type of mold would be set up coaxial with the melting crucible and filled from the top, as shown in Fig. 2. In typical operation, the top of the graphite mold in position is heated to about 2300 F (1260 C), and the bottom of the graphite mold would be in the range of 1300-1400 F (704-760 C). This produces excellent directional solidification and gives a maximum yield of metal cast. Design of graphite molds for simple shape castings is therefore relatively simple. A riser volume of 20 per cent is used, and a shrinkage allowance of 1.8 to 2.2 per cent is made depending on the degree of restraint caused by the graphite mold.

A satisfactory mold wash consists of a thin zirconite base wash, which is then covered with a sec-

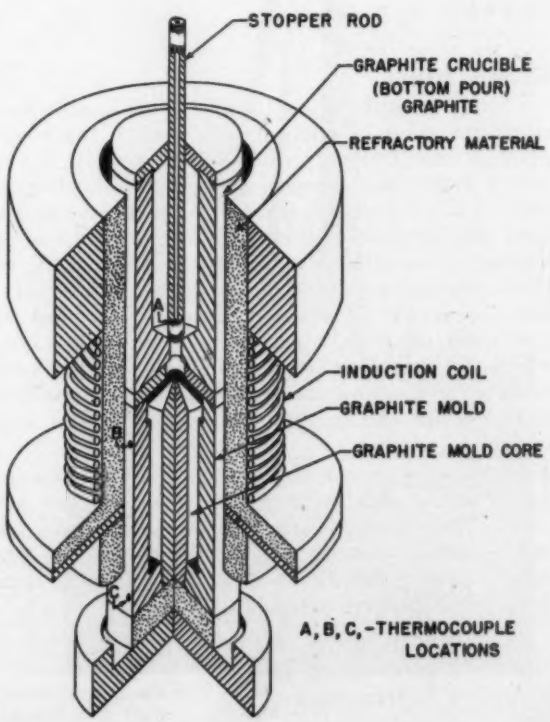



Fig. 2 — Uranium bottom pour induction melting and casting furnace used at Watertown Arsenal.

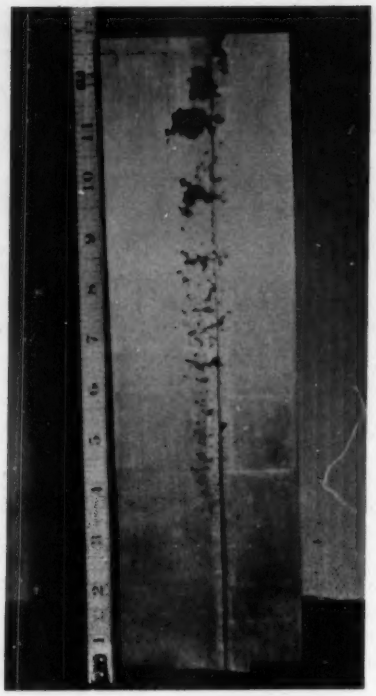


Fig. 3 — 3⅝-in. diameter uranium, 8 per cent molybdenum alloy. This casting was originally 18 in. long. Centerline shrinkage extended almost throughout the casting.

ond coating of fine mullite. At Watertown Arsenal both washes are normally put on by brushing rather than spraying. After being coated, the molds are packed in an oven at 400 F (260 C) for at least 2 hr. This procedure results in fine casting finishes, which will show the brush marks on the mold coating and will result in negligible carbon pickup from the mold.

### *Ceramic Molds*

Sintered ceramic molds have been used, and generally are more economical in use than machined graphite molds. Better dimensional control is obtained with ceramic molds, and preheating prior to pouring is not necessary providing the ceramic molds have been fired at high temperature. No attempt at heating ceramic molds for the establishment of thermal gradients would normally be made; therefore, careful gating and risering techniques must be used. The uranium 8 per cent molybdenum alloy, which is the most important of the structural uranium cast alloys, shows a shrinkage comparable to that which will be obtained in a low carbon steel casting.

An example demonstrating the type of shrinkage which occurs is shown in Fig. 3, which is a saw cut section through a 3⅝-in. diameter billet approximately 18 in. high. It was bottom gated. The sintered zircon mold used was of the following composition for each 100 lb batch:

80 lb zircon sand
20 lb zircon flour
1500 cc water
500 cc phosphoric acid (C.P.)
712 grams aluminum hydroxide

It can be seen from examination of Fig. 3 that the zircon mold caused freezing from the outside diameter in, and little feeding along the length of the billet was obtained. This shrinkage is fairly comparable to what would have been obtained with a low carbon steel casting cast in a zircon mold. Castings made in the zircon mold show a fine surface finish and have no evidence of any chemical reaction. Another mold material used at Watertown Arsenal is commonly a process which has been widely described in the literature.[1]

The refractory material used for uranium is a fine mullite. Other aspects of the molding procedure with this mold material are well described in the referenced article. Castings produced in this manner have exceptionally fine surface finish and excellent dimensional reproducibility. Figure 4 shows some small projectile castings which have been cast to close tolerances in the molding process. These little projectiles, being machined from bar stock, are currently replacing tungsten base alloys at a great reduction in cost. Figure 5 shows another test casting poured in this molding material.

The pattern shrinkage allowance, which should be made in using this process is about one per cent, similar to that for steel. Here again the uranium casting will show fine detail and fine marks left in the mold during the molding process.

### *Melting and Pouring*

As outlined previously, molten uranium will rapidly form carbides, oxides, nitrides and hydrides. This extreme chemical reactivity precludes the melting of uranium in air, in certain oxide crucibles and in carbon or carbide crucibles, unless a protective wash is applied to these materials. The most frequently used crucible material is graphite. The desired crucible is normally machined from large blocks and given a coating of a thin, adherent stable oxide. For convenience, at Watertown Arsenal the same wash is used for molds as for crucibles.

A creamy solution of zirconite in water is thinly spread over the crucible and allowed to dry. Following this, a second coat of a proprietary coating is brushed on and baked out at 400 F (260 C) for 2 hr. Properly applied, this coating will adhere to the crucible during melting, and carbon pickup is insignificant if not undetectable. Graphite crucibles can be used for tilt pouring or bottom pouring. A typical resultant analysis is:

| Heat | Mo | Ti | C | O | H | N | Si | Fe |
|---|---|---|---|---|---|---|---|---|
| Desired, % | 8/8.5 | 1/1.2 | <0.015 | <0.006 | <0.0002 | <0.0002 | <0.02 | <0.060 |
| E-76, % | 8.20 | 1.00 | 0.0115 | 0.0034 | 0.00016 | 0.0013 | 0.006 | 0.0084 |

Figure 2 shows a conventional crucible-mold setup for melting and casting in graphite. Naturally, little induction stirring is obtained when a graphite crucible is used, but the degassing action of the high vacuum will cause adequate stirring for homogeniza-

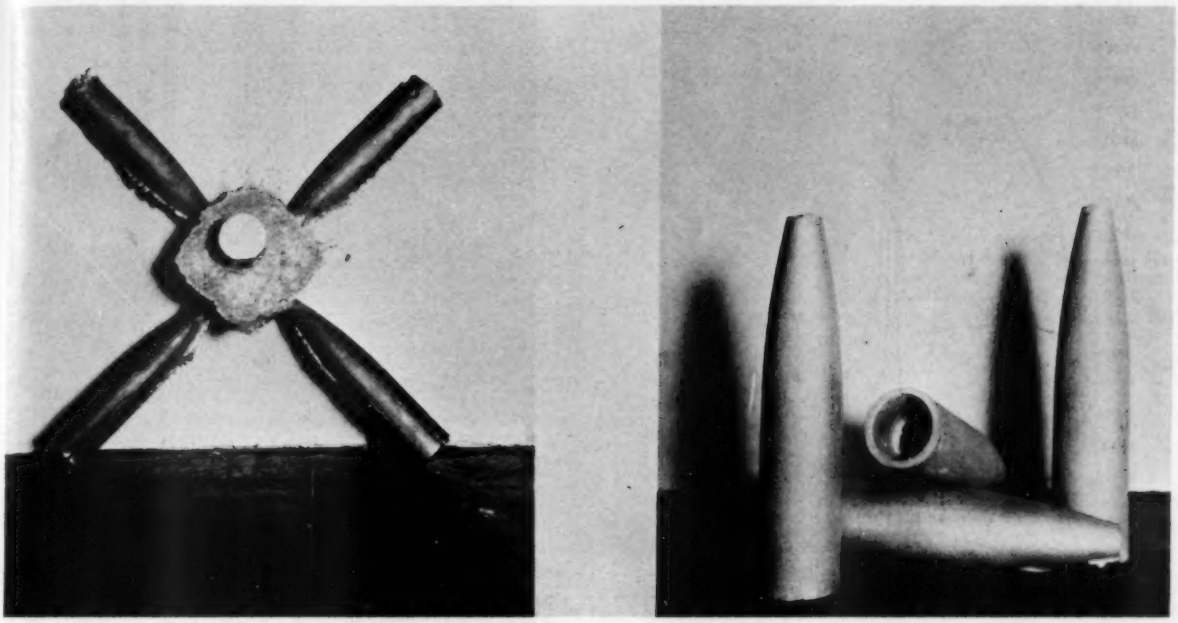


Fig. 4 — Uranium projectile casting made by patented process[1]. Castings attached to sprue-riser are at left, and cleaned projectiles are at right.

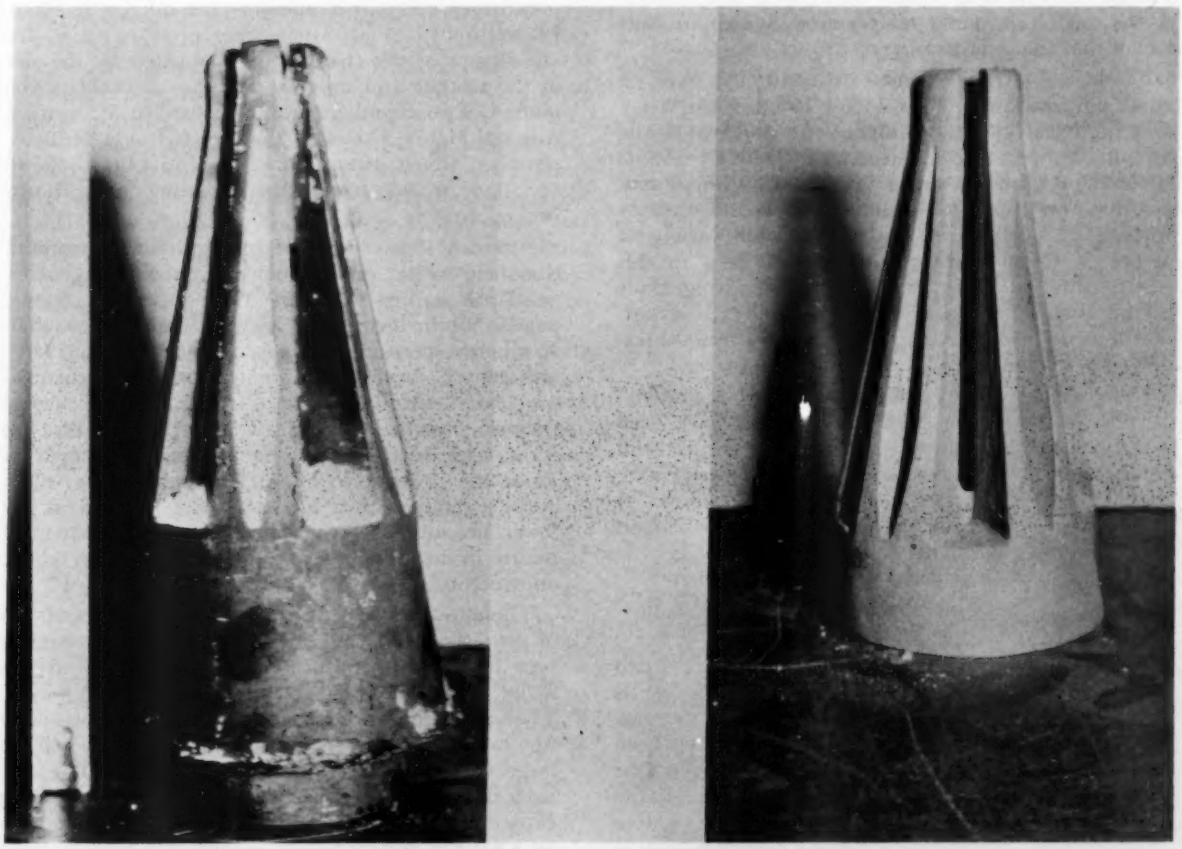


Fig. 5 — As-cast uranium castings shown at left, and after cleaning and sand blasting at right.

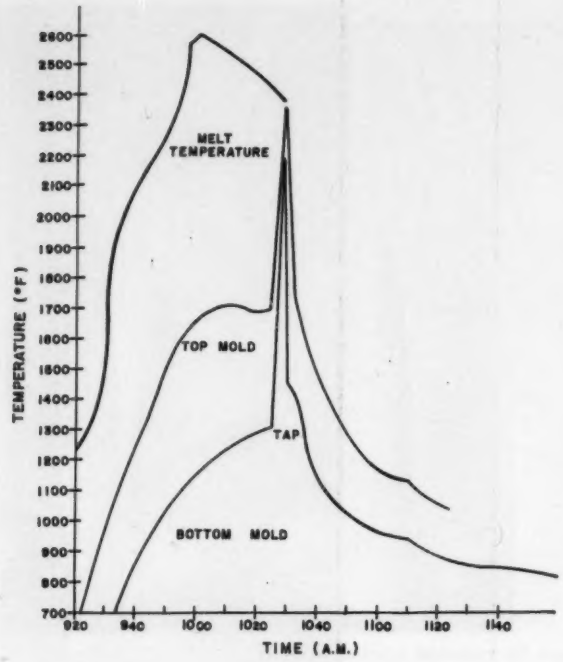



Fig. 6 — Thermal history of important portions of conventional melting and casting used shown in Fig. 2.

tion if the alloys are held for about 5 min at 2550 F (1399 C) or above. It should be noted in Fig. 2, that three temperatures are recorded throughout the melting and casting cycle. The melt temperature *A*, the top of the mold temperature *B* and the bottom of the mold temperature *C*.

To make perfectly sound castings in this type of unit, temperature of *C* is 1300 F-1400 F (704-760 C), *B* is high enough so the top of the mold is at the liquidus temperature about 2350 F (1288 C) and *C* is 100-200 F (38-94 C) above the liquidus. These temperature conditions have been found satisfactory for sections of ¾ to 8 in. and for a wide variety of shapes. A thermal history of a typical heat in this casting unit is shown in Fig. 6. The casting made in this instance was a bushing about 6 in. in outside diameter 14 in. long, with a ¾-in. wall thickness.

It should be noted that the power input and relative positions of the crucible and mold have been controlled so that the charge has been melted and homogenized by the time the proper gradients have been established in the mold. Also noteworthy is the rapid solidification of the casting even though the mold was thoroughly preheated.

### *MgO Crucibles*

Sintered MgO crucibles have been used for small heats, but for larger heats rammed MgO lining is satisfactory for melting heats to be tilt poured. Tilt poured heats at Watertown Arsenal have been poured from both graphite and rammed MgO. One disadvantage of tilt pouring is that highly radioactive particles are dispersed throughout the vacuum chamber, and stringent health physics precautions must be observed. Use of a bottom pour setup, as shown in Fig. 2, will give the minimum amount of radio activity in the vacuum chamber.

Induction heating of the crucible and charge (and often mold) for melting in vacuum is the universal practice. The equipment is similar to air induction melting equipment, except that the field voltage is limited to 400 volts as higher voltages cause "corona" type of electrical discharge in vacuum as well as arcing. The power supply used at Watertown Arsenal is a 600 kw motor-generator set of 960 cycle frequency. The power requirements for melting uranium are not great due to its low specific heat and the low melting point.

The furnaces, crucibles and molds are housed in a chamber which can normally be evacuated to a pressure of about one micron. One micron is about 1.3 millionths of an atmosphere pressure. The size and design of the chamber is predicated by the size of the furnace and quantity and size of molds to be poured. Two chambers are being used at Watertown Arsenal. Figure 7 shows a small, 4 ft inside diameter chamber, which is used to melt and cast up to 300 lb of alloy in the conventional casting unit shown in Fig. 2.

Figure 8 shows the 8 ft inside diameter vacuum chamber, which can be used either for large conventional casting or for the tilt pouring of ceramic molds. While melting is always begun at less than 10 microns, pressure increases up to 200 microns have not caused any significant change in mechanical properties of soundness of the castings. Two stage vacuum pumping is used for both chambers. That is, a large mechanical roughing pump is used to lower the pressure to about 1000 microns at which point a pump with a high pumping capacity at the lower pressures is activated to assist the mechanical pump in lowering the chamber pressure to about one micron.

These pumps are called booster pumps and are of two types—oil and mechanical. The mechanical type has a more uniform pumping capacity over a wider range of pressures than does the oil type. Therefore, the mechanical pump is recommended for melting chambers which may be exposed to large uncertain gas loads under vacuum. The furnace shown in Fig. 8 has adequate vacuum pumping capacity consisting of a 780 cfm roughing pump with an 8000 cfm blower-booster pump. The total volume of the chamber is about 450 cu ft. The stainless

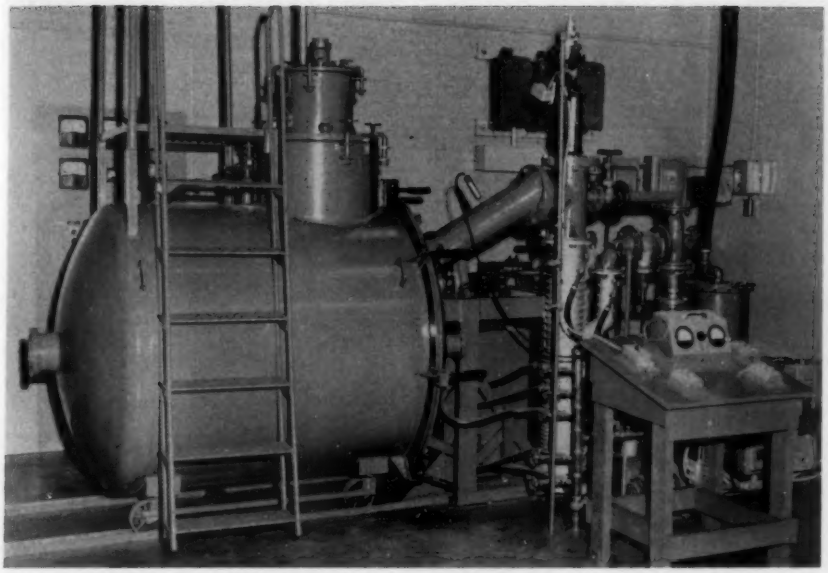


Fig. 7 — Vacuum chamber and pumping system used in conventional graphite crucible mold arrangement shown in Fig. 2.

steel shell is water cooled as is conventional in vacuum melting furnaces.

*Inspection*

In addition to normal dimensional inspections, the quality of uranium alloy castings may be controlled by most nondestructive test methods, except those which are designed for ferromagnetic materials. Dye check inspection of all uranium parts made at Watertown Arsenal is routine. In addition, full use of radiography is made, although long exposure times at high voltages are required.

Figure 9 is an x-ray of the riser and upper section of a bushing casting made in a graphite mold. It shows a sound casting-riser interface, with the shrinkage appearing as a white cloudy area confined to the upper part of the riser. Figure 10 shows two radiographs at different densities of the casting shown in Fig. 5. The riser section has been removed and the casting is of excellent soundness.

Careful chemical analysis is essential to adequate quality control of uranium castings. In particular, the mechanical properties and ductile-brittle transition temperature is sensitive to hydrogen content. The hydrogen content is normally less than two ppm in the as-cast part. However, heating in saltpots and air can raise this as high as 6.3 ppm. This is extremely important, because pickup of as little as 2.5 ppm of hydrogen can reduce the room temperature ductility to a fraction of its original value.

Another valuable aid in establishing a casting practice is macro-etching. This will show general grain structure as well as areas of microporosity which will not be detected by x-ray. A macro-etch of a ½ per cent Mo, ½ per cent Cb, uranium alloy taken from a vertical section of a 6 in. OD, ¾-in. wall thickness bushing is shown in Fig. 11. This grain structure does not give adequate ductility. Figure 12 shows a horizontal section of a 2 per cent Mo bushing casting which has a fine grain size. Macro-etches of 8 per cent Mo alloys show a finer grain structure

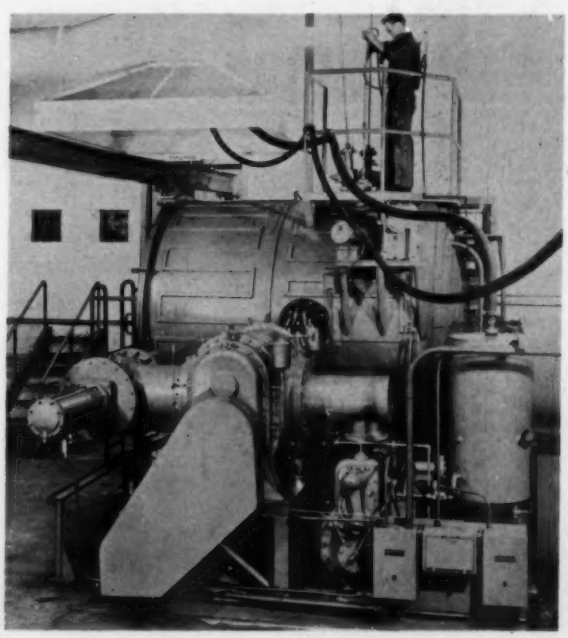


Fig. 8 — Vacuum system for 2200 lb uranium vacuum melting furnace which can be used for conventional bottom pouring or tilt pouring ladles. Chamber is of 8 ft inside diameter and 9 ft long. Melting power is 300 kw.



Fig. 9 — Radiograph of riser and upper part of casting described in text. Shrinkage is confined to the riser section at the top of photo. Casting is sound.

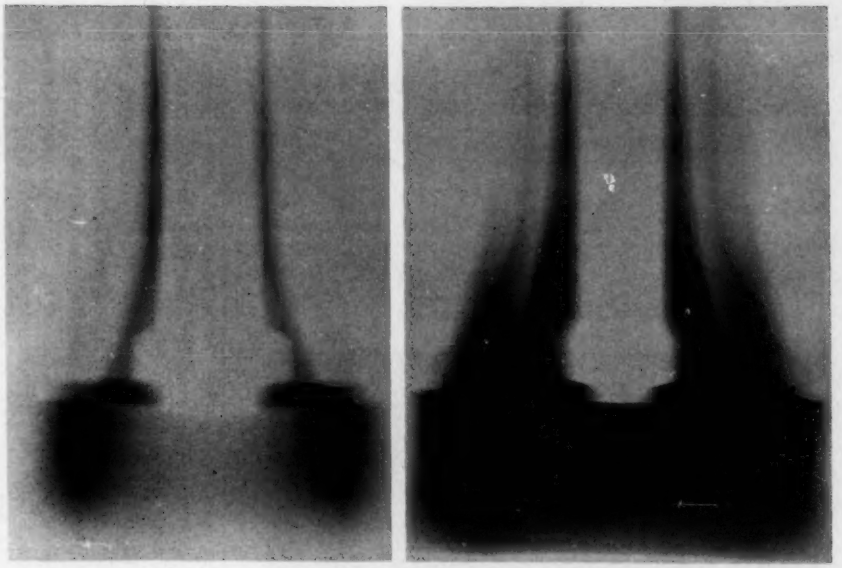


Fig. 10 — Radiographs of rather complicated uranium casting shown in Fig. 5. Films are of different density in order to obtain proper accuracy in the thin and thicker sections of casting.

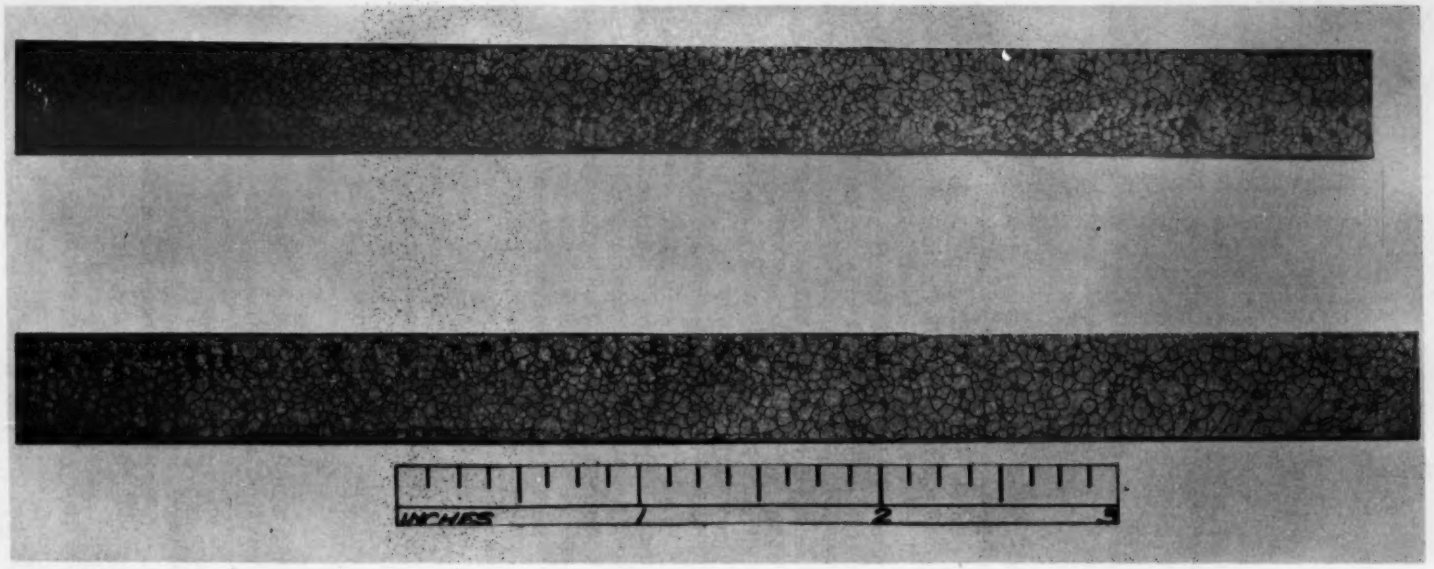



Fig. 11 — Macro-etch of a uranium ½ per cent molybdenum, ½ per cent columbium alloy. Section at top is bottom portion of casting, and section at bottom is upper portion of casting. Note that grain size is large and nonuniform.

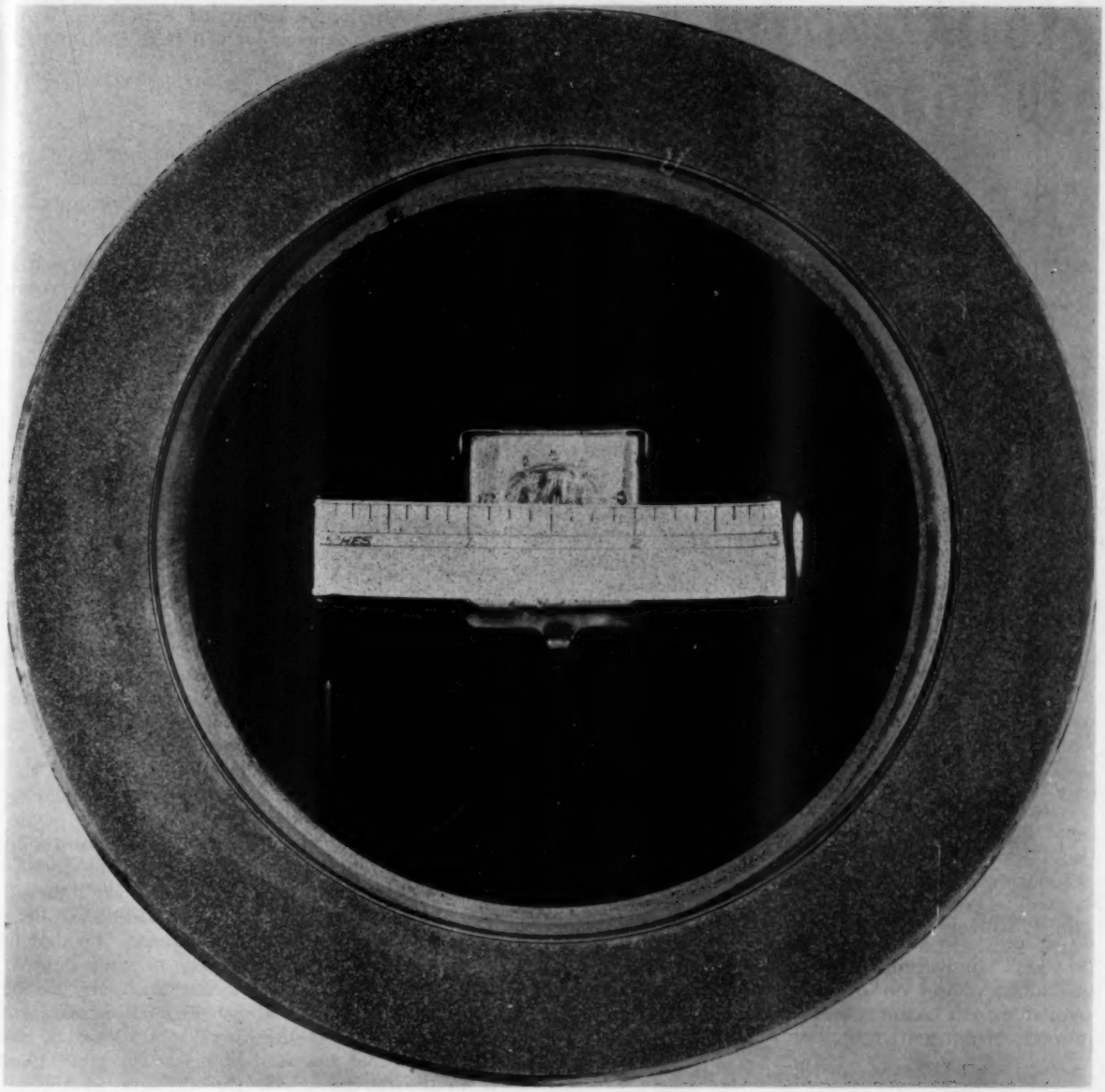


Fig. 12 — Macro-etch of horizontal slice through a vertically cast bushing in uranium, 2 per cent molybdenum alloy. The grain size is fine and will yield good ductility for this analysis. Reduced slightly in reproduction.

than that of Fig. 12. This type of structure will give the optimum properties outlined below for heat treated castings.

## PHYSICAL METALLURGY

The physical metallurgy of uranium alloys is wonderously complicated, and there are many excellent references on the subject.[2,3] At present, the alloys of prime interest are those which have high strength and corrosion resistance. The alloys which satisfy this requirement are metastable gamma phase alloys. Additions of molybdenum and columbium will make gamma uranium (body centered cubic structure) stable at normal temperatures much in the same manner that chromium will make austenite in steel stable at normal temperatures.

Thus, while the equilibrium phase diagram Fig. 13 shows alpha and delta phases at room temperatures, one finds that, in the range of about 7 to 12 per cent Mo, the high temperature gamma phase is stable. This one phase alloy has excellent corrosion resistance to mildly corrosive media—far better than mild steel. Unprotected, machined castings will stay bright and shiny for periods in excess of one year. However, if these metastable gamma alloys are slowly cooled during casting they will contain some alpha and delta phases, and must be reheat treated in order to make them ductile and fully corrosion resistant.

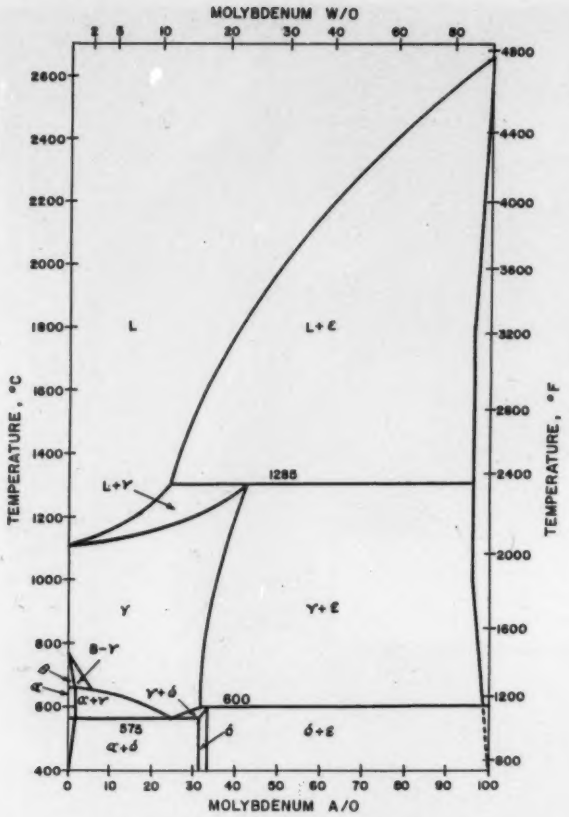



**Fig. 13 — Uranium-molybdenum equilibrium phase diagram. Under nonequilibrium cooling the gamma phase is stable at room and slightly elevated temperatures from 7 to 12 weight per cent molybdenum.**

In normal production at Watertown Arsenal the 8 per cent Mo castings are heated to 1600-1650 F (871-899 C) in vacuum for several hours, depending on thickness and oil quenched. The castings heat treated in this manner will yield these mechanical properties consistently:

| | |
|---|---|
| Y.S., 0.1 per cent offset, psi | 130-135,000 |
| T.S., psi | 130-140,000 |
| Elongation, per cent | 15-17 |
| Reduction of Area, per cent | 40-50 |
| Charpy impact at −40 F, ft-lb | 3-8 |

A short stress relief treatment at 950 F (510 C) can be given the castings without lowering ductility. However, holding for more than a few hours will cause the tensile strength to drop to about 100,000 psi, and cause the ductility to drop to zero.

For the future, the possibilities in uranium alloys are manifold—there are alloys which form martensitic phases,[4] there are alloys which can be age hardened and modifications of the metastable gamma alloys will be of higher strength than now available. In alloy development programs at Watertown Arsenal tensile strengths in excess of 200,000 psi have been obtained. Alloys of this strength do not have as good ductility or corrosion resistance as is desirable, so it can be concluded that the physical metallurgy of uranium alloys will be a fruitful field for study in the years to come.

## SUMMARY

The processing techniques for making structural uranium alloy castings at Watertown Arsenal have been described. Important aspects of the practice are:

1. Uranium must be melted in vacuum or inert gas. Vacuum melting is preferred to insure proper quality.
2. Mold materials must be free of minute traces of absorbed gases, i.e., graphite must be heated to 1300 F (704 C) under high vacuum in order to eliminate gas porosity.
3. Uranium can form a low melting eutectic if it contacts bare iron or steel at temperatures in excess of 1340 F (727 C). It dissolves carbon and reacts with some common refractories so that short melting cycles and pouring at the lowest temperatures should be used to reduce contamination.
4. Careful chemical analysis of heats must be made in order that good ductility can be assured. It has been shown that hydrogen in as little amounts as 2.5 ppm can reduce ductility to a small fraction of its original value.
5. Vacuum heat treating is required to assure good ductility, because uranium can dissolve hydrogen from air and molten "neutral" salts at the normal heat treating temperatures encountered and the castings will be drastically embrittled thereby.
6. Adequate health physics precautions such as the use of film badges, monitoring radiation and good housekeeping are required for foundry operations regardless of whether or not the uranium is natural, enriched or depleted. This is required because the radioactively hot isotopes and daughter products segregate during melting, and if airborne can cause a level of radioactivity in the vacuum chamber far above the maximum allowable exposure. Castings made of alloys of depleted (D-238) uranium will show a small level of radioactivity, which can be ignored for all practical purposes.

For current engineering applications requiring a maximum density or great gamma ray absorption, combined with a high strength and good corrosion resistance uranium base alloy castings are not only economically competitive with other materials, but are indeed unique in their ability to satisfy these requirements.

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# LOW MELTING ALLOYS FOR PATTERN SHOP USE

by O. J. Seeds


## ABSTRACT

**Use of bismuth (low temperature melting) alloys in making of duplicate patterns for matchplate construction is discussed. Factors of economy in processing and speed in construction are important considerations in pattern shop uses of these materials. In addition to pattern fabrication methods are described for use of low melting metals applied with a spray gun — to coat wood patterns to increase wear and prevent warpage, to construct metal shell patterns for core boxes and dryers and to make metal faced plaster reinforced molds for plastic patterns. Bismuth alloys provide a quick, easy method for fastening patterns in matchplates, producing patterns compensated for shrinkage, altering or repairing both patterns and matchplates and for salvaging worn core boxes and for checking core prints and core boxes.**


## INTRODUCTION

Metal master working patterns are frequently used in match plate production for various reasons, such as greater durability and ease of removal from wet molds.

Wet molds (plaster) offer considerable resistance to removal of wooden patterns, and those having weak thin sections requiring draw screws for withdrawal may be broken. Metal patterns are most readily withdrawn from plaster molds.

A variety of metals is available for making master patterns and they are found roughly in two classes — high temperature melting metals and low temperature melting alloys.

High temperature melting metals, including aluminum, antimonial-lead, copper, zinc, tin, white metal, etc., have several major disadvantages in production of duplicate patterns:

1. They shrink from $\frac{3}{64}$-in. – $\frac{9}{64}$-in./ft, and this factor requires time for computation of allowances to offset this shrinkage.
2. High melting temperatures require special melting and pouring equipment.
3. They cannot be poured into wood molds without damaging the mold, and they cannot be poured into plaster molds which are not fully dehydrated without danger of steam blowing.
4. Shrinking pattern metals pull away from molds on solidification, and thus fine detail is poorly reproduced resulting in need of chasing and cleaning, which is extra work.

Low melting alloys for pattern fabrication and those most frequently used are listed in the table. The outstanding characteristics which commend the bismuth alloys (bismuth is the principal common component) for pattern shop uses are:

1. Nonshrinkage.
2. Low pouring temperature – some can be poured below boiling temperature of water.
3. Expansion or growth – some of the bismuth alloys grow over a period of several hundred hours as much as $\frac{1}{16}$-in./ft, and this factor can be put to good use.

Compositions are given in the table to enable anyone to make them up as desired. However, it is generally more economical to use the proprietary commercially available alloys as they carry a guarantee of composition and can be relied upon for uniformity of performance.

**LOW MELTING ALLOYS USED IN PATTERNMAKING**

| Alloy No. | Melting Temp., F (C) | Melting Range, F (C) | Bismuth, % | Lead, % | Tin, % | Cadmium, % |
|---|---|---|---|---|---|---|
| 1* | 158 (70) | – | 50.0 | 26.7 | 13.3 | 10.0 |
| 2 | – | 158-190 (70-88) | 42.5 | 37.7 | 11.3 | 8.5 |
| 3* | 255 (124) | – | 55.5 | 44.5 | – | – |
| 4 | – | 281-338 (138-170) | 40.0 | – | 60.0 | – |
| 5* | 281 (138) | – | 58.0 | – | 42.0 | – |

*Eutectic compositions.

The reasons low melting alloys are ideal for pattern metals are:

1. Nonshrinking. They do not shrink as other types of pattern metals, but actually expand when freezing pushing into fine mold detail assuring perfect reproductions.
2. These metals cast well in sand, plaster, wood or other materials, and much time is saved because less hand chasing and clean up are required.

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3. Thin sections are easily cast because of high fluidity of the metals when fully liquid, especially alloys 1, 3 and 5 in the table. Two and 4 are also fully fluid above their liquidus points, but it will be noted these points are much higher than those of the eutectic composition.
4. Low melting and pouring temperatures (about one half of other pattern metals) result in faster melting with fuel savings.
5. Simple melting equipment will serve. An ordinary iron pan or ladle over a gas flame or electric hot plate is satisfactory for melting.
6. Gates can be cast integrally with patterns, and since they are easily soldered to brass runner time and money are saved in master pattern work (Fig. 1).
7. Low melting point metal patterns are ready for use immediately after solidification, requiring no curing time.
8. Full salvageability. After low melt alloy patterns have served their purpose at the foundry they are returned to melting pot for reuse indefinitely (Fig. 2).

## FABRICATING METAL PATTERNS

A suggested technique for fabricating metal patterns with alloy 3 in sand molds is:

1. Dust wood pattern with lycopodium for parting.
2. Face pattern with about ¼-in. Windsor Locks sand, after which fill flask with a fine Albany sand.

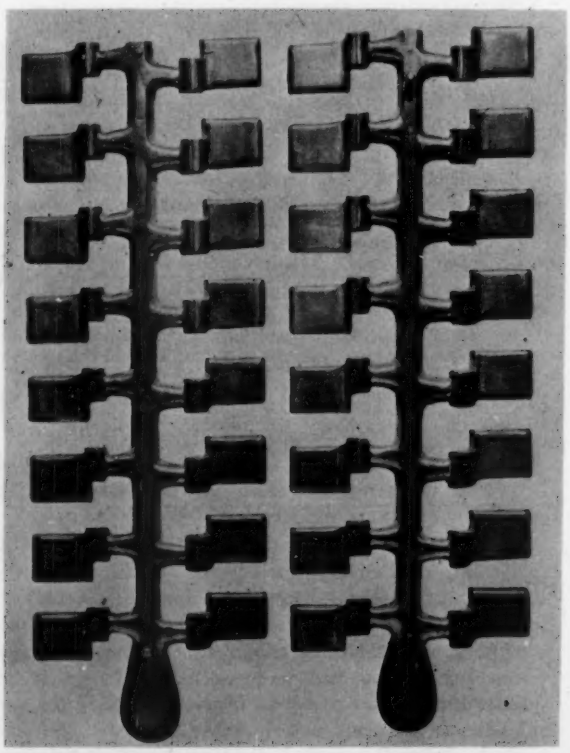


Fig. 1 — Fusible alloy patterns with integrally cast runner and gates.

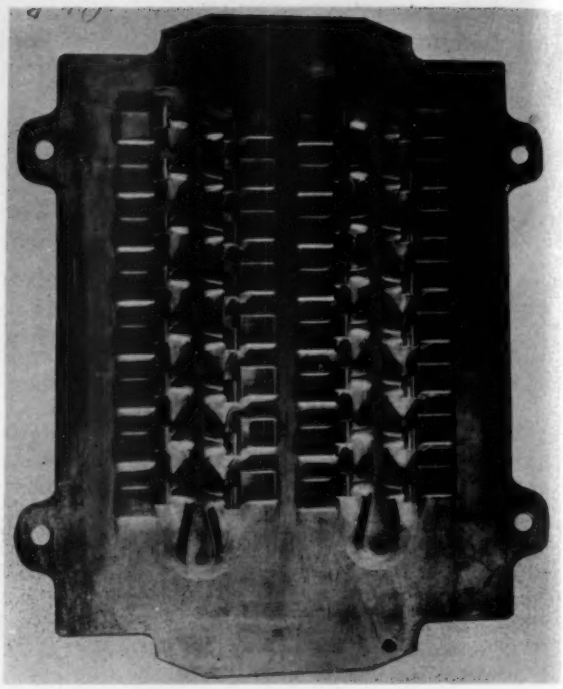
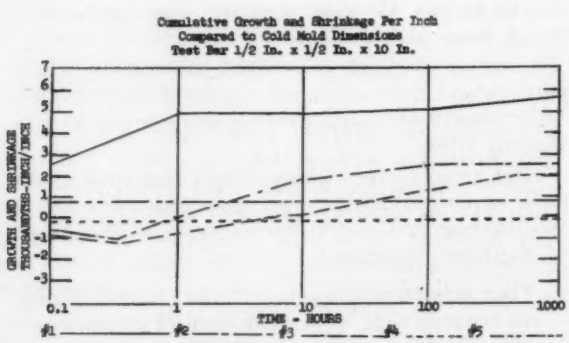


Fig. 2 — Aluminum matchplate molded patterns, alloy patterns returned to pattern shop for remelting.

3. Heavy solid molded castings should be poured with heavy riser on opposite side from gate or runner leading into casting.
4. Some experimenting with pouring temperature may be necessary, and a thermometer should always be used to check metal temperature when pour is made. Long, thin castings require about 310 F (154 C) pouring temperature. Heavy, thick solid castings require about 270 F (132 C) for best results.




For plaster molds use a mixture of one part dental plaster and one part powdered silica. Using alloys 3, 4 or 5 will require the plaster mold to be thoroughly oven dried (1 and 2 can be used in partially dried plaster molds). Oven dry molds should be dusted with thin layer of table salt, sodium chloride or calcium chloride to any moisture. Remove salt before casting.

When soldering gate to runner, tin runner with ordinary tin-lead (50/50) solder. With reduced heat in soldering iron join gates to runner with a solder of alloy 3 one part and 50-50 solder one part. Zinc, ammonium chloride flux is satisfactory. Alloy 3 is the choice of a number of patternmakers, because it expands enough to offset metal removed by steel wool clean up after sand casting.

Higher dimensional accuracy of patterns is obtained by use of alloys 4 and 5 when casting in plaster molds. Better surface finishes are produced when casting in plaster molds if the mold walls are kept as thin as possible to accelerate cooling. The quicker solidification occurs, the better the surface finish will be.

Wood molds are sometimes convenient or desirable to use for casting patterns. With low melting alloys, such as alloy 3, no special treatment of a wood mold is necessary, except to remove any finish which may have been applied. There is no danger of discoloration or damage to the wood. Low melt alloy castings are as smooth as the wood itself, and any wood grain visible on surface of castings is easily removed with fine steel wool or a light polish on a cotton buffing wheel.

### *Core Box and Dryer Patterns*

A rapid, accurate method for making core box and dryer patterns utilizes alloy 2, and results in great savings in manufacturing cost as well as superior castings.

The method was worked out as a solution to the problem of producing master shell patterns of wood having uniform wall thickness. Wood shell patterns of $\frac{3}{8}$-in. wall thickness, with involved curved sections are practically impossible to make.

The method uses a metal sprayer or gun designed for use with bismuth alloys. The procedure is:

1. A core plug is made of wood and mounted on a plywood or masonite board (Fig. 3).
2. Graphite or powdered mica is brushed and polished on the pattern (plug) and board. This is a separator.
3. Alloy 2 is sprayed on all surfaces of pattern and top of board to $\frac{3}{16}$-in. thickness (Fig. 4).
4. Ribs are added to give reinforcement, bosses and dowels are added for blow holes (Fig. 5).

Larger core boxes are made by spraying a light shell of low melting alloy approximately $\frac{1}{64}$-in. thick. A layer of cheese cloth cut to fit the surface of the sprayed core plug is bonded to the surface with shellac. Added thickness is obtained by applying layers of burlap which have been dipped in creamy plaster of paris, and carefully smoothing over the surface. When plaster is dry, add reinforcement ribs and feet to the shell pattern.

Such patterns are usually strong enough to be molded. However, it is recommended to use the plug as a ram up block to support them while ramming.

Because the low melting alloy surface is in direct contact with the core plug there is a direct transfer, and no loss of dimension occurs. Smooth surfaces, curvatures, sharp corners and angles are all reproduced with highest accuracy with minimum of time and

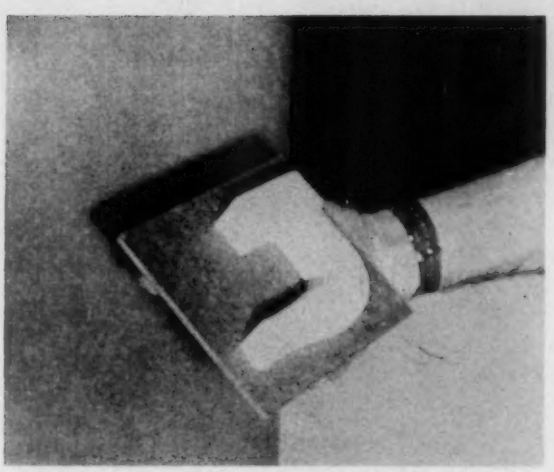


Fig. 3 — Core plug mounted for spraying with alloy 2.

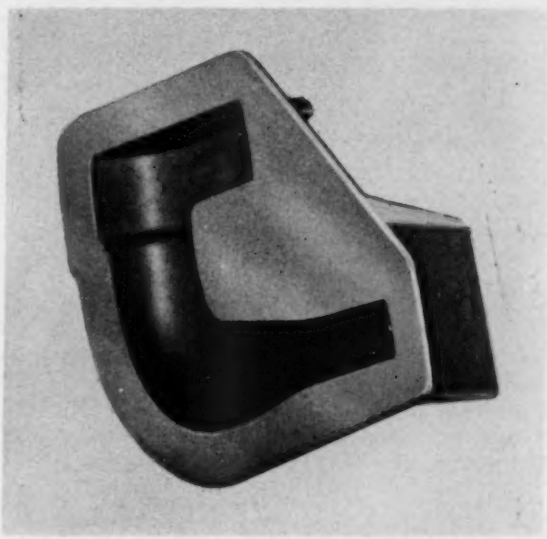


Fig. 4 — Sprayed shell pattern accurately reproduces contour, dimensions and finish of plug.

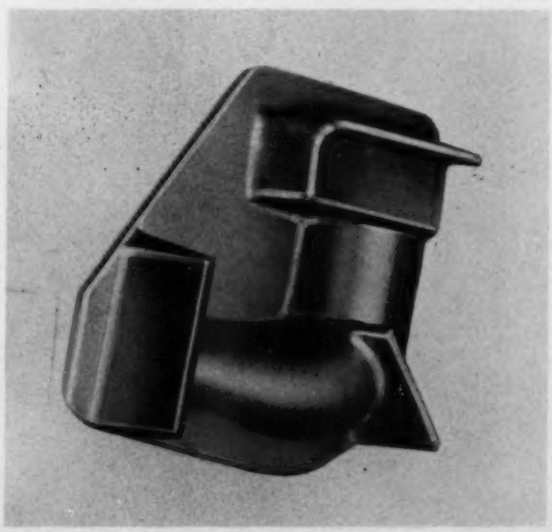


Fig. 5 — Showing bottom of pattern with supporting feet of wood glued on after spraying.

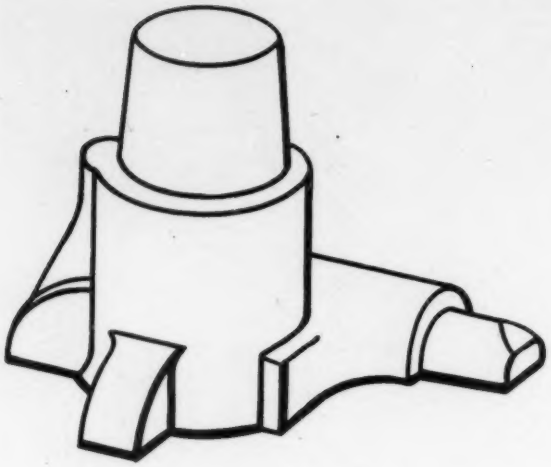
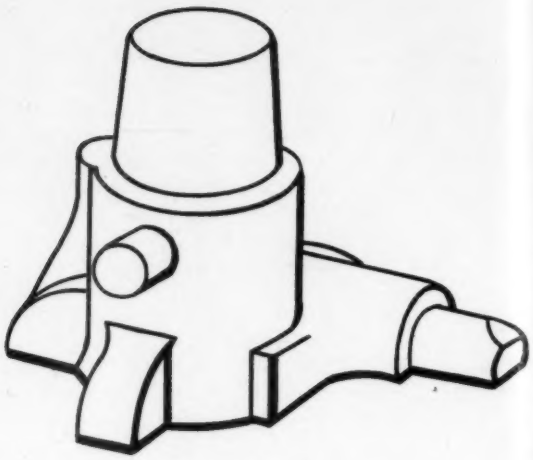


Fig. 7 — Boss added (held with brad and wax).

Fig. 6 — Solid pattern to be altered.



equipment. Uniform thickness shell patterns can be constructed in this manner at a fraction of time and cost required to produce them in wood.

### *Shrinkage Added to an Exact Size Pattern Produces Patterns Compensated for Shrinkage*

Occasionally it is necessary to reproduce a metal casting for which no pattern or blueprint exists. To make a pattern or several duplicates, enlarged to offset the shrinkage of the casting metal, use the following procedure:

1. Place casting on mold board, flask and ram sand to make a mold. Plaster can be used if desired.
2. The first known user of this idea chose alloy 2, melted it in a double boiler and cast it in the sand mold. After aging 4 hr it was noted the casting was 0.004 in./in. larger than original casting. While the growth chart shows only 0.0018 in./in. growth in test bar ($\frac{1}{2}$ in. x $\frac{1}{2}$ in. x 10 in.) in 4 hr, the larger amount can be accounted for by rapping to withdraw casting. Also, other shapes and heavier masses may give different results, so the chart and tabular data should be used only as a guide.

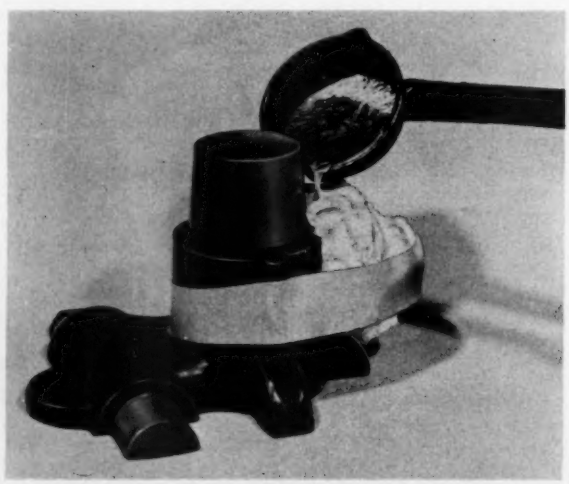


Fig. 8 — Pouring fusible alloy 5 to fill loose piece impression and section of pattern.

The alloy casting aged 4 hr is placed on the mold board and a new sand or plaster mold is made, and casting and aging process repeated.

Again, a third mold and casting are made and the casting is aged 4 hr.

From this third enlargement a plaster mold is made, and the required number of castings taken from it. Final castings aged 4 hr will be found to have increased by $\frac{3}{16}$-in./ft, the amount needed to equal shrinkage of casting metal.

### *Altering Solid Pattern With Loose Piece*

When altering a straight draw solid pattern by addition of a boss, for example, complete rebuilding to a split pattern can be avoided by making a loose piece in a low melting alloy (Fig. 6).

First make the wood boss and brad it to pattern (Fig. 7). Fillet with wax. Take a plaster impression of the boss and adjacent section of pattern (Fig. 8). When plaster is set, remove and cut out dovetail section as shown in Fig. 9. Fasten plaster impression over the chiselled out section and pour alloy 5 to fill the cavity.

When alloy has solidified, remove plaster and trim the loose piece to give proper clearance (Fig. 10). The alloy expands slightly when it freezes, hence it will need relieving to facilitate removal when drawing from mold. Drill blind hole in back of loose piece so

wire hook can be used to withdraw loose piece from mold.

### *Metallizing Wood Patterns and Core Boxes*

Warpage and loosening of glued joints and fillets can be minimized if not prevented completely by coating wood patterns with low melting metal.

It is easy to apply low melting metal 2 with a sprayer designed especially for service with these metals.

Wood patterns are given two coats of shellac, and when the second coat is tacky spraying begins. Any desired thickness of metal may be applied, but about 0.003 in. is considered sufficient. Advantages of this procedure, in addition to minimizing warpage, are — increases life of pattern to almost that of metal patterns, permits easy repair of nicks and dents, alterations can be done readily, metal can be cut by ordinary wood working tools and respraying to cover new surfaces joins smoothly to old coating.

### *Matchplates Repaired or Altered*

Changes, alterations and repairs of matchplates can be quickly and inexpensively made with low melt alloys (Fig. 11).

To change shape or size of one or a series of patterns on a matchplate, build up and carve to exact shape and size of one pattern with wax or clay (Fig. 12). Take a plaster impression of this built up pattern (Fig. 13). Remove plaster when set and mark off margin of built up area with scriber. Rout the marked off area at least ⅛-in. deep (Fig. 14). Drill hole through this depression to back of plate. Replace plaster over the routed depression and pour alloy 5 into the hole (Fig. 15). Alloy will flow against plaster, picking up detail of the alteration and will lock into the depression, making a perfect in-lay on the plate (Fig. 16). Much time can be saved in repairing damage plastic patterns by use of low melting alloy and this technique.

Assume that only one of several identical patterns is to be repaired. Since all patterns are alike, take plaster impression of corresponding area on adjacent pattern. While waiting for plaster to set, have low melting alloy in process. When plaster has set, remove from



Fig. 9 — Dovetail section under boss location chiselled out. Plaster impression seen at right.

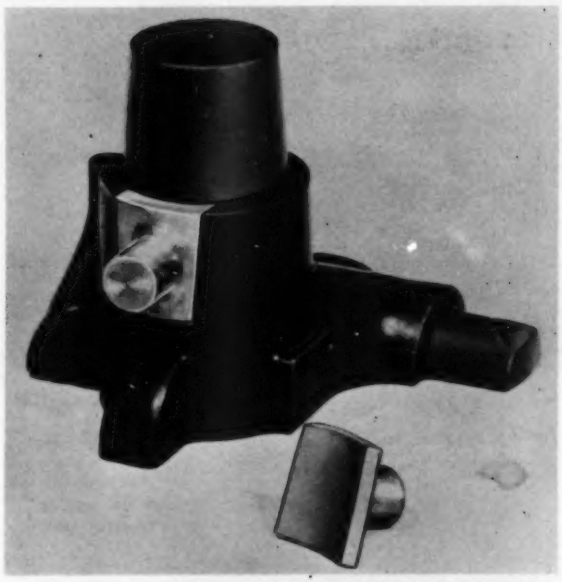


Fig. 10 — Finished loose piece in position on pattern.

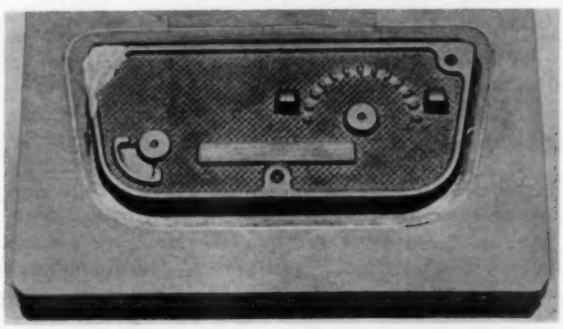


Fig. 11 — Broken corner of matchplate pattern to be repaired.

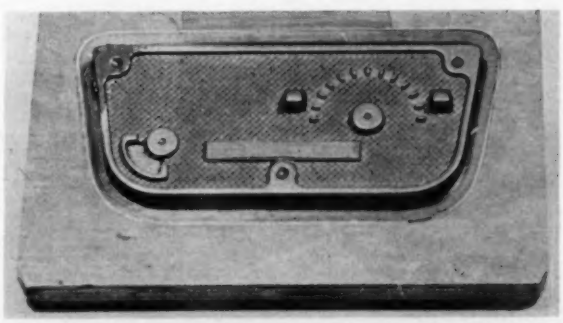


Fig. 12 — Corner rebuilt with wax.



Fig. 13 — Plaster impression being taken of rebuilt section.

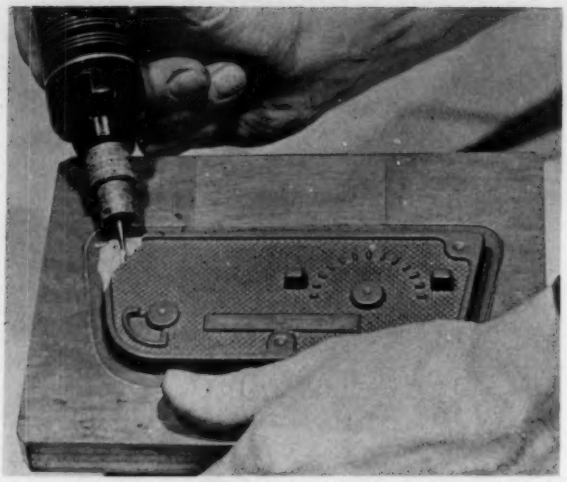


Fig. 14 — Broken section being undercut to anchor fusible alloy. Pouring hole drilled through to other side.

plate and rout area under impression. Eliminate feather edges and drill hole through routed area to other side of plate. Replace plaster and pour molten alloy 5. Plastics require careful measuring and mixing of ingredients plus several hours curing time after pouring.

Also, plaster must be thoroughly dry to assure proper cure. Plastic has limited shelf and pot life, and any excess must be discarded. Low melting metals have indefinite shelf and pot life, require no curing and thus save both time and money by giving tools ready to use immediately after solidification.



Fig. 15 — Pouring alloy to fill cutout section and plaster (impression replaced on pattern section). Pouring sprue held with clay.

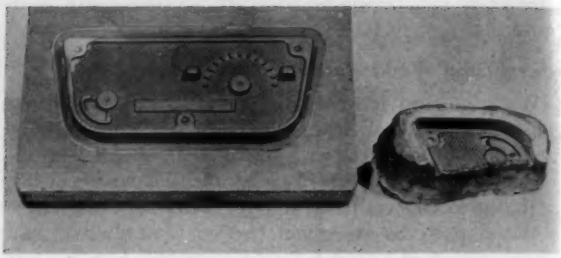


Fig. 16 — Plaster impression removed showing perfect inlaid repair, pattern ready for immediate use.

### *Aluminum Matchplates Raised Letters*

By employing same technique, omission of lettering on master pattern can be easily corrected on completed matchplate.

Glue or shellac desired letters on one pattern and take a plaster impression. When plaster has set, remove from plate. Scribe area around lettering, remove letters and rout the area at least 1/8-in. deep with under cut sides. Drill hole to other side of plate through routed area. Repeat routing of corresponding area for remainder of patterns. Replace plaster over area and pour full with alloy 5. Repeat for balance of patterns. This is possibly the quickest method known for this kind of alteration.

### *Wood Ram-Up Core Boxes Metal Linings*

Ram-up cores are useful in molding patterns which have pockets or sections otherwise difficult to draw (Fig. 17). Making core boxes to fit this type of core is sometimes costly, because the core must not be too tight nor too loose. Adjustments are time consuming. A rapid and accurate method of core box making is:

Make well or dam of clay around pattern section to be cored. Taper sides of dam to form section of core, which will be held in green sand mold. Fill pattern section and well with plaster of paris. When set, remove plaster and trim down to loose fit in the pat-

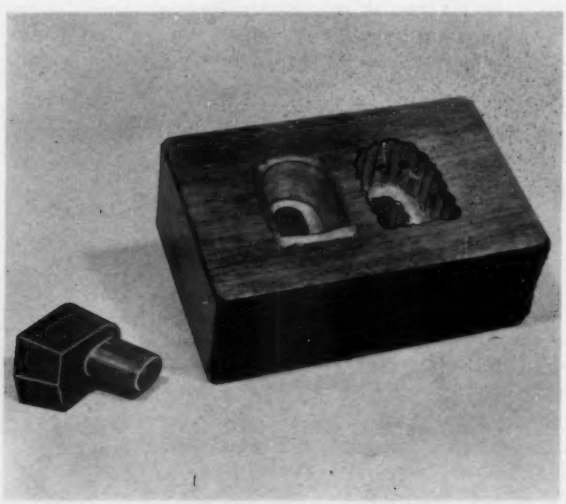


Fig. 17 — Two cavity ram-up core box showing core, fusible alloy lined cavity (left) and roughed out cavity (right) before ramming.

tern. Chop out a cavity in a block of pine or mahogany 1/8-in. larger all around than plaster core pattern. This cavity should be fairly rough and under cut at various places. Drill hole through cavity to outside. Fasten pattern with brad, glue or shellac to a flat surface.

Place block with cavity centralized over pattern. Weight down block and pour in alloy 5. As illustrated in Fig. 18, an extension sprue is to be used to assure a pressure head which helps to pick up all detail on pattern.

Low melting alloys are poor conductors and will solidify slowly in wood and plaster molds, therefore sufficient time for freezing must be allowed – 2, 3, 5 min or more, as needed.

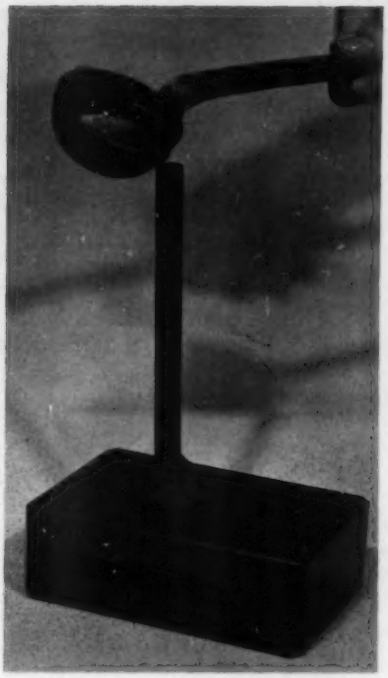


Fig. 18 — Ram-up core box in position over core model. High pouring sprue is optional but its use assures complete fill.

### *Worn Core Boxes Salvaged*

It is no longer necessary to discard worn or damaged wood core boxes. An inexpensive procedure has been developed to restore damaged wood core boxes.

Build a pouring box around the core box. Fill core cavity and pouring box and screen off. When set, remove plaster core and smooth it to perfection by carving or sanding any projections due to dents or nicks in core box.

Using a burr, rotary file or other cutting tool, cut away about 1/8-in. of inside surface of core box cavity. Drill one or more holes (1/4-in.) through core box wall to the cavity.

Replace core on box and clamp or tape together. Warm up core and box before pouring. About 250 F (121 C) for wood – 280 F (138 C) for metal box (core should be baked to eliminate moisture). Through extension sprue pour alloy 3.

### *Epoxy Resin Patterns Fabrication*

Epoxy resins have grown rapidly in popularity among patternmakers recently. To offset some of the extra time required in fabrication of patterns from these materials as well as increasing mold life, a method for using low melting alloy has been developed.

While plaster molds can be used for pouring and curing epoxies they must be dried out first and this requires much time, depending on thickness and volume of plaster.

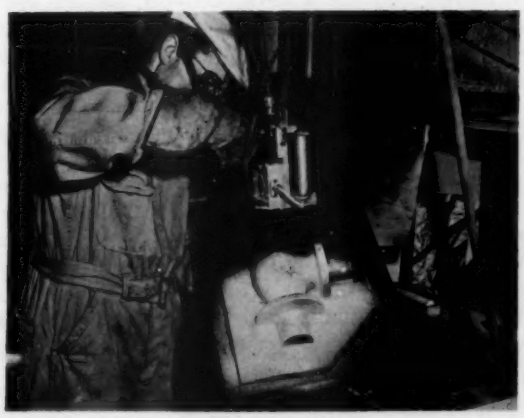


Fig. 19 — Spraying master pattern with thin coat of low melting alloy.



Fig. 20 — Plaster is poured to back up metal coating.

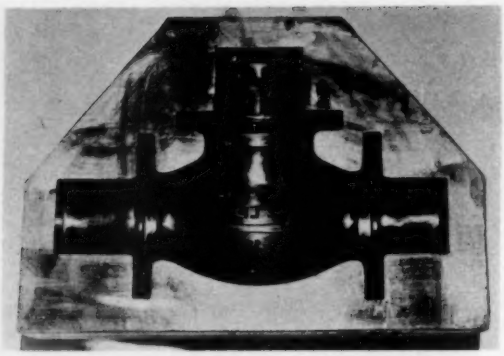


Fig. 21 — Metal faced mold ready for use, dehydration of plaster unnecessary.

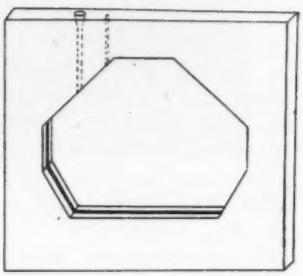


Fig. 22 — Matchplate with pouring hole, vent holes and groove in side opening ready for insertion in pattern.

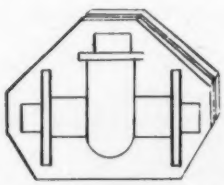


Fig. 23 — Pattern with matching groove ready for insertion.

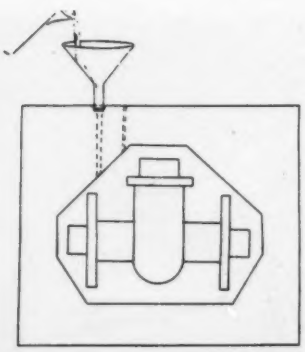


Fig. 24 — Fusible alloy poured to lock pattern in matchplate.

By applying a thin layer of low melting alloy 2 to surfaces of pattern before pouring plaster much time can be saved. That is, the mold will be ready for use as soon as the plaster has set, and no drying out is necessary.

The pattern is fastened down in the same manner as for pouring plaster. After coating pattern with graphite, wax or other parting medium and polishing vigorously, spray on a coating of alloy 2 to a thickness of $\frac{1}{16}$-in. to $\frac{1}{8}$-in. (Fig. 19). Apply coat of quick dry lacquer or shellac to sprayed surface, and when dry pour plaster to fill flask which has been placed around pattern (Fig. 20).

Immediately after plaster has set, pattern may be removed from mold and it is ready for use (Fig. 21). Besides economy in time of construction such metal faced molds offer the benefit of much longer life than unfaced plaster. From 10 to 20 casts can be obtained from a metal faced mold, which is considerably more than can be obtained from regular plaster molds.

A proprietary process for anchoring patterns in matchplates is a novel method of mounting plastic patterns which eliminates nails or screws for fastening. The use of this process has effected appreciable savings in matchplate production there and elsewhere.

Patterns are made of plastic in the usual way in plaster molds and cured. An opening of same shape and size of pattern for loose fit is cut in the wood or metal matchplate. Outside of pattern and inside of hole are grooved at the midpoint and vent and pouring holes drilled to groove in plate (Fig. 22). Pattern is inserted in opening and joint taped to prevent leakage (Fig. 23). Fusible alloy 1 is poured to completely fill groove (Fig. 24). When solid, the metal spline maintains positive anchorage.

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Thanks are also due to Miss Christine Hansen, Secy., and R. S. Darnell, Asst. Mgr. of Cerro Alloy Sales Dept. for valuable assistance in preparation and proof reading.

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# CENTRIFUGALLY CAST STEEL CYLINDERS RAPID GAS HEATING TECHNIQUES

by A. Ayvazian


## ABSTRACT

**An investigation was carried out to study the effects of rapid gas heat treating of centrifugally cast 4330 thick walled steel cylinders in the 150,000-180,000 psi yield strength range. The cylinders were heated for hardening and tempering in specially designed rapid heating gas furnaces. Temperatures used were higher than those ordinarily employed in conventional practice to compensate for the shorter holding time commensurate with the furnace performance capability. Similar cylinders were also heat treated by conventional techniques. Cylinders heat treated by both methods were subjected to metallurgical examinations, and a comparison made of the properties indicated favorable results.**


## INTRODUCTION

The success of many Ordnance applications is attributed to a combination of certain mechanical properties, mainly, ductility and impact properties. Any methods or processes which could improve these properties of Ordnance employed steels would be a considerable accomplishment. Recently, the never ending search for such practices led to the investigation of a high speed gas heating process. In essence, the concept of this process is to subject the charge in the furnace to temperatures in excess of the final desired temperature for a short time, whereby an appreciable thermal gradient is attained between the surface and center of the workpiece.

This is followed by an equalization period of time in order to heat the material uniformly throughout its cross-section. The introduction of this process into the heat treating field as an accepted mode of through heating is not a recent innovation, since the theory and process have been in existence since World War II. However, improved heat processing, attributed primarily to the development of recent ceramic burners, has renewed interest in the application of rapid gas heating. Also, successful employment of rapid heating in the metal forming industry warranted that this manner of heating steel be utilized in order to observe the potentialities of such a process for heat treating of Ordnance steel components.

Exploring the characteristics of this heating technique, it is found that reduction of time and the employment of elevated temperature are the predominant factors which distinguish this method of heating from conventional methods. Heating rates from one to 5 min/in. of thickness are anticipated for plain carbon, low and medium alloyed steel in comparison to the conventionally accepted rates of 30 to 60 min/in. of thickness.

This paper describes the first phase of an investigation conducted to study the feasibility of utilizing this process by comparing the mechanical properties obtained from the rapid gas heat treatments vs. conventional gas heat treatments of centrifugally cast 4330 thick walled steel cylinders in the 150,000-180,000 psi yield strength range.

## PROCEDURE

### *Test Casting*

The test castings selected for this investigation were five actual 90mm centrifugally cast 4330 steel full length gun tubes. Similar castings were being produced and heat treated by conventional methods to meet stringent mechanical properties required by gun tube specifications. It was conceived that, if these test cylinders, heat treated by the rapid heating process, produced mechanical properties comparable to those being obtained by the conventionally treated cylinders, then this would be one criterion by which this process could be evaluated.

The cylinders to be tested were produced from

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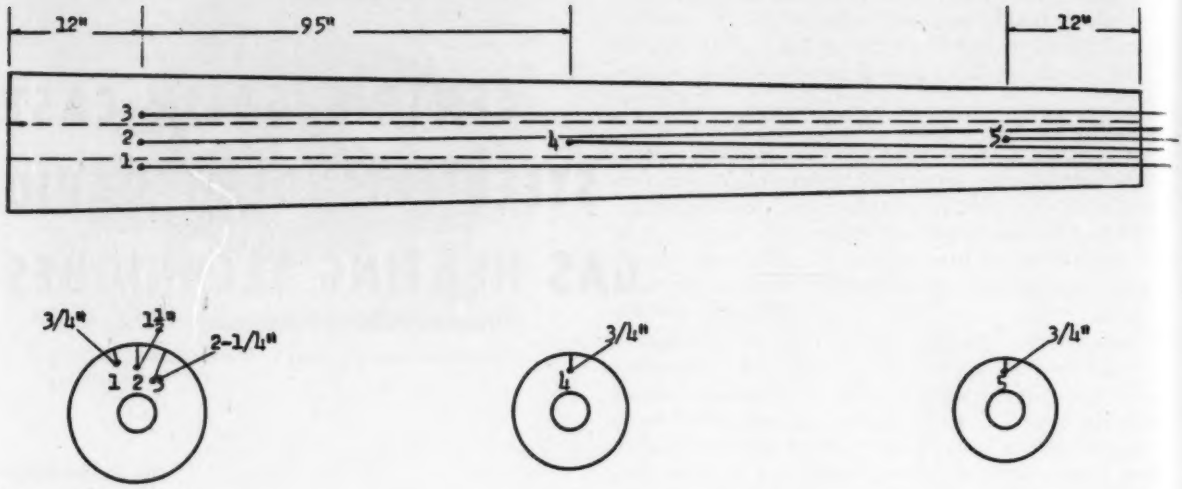



Fig. 1 — Schematic showing location and depth of thermocouples.

different heats with final typical chemical composition of:

| C | Mn | P | S | Si | Ni | Cr | Mo | V |
|---|---|---|---|---|---|---|---|---|
| 0.285 | 0.64 | 0.011 | 0.019 | 0.24 | 2.24 | 0.84 | 0.47 | 0.22 |

### *Preparation and Heat Treatment of Test Specimens*

The cylinders, numbered 5, 6, 7, 8 and 9, were normalized at 1750 F (954 C) prior to the tests. After the normalize, they were machined so that the dimensions were approximately 17 ft long with outside diameter of 8½-in. at one end tapering down to about 5½-in. outside diameter at the other end. Each cylinder had a bore of about 2½-in. Chromel-alumel thermocouples were attached to each cylinder, as shown in Fig. 1. Three thermocouples were located one foot from the heaviest end, and penetrated into the surface at distances of ¾, 1½ and 2¼-in. A fourth thermocouple was placed at midlength, ¾-in. into the surface and a fifth thermocouple was placed one ft from the other end, also ¾-in. into the surface. These five thermocouples were attached to a multiple-point recorder which recorded the temperature distribution in the cylinder as it was heated.

Prior to heating the test cylinders, preliminary tests were conducted with surplus specimens in order to arrive at some procedure which would produce yield strength values within the 150,000-180,000 psi range. From these tests, the most applicable treatments were selected for the five test cylinders. The hardening treatment in all five specimens was similar and consisted of vertically placing the specimens which were to be treated, one at a time, into a specially designed vertical furnace with 36 ceramic burners.

The furnace, which contained two controlled zones, was operating at 2300 F (1260 C) at the bottom zone

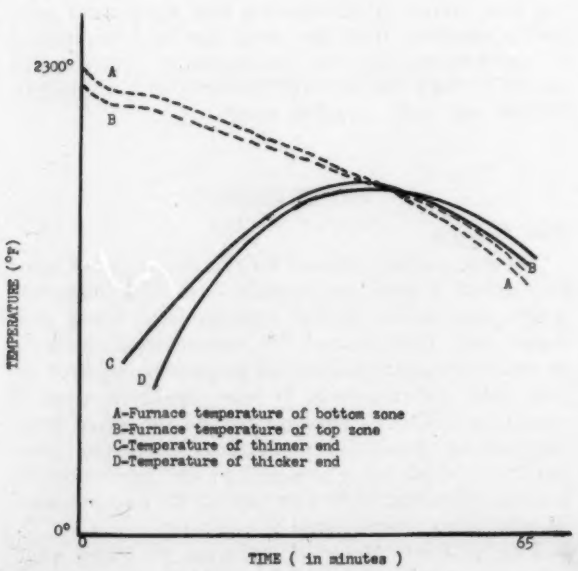



Fig. 2 — Schematic graph showing general shape of curves representing austenitizing cycle of cylinder vs. furnace temperature.

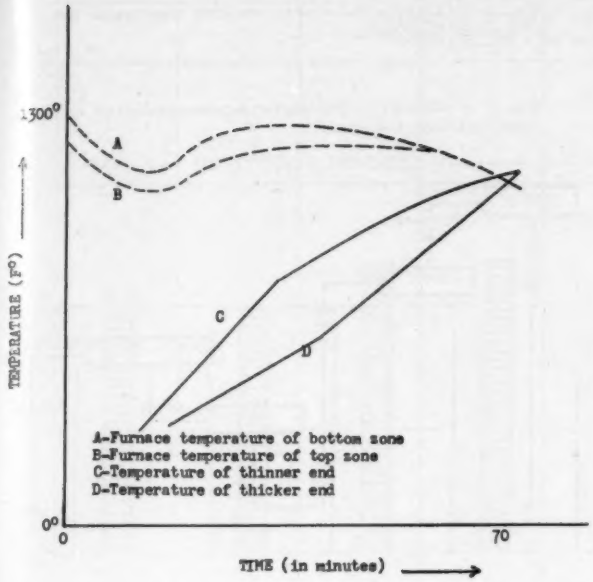



Fig. 3 — Schematic graph showing general shape of curves representing tempering cycle of cylinders vs. furnace temperature.

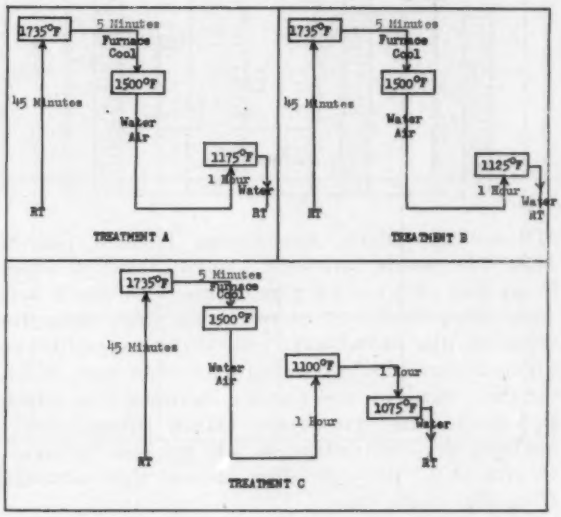



Fig. 4 — Schematic showing treatments employed in final tests.

and 2100 F (1143 C) at the top zone when the specimen was placed into the furnace. The differential in furnace temperature was to compensate for the taper in the cylinder. As the temperature of the specimen gradually increased, the furnace temperature was progressively lowered (Fig. 2). The specimen was periodically rotated in the furnace to allow for uniform heating. In 45 min both furnace and specimen reached 1735 F (946 C), the austenitizing temperature.

The burners were shut and the specimen allowed to cool to 1500 F (810 C) for 15 min prior to water quenching. The quenching consisted of completely submerging into cold water for 2 min, followed by a tapered quench for 1½ to 2 min. Subsequently, the cylinder was cooled in water to near room temperature after a 4 min equalization period in air at around 600 F (316 C). The complete heating and cooling process was followed on the multiple-point recorder.

### *Tempering Procedure*

The technique employed for the tempering was similar to the hardening procedure; however, five slightly different final temperatures were used. The specimen was lowered into another special vertical furnace containing 16 burners. The bottom zone was operating at 1300 F (704 C) and the top zone at 1200 F (649 C). In about 50 min, when both the specimen and the furnace reached a desired temperature (Fig. 3), the burners were shut and the specimen allowed to soak for 10 min prior to water quenching to room temperature. The final temperatures and times are shown in Figs. 4 and 5.

In contrast to the above treatments, the conventional production heat treatment for such cylinders consists of slowly heating to 1600 F (871 C), holding 5 hr, furnace cooling to 1500 F (810 C), holding 2 hr and taper quenching in water until cylinder reaches 600 F (316 C), holding in air for an equalization period prior to cooling in oil to room temperature. The tempering cycle consists of heating to 600 F (316 C), holding 3 hr, raising temperature to 1000 F-1060 F (538-571 C), holding 5½-hr prior to water cooling (Fig. 6).

Mechanical properties of cylinders 1, 2, 3 and 4, heat treated in this manner, are listed in Tables 1 and 3, in order to provide data with which the rapidly heat treated test cylinders can be compared.

### *Mechanical Tests*

**Tensile Tests.** The testing of the cylinders was performed in accordance with Military Specification

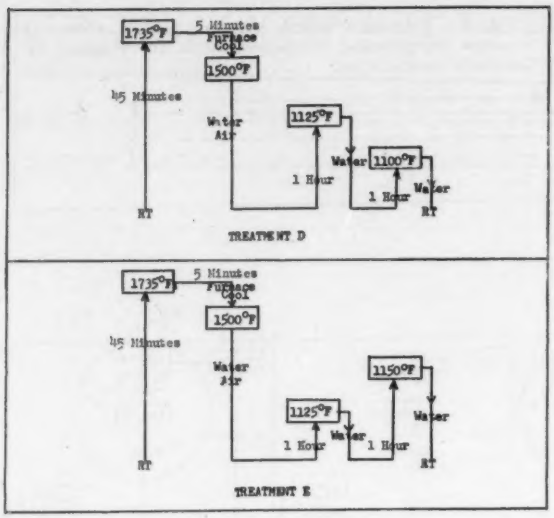
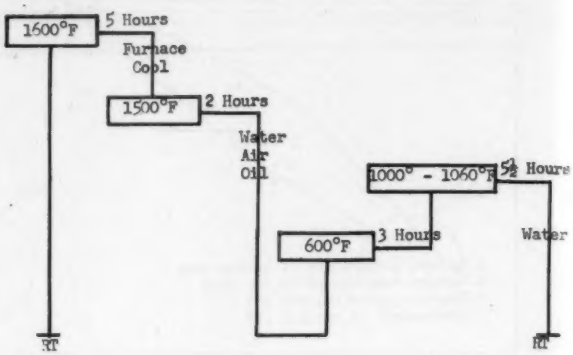



**Fig. 5 — (Left) — Schematic showing treatments employed in final tests.**

**Fig. 6 — (Below) — Schematic representation of conventional heat treatment.**




MIL-S-10026 (ORD), Amendment 1 dated Jan. 5, 1950. The tensile bars were standard 0.357 in. taper shank bars with 1.40 in. gage length. Two tensile bars were taken from each of two ¾-in. thick disks, cut perpendicular to the major axis of the test cyclinders. Figure 7 shows the approximate locations from which the disks were cut, and also the locations from which the tensile bars were taken. Tensile strength, yield strength, per cent elongation and per cent reduction in area at 0.1 per cent offset method were obtained from the tensile bars.

**Charpy Bars.** The Charpy bars were standard 0.394 in. square by 2.156 in. long V-notched impact bars. A total of three bars was taken from each cylinder, two from the thicker end and one from the other end (Fig. 7).

### *Microstructural Examination*

All test cylinders were subjected to microstructural examination after the rapid heat treatments in order to correlate the changes, if any, with those obtained in the microstructure of the conventionally heat treated cylinders. The photomicrographs were taken from two small cube specimens, one cut from an area close to the outer surface and the other from an area close to the bore, as shown in Fig. 7.

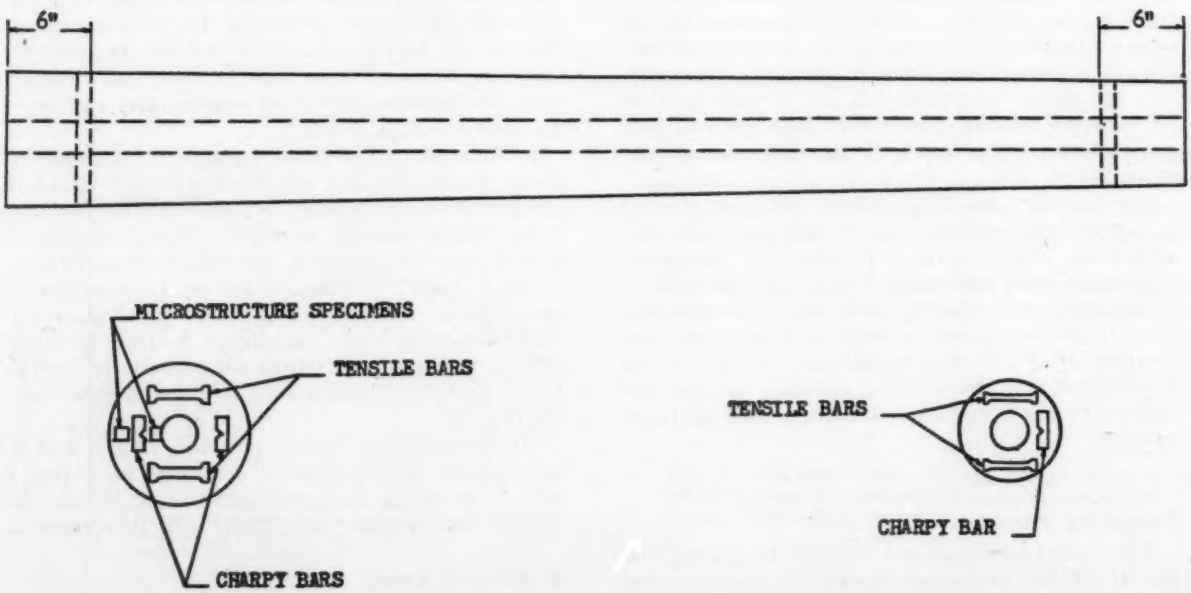



**Fig. 7 — Schematic showing location from which test specimens were taken.**

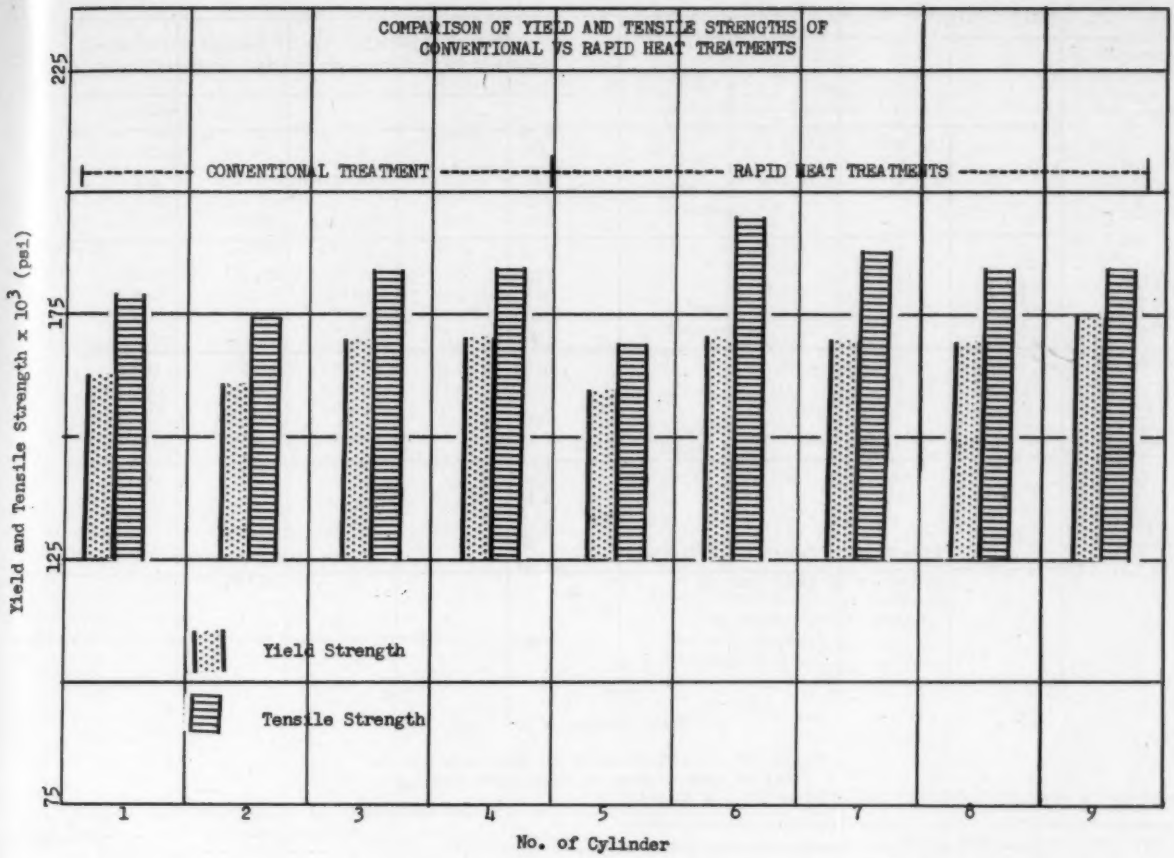



Fig. 8 — Comparison of yield and tensile strengths of conventional vs. rapid heat treatment.

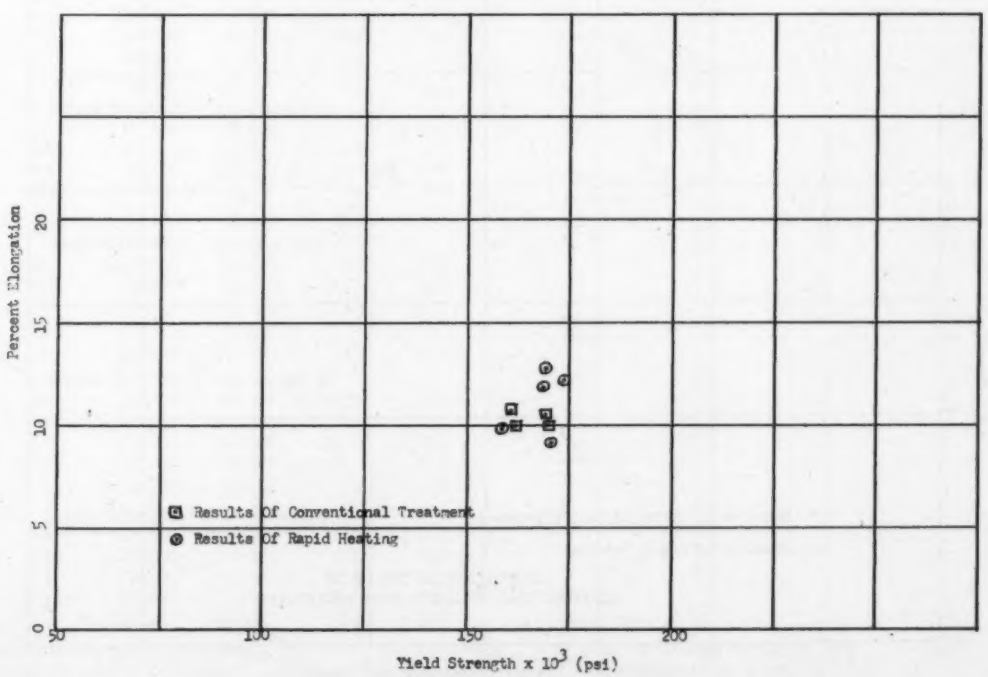



Fig. 9 — Comparison of per cent elongation of conventional vs. rapid heat treatment.

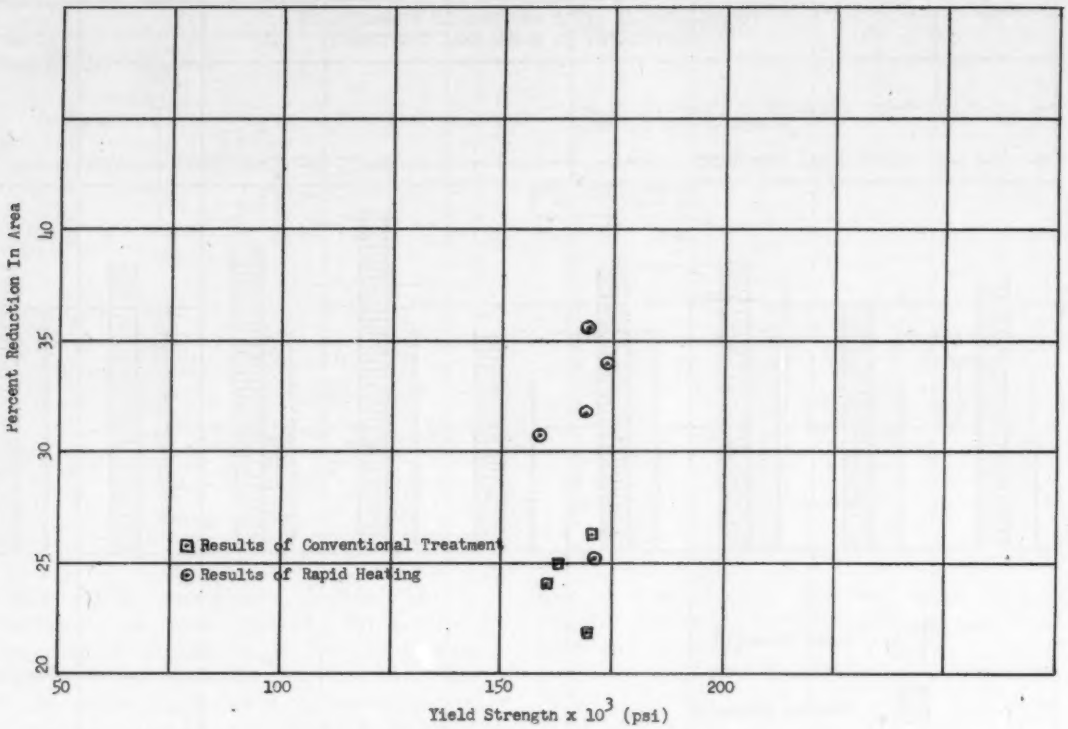



Fig. 10 — Comparison of per cent reduction in area of conventional vs. rapid heat treatment.

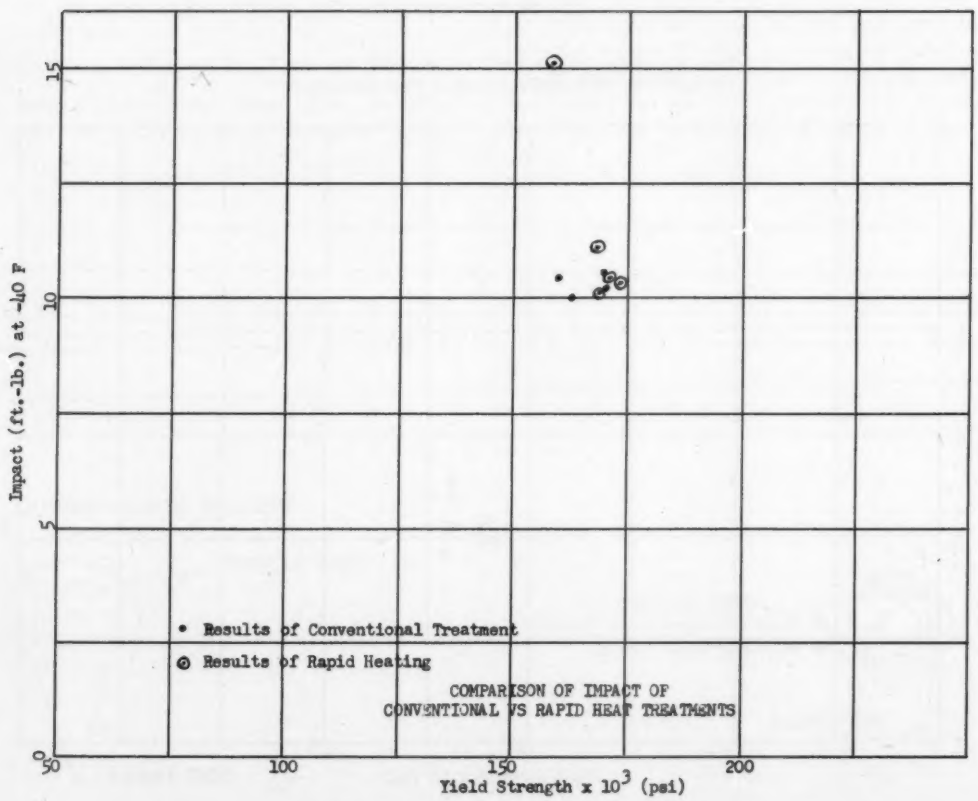



Fig. 11 — Comparison of impact of conventional vs. rapid heat treatment.

TABLE 1 — RESULTS OF TENSILE TESTS
CONVENTIONALLY HEAT TREATED CYLINDERS

| Cylinder No. | Location of Test Specimen* | Y.S., 0.1 % | T.S. | Elong., % | R.A., % |
|---|---|---|---|---|---|
| 1 | B-6 | 165,000 | 180,500 | 9.3 | 21.6 |
| | B-12 | 165,000 | 181,500 | 11.4 | 27.0 |
| | T-12 | 158,000 | 173,000 | 9.3 | 26.5 |
| Average | | 162,600 | 178,300 | 10.0 | 25.0 |
| 2 | B-6 | 161,500 | 172,500 | 13.6 | 23.1 |
| | B-12 | 164,000 | 179,000 | 10.7 | 24.6 |
| | T-12 | 156,500 | 171,000 | 7.1 | 24.6 |
| Average | | 160,600 | 174,200 | 10.7 | 24.1 |
| 3 | B-6 | 171,750 | 185,000 | 10.0 | 19.6 |
| | B-12 | 165,500 | 179,000 | 11.4 | 24.6 |
| | T-12 | 170,500 | 185,500 | 10.0 | 21.6 |
| Average | | 169,275 | 183,200 | 10.5 | 21.9 |
| 4 | B-6 | 171,250 | 185,500 | 17.9 | 24.1 |
| | B-12 | 171,000 | 184,500 | 7.1 | 25.5 |
| | T-12 | 168,750 | 180,500 | 5.0 | 29.8 |
| Average | | 170,300 | 183,500 | 10.0 | 26.4 |

*B — Bottom (thicker end of cylinder).
T — Top (thinner end of cylinder).
6 — o'clock position.
12 — o'clock position.

TABLE 2 — RESULTS OF TENSILE TESTS

| Cylinder No. | Location of Test Specimen* | Y.S., 0.1 % | T.S. | Elong., % | R.A., % |
|---|---|---|---|---|---|
| | **Rapid Heating — Treatment A** | | | | |
| 5 | B-6 | 155,500 | 166,000 | 3.6 | 22.6 |
| | B-12 | 160,100 | 167,700 | 9.3 | 15.1 |
| | T 6 | 158,500 | 168,500 | 12.1 | 46.5 |
| | T-12 | 160,000 | 171,000 | 13.6 | 38.5 |
| Average | | 158,500 | 168,300 | 9.7 | 30.7 |
| | **Rapid Heating — Treatment B** | | | | |
| 6 | B-6 | 174,500 | 203,600 | 7.1 | 23.1 |
| | B-12 | 172,000 | 203,200 | 4.3 | 8.2 |
| | T-6 | 168,000 | 184,000 | 12.9 | 34.5 |
| | T-12 | 170,000 | 185,600 | 12.1 | 34.9 |
| Average | | 171,100 | 194,100 | 9.1 | 25.2 |
| | **Rapid Heating — Treatment C** | | | | |
| 7 | B-6 | 166,000 | 189,600 | 11.4 | 34.0 |
| | B-12 | 173,500 | 193,000 | 11.4 | 32.2 |
| | T-6 | 169,500 | 183,800 | 14.3 | 41.5 |
| | T-12 | 169,500 | 183,600 | 13.6 | 34.5 |
| Average | | 169,600 | 187,500 | 12.7 | 35.6 |
| | **Rapid Heating — Treatment D** | | | | |
| 8 | B-6 | 162,500 | 183,500 | 12.9 | 34.0 |
| | B-12 | 167,000 | 186,000 | 12.1 | 32.6 |
| | T-6 | 171,000 | 184,500 | 10.0 | 29.4 |
| | T-12 | 172,000 | 186,500 | 12.1 | 31.2 |
| Average | | 168,100 | 185,100 | 11.8 | 31.8 |
| | **Rapid Heating — Treatment E** | | | | |
| 9 | B-6 | 172,500 | 184,500 | 10.0 | 25.5 |
| | B-12 | 172,000 | 186,000 | 12.1 | 35.8 |
| | T-6 | 176,000 | 185,000 | 12.9 | 40.2 |
| | T-12 | 175,500 | 184,500 | 13.6 | 34.5 |
| Average | | 174,000 | 185,000 | 12.2 | 34.0 |

*B — Bottom (thicker end of cylinder).
T — Top (thinner end of cylinder).
6 — o'clock position.
12 — o'clock position.

TABLE 3 — RESULTS OF CHARPY TESTS AT −40 F

| Cylinder No. | Location of Test Specimens* | Conventional Treatment, Ft-Lb |
|---|---|---|
| 1 | B-3 | 8 |
| | B-9 | — |
| | T-3 | 12.0 |
| Average | | 10.0 |
| Required | | 12.0 |
| 2 | B-3 | 10.5 |
| | B-9 | — |
| | T-3 | 11.0 |
| Average | | 10.8 |
| Required | | 12.0 |
| 3 | B-3 | 10.6 |
| | B-9 | — |
| | T-3 | 11.5 |
| Average | | 11.1 |
| Required | | 11.0 |
| 4 | B-3 | 10.5 |
| | B-9 | — |
| | T-3 | 10.0 |
| Average | | 10.3 |
| Required | | 10.0 |

*B—Bottom (thicker end of cylinder).
T—Top (thinner end of cylinder).
3—o'clock position.
9—o'clock position.

TABLE 4 — RESULTS OF CHARPY TESTS AT −40 F

| Cylinder No. | Location of Test Specimens* | Rapid Heating Treatments, Ft-Lb | | | | |
|---|---|---|---|---|---|---|
| | | A | B | C | D | E |
| 5 | B-3 | 15.9 | — | — | — | — |
| | B-9 | 14.5 | — | — | — | — |
| | T-3 | — | — | — | — | — |
| | T-9 | — | — | — | — | — |
| Average | | 15.2 | | | | |
| Required | | 13.0 | | | | |
| 6 | B-3 | — | 8.4 | — | — | — |
| | B-9 | — | 9.7 | — | — | — |
| | T-3 | — | 12.4 | — | — | — |
| | T-9 | — | 13.0 | — | — | — |
| Average | | | 10.9 | | | |
| Required | | | 10.0 | | | |
| 7 | B-3 | — | — | 10.0 | — | — |
| | B-9 | — | — | 7.5 | — | — |
| | T-3 | — | — | 11.5 | — | — |
| | T-9 | — | — | 12.4 | — | — |
| Average | | | | 10.4 | | |
| Required | | | | 11.0 | | |
| 8 | B-3 | — | — | — | 10.9 | — |
| | B-9 | — | — | — | 11.5 | — |
| | T-3 | — | — | — | 14.9 | — |
| | T-9 | — | — | — | 10.9 | — |
| Average | | | | | 12.1 | |
| Required | | | | | 11.0 | |
| 9 | B-3 | — | — | — | — | 10.6 |
| | B-9 | — | — | — | — | 11.5 |
| | T-3 | — | — | — | — | 10.6 |
| | T-9 | — | — | — | — | 9.5 |
| Average | | | | | | 10.6 |
| Required | | | | | | 10.0 |

*B—Bottom (thicker end of cylinder).
T—Top (thinner end of cylinder).
3—o'clock position.
9—o'clock position.

## RESULTS AND DISCUSSION

The commercial mechanical property specifications for these cylinders is 150,000-180,000 psi yield strength at 0.1 per cent offset, 22.0-20.4 per cent reduction in area and 14.0-10.0 ft-lb impact at −40 F for this yield strength range. In Table 2 are the tensile properties obtained from the test cylinders which were subjected to the rapid heat treatments. These results, compared with those listed in Table 1, which represents typical tensile properties obtained when similar cylinders are conventionally heat treated, indicated a trend existed toward improved tensile properties.

This can more readily be observed in Fig. 8, which is a bar graph of the average yield and tensile strengths listed in Tables 1 and 2. Figures 9 and 10, show that for comparable yield and tensile strengths, the ductility values of the rapidly treated specimens were superior in most cases to those obtained from conventional treatments. This was especially noticed in Fig. 10, which represents the average results of the per cent reduction in area. A substantial improvement was seen in this important mechanical property.

The specimens which were subjected to a double temper disclosed better and more uniform results than the single temper. This indicated that the shorter tempering cycles apparently were not sufficient to completely temper the specimens. A complete list of the per cent elongation and per cent reduction in area values are found in Tables 1 and 2.

### *Impact Values*

Since it was of vital importance that cylinders of the type used in the studies contain a high degree of resistance to shock, the acquisition of any significant improvement in the impact properties would be considered a worth-while contribution. The data resulting from the five tests indicated impact values comparable to those obtained from the production treated cylinders (Fig. 11).

These impact values would be accepted under production standards. However, they were, as was the case of the conventionally heat treated cylinders, not ideal, since they were minimum average values in most instances (Tables 3 and 4). Some of the single values listed in Table 4 were encouraging, but these were offset by some low values which when averaged with the higher values, gave minimum results. Acquisition of the low values was disappointing, and it seemed to point to the fact that perhaps some incomplete transformation due to nonuniform heating was responsible. However, no studies were conducted during this investigation to substantiate this factor.

Microstructural examination of the test cylinders disclosed similar basic structures to those obtained after conventional heat treatments. The microstructure of the conventionally treated specimens revealed a small amount of tempered bainite in a matrix of tempered martensite, with precipitated carbides scattered throughout the structure especially along grain boundaries. The microstructure of the rapidly heated specimens also disclosed tempered bainite in a matrix of tempered martensite. However, a slightly greater amount of tempered bainite was present.

The amount of resolved carbide was dependent on the tempering temperature and time at temperature. The treatments requiring longer times indicated more resolved carbides present. The overall structures of these specimens were rather heterogeneous, as compared to the conventionally treated structures. However, since the mechanical properties were at least comparable to those obtained from the conventionally treated cylinders, no attempt was made to study this parameter in this phase of the investigation.

### *Grain Size*

The grain size for the conventionally heat treated cylinders ranged from A.S.T.M. 8 to 9 throughout, while those of the rapidly heated cylinders ranged from A.S.T.M. 6 to 7 near the surface and 8 to 9 close to the bore.

In view of the test results, the application of rapid gas heating as a method of heat treating has shown some significant potentialities. Up to recent times, heating of steel for certain types of fabrication was generally done at the rate of 20 to 35 min/in. of diameter, and in many instances with highly alloyed steels even up to 60 min/in. was advocated. There was a certain amount of reluctance to heat at faster rates than mentioned above in order to avoid internal cracks caused by thermal stresses. However, no such results were experienced in this investigation.

## CONCLUSION

Due consideration was given the new mechanical properties which were comparable, and in some instances superior, to those established by the conventional heat treating practices. Working within the prescribed yield strength range, the final tests indicated that most of the reduction in area properties increased in value while some of the elongation values also showed improvements. The impact properties, however, did not disclose any significant changes.

It is apparent from these studies that in order to improve the mechanical properties and to obtain a more uniform structure, a different initial prehardening treatment is necessary. These studies consisted primarily of determining the feasibility of utilizing this rapid process for heat treating cylindrical objects by comparison of mechanical properties. No quantitative attempt was made to establish any substantially different heating cycles than those used during the investigation.

The temperatures employed were basically those used in the conventional heating cycles, with some modifications to compensate for the shorter holding times. These modifications included higher austenitizing and tempering temperatures than generally used in order to obtain the required mechanical properties. Also double tempering was employed in several treatments in anticipation of obtaining more uniformity in structure and properties.

Since favorable results emanated from this work, further work will be conducted in order to study the difficulties and questionable parameters which were encountered during this investigation.

# ALUMINUM FURNACE REFRACTORIES

by C. H. Schweinsberg and J. L. Dolph


## ABSTRACT

**Research in the development of refractories for aluminum furnaces has been particularly active in recent years. New techniques in casting and forming aluminum from an ever increasing variety of alloys has made this work important. Research leading to recent developments, as well as some test procedures, will be briefly reviewed, and compositions, properties and experience with brick, castables, ramming mixes and coating materials now available will be discussed.**


## INTRODUCTION

Refractories developed specifically for aluminum melting furnaces provide for substantially all the different operating conditions. There are a number of materials available to meet these varying conditions, and proper selection is necessary for best economy. Frequently, several types of refractories are tried in service to determine which is best suited to an individual operation.

A survey of literature and reports on refractories for aluminum melting furnaces reveals the wide selection of materials proposed for this application. Although some have been discarded as impractical or lacking in properties now considered essential, the work done has contributed greatly to the problems involved. Refractories which have been investigated in service include graphite, magnesite, chrome, chrome-magnesite, silicon carbide, zircon, fused alumina, magnesia-alumina spinel and fused-cast millite.

## TEST PROCEDURES

Research on this particular problem has been particularly active in recent years, and new products developed are giving excellent service in numerous installations.

Long ago it was learned that molten aluminum and its alloys removed silicon from refractories, and this was identified as a thermite type, heat releasing, reaction.[1] Since silicon was objectionable in many alloys, the use of higher alumina refractories with lower silica content became widely used along with other types which resisted reaction with molten metal.

For a long time studies were made by cutting cups or troughs in brick to be tested with molten aluminum. An alloy of aluminum, such as 7075, was placed in the cup, the brick heated to a predetermined controlled temperature and held there for a definite time, usually 72 hr. After cooling, the brick was cut through the cup, and the degree of penetration or reaction was measured in comparison with a control sample.

This method is satisfactory to roughly indicate the relative merit or various compositions, but it does not simulate actual furnace conditions as well as the immersion test now used. In the immersion test, samples are partially submerged or suspended in molten aluminum in a laboratory furnace, such as the one shown in Fig. 1.

This furnace is heated by electric elements and the temperature can be closely controlled. The hearth of this particular furnace can be removed through the bottom and also serves for testing purposes. Figure 2 shows the cross-section of a test hearth built of two types of brick, differing in their ability to resist penetration of aluminum.

The modulus of rupture of a brick is a test of transverse strength and is generally considered an indication of ability to withstand mechanical abuse. The test data usually quoted are obtained at room temperatures and do not necessarily have a relation to the strength at operating temperatures. In Table 1, results are shown of tests run at 75 F (24 C) and at

**TABLE 1 — MODULUS OF RUPTURE — LB/SQ IN.**

| | At 75 F (24 C) | | At 1400 F (760 C) | |
|---|---|---|---|---|
| | Unused Brick | After Immersion In Molten Aluminum | Unused Brick | After Immersion In Molten Aluminum |
| Fireclay Brick | 1550 | Cracked | 1640 | 1220 |
| 90% Alumina Brick | 1370 | 5050 | 1410 | 1000 |
| 85% Alumina Brick, Phosphate-Bonded (Unburned) | 1640 | 1810 | 2780 | 2390 |

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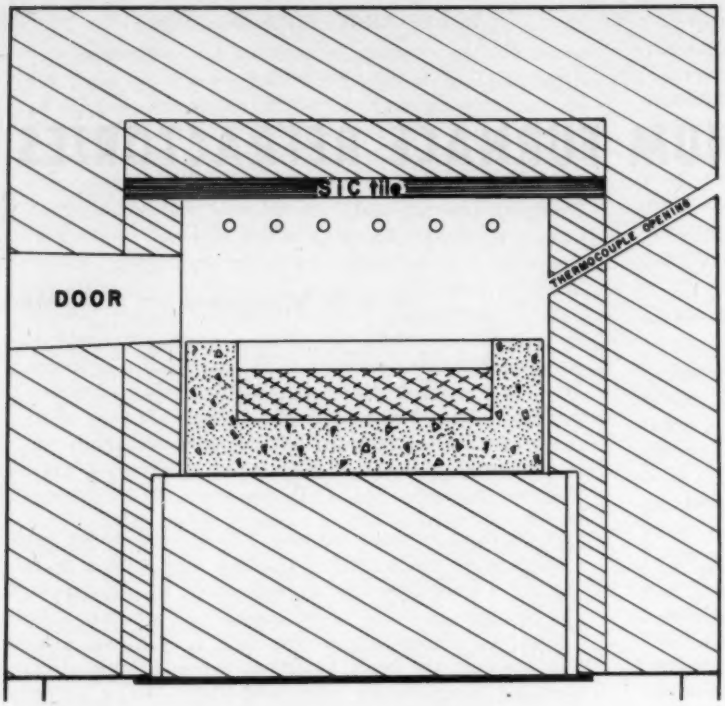



Fig. 1 — Cross-section of laboratory aluminum furnace.

1400 F (760 C) on brick before and after immersion in molten aluminum.

Of particular significance is the fact that bricks which are penetrated by aluminum in the immersion test show a resulting loss of strength in the hot modulus test.

Table 2 shows the modulus of rupture of two mortars tested under the same conditions as the brick in Table 1. The sodium silicate bonded mortar lost over 70 per cent of its transverse strength at 1400 F (760 C) due to penetration of molten aluminum, while the phosphate-bonded mortar showed no loss of strength. The effect of the aluminum reinforcement was apparent in the sodium silicate bonded mortar. The penetrated aluminum is soft at operating temperatures and is ineffective when strength is needed.

TABLE 2 — MODULUS OF RUPTURE — LB/SQ IN.

| | At 75 F (24 C) | | At 1400 F (760 C) | |
|---|---|---|---|---|
| | Unused Brick | After Immersion In Molten Aluminum | Unused Brick | After Immersion In Molten Aluminum |
| High-Alumina Mortar, Sodium Silicate Bond | 1450 | 1890 | 1480 | 410 |
| High-Alumina Mortar, Phosphate Bond | 900 | 1020 | 1230 | 1320 |



Fig. 2 — Cross-section of hearth from laboratory furnace.

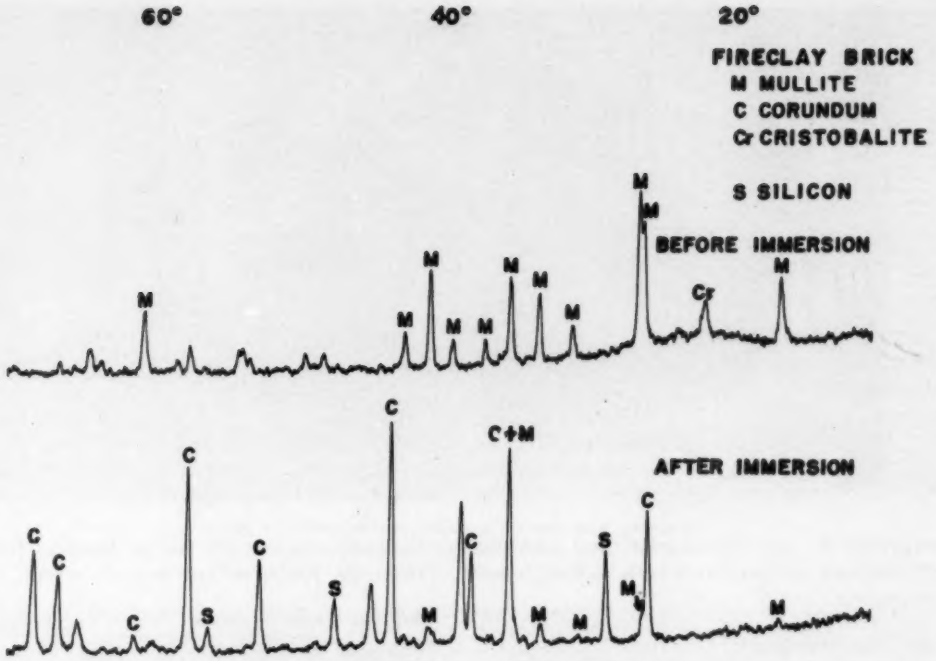



Fig. 3 — X-ray defraction of fireclay brick before and after immersion in molten aluminum.

It has been established that silicon pickup and dross buildup is generally closely proportional to the penetration and the silica content of the refractory.[2]

Figure 3 shows an x-ray diffraction pattern which graphically illustrates the crystalline phases exhibited by fireclay brick, containing approximately 40 per cent alumina ($Al_2O_3$), before and after immersion in molten aluminum for three days at 1500 F (816 C). The original brick was composed principally of mullite ($3\,Al_2O_3 \cdot 2\,SiO_2$), some cristobalite and a moderate amount of amorphous material which is probably glass.

After immersion, there was no detectable cristobalite, the mullite had been reduced by 85 per cent and the glass is significantly less. However, there are now present sizable amounts of corundum, elemental silicon and aluminum metal. Obviously, the silica has been reduced to silicon and picked up by the aluminum.

This reaction would decrease with increased alumina content, and should be negligible with refractories in the 85 to 99 per cent alumina class.

Laboratory results are important for determining the types or composition of refractories that could be economically adapted to operating conditions. Test panels in production furnaces also are important, and frequently will give positive indication of the service that may be expected. However, the only true test is a complete installation in a furnace operating under conditions for which the refractory is designed.

## TYPES OF REFRACTORIES

In the upper side walls and roofs, high-duty or super-duty fireclay brick, plastics or castables are usually adequate, with the trend toward the super-duty materials. The critical parts in an aluminum melting furnace are the bottom and side walls to a point approximately 12 to 18 in. above the sill level. Dense, high-duty fireclay brick of the class used in iron blast furnaces continue to be used widely, particularly where alloying and fluxing conditions are not severe.

High-alumina bricks of 60 per cent $Al_2O_3$ content are also used extensively, and one operator has described this type as the workhorse of his furnace. Frequently, 60 per cent $Al_2O_3$ brick have proved to be a satisfactory compromise between high-duty fireclay brick and the types containing 85 per cent or more $Al_2O_3$ which costs more. Bricks having an alumina content of 90 per cent have been outstanding when operating conditions are severe and when critical alloys are being melted.

High-duty fireclay brick and high-alumina brick of both 60 to 90 per cent $Al_2O_3$ classes are all reacted and penetrated to a degree which in some cases can be damaging. The cup test in Figure 4 shows a high-duty fireclay brick after 30 hr at 1700 F (927 C) using 7075 alloy. It will be observed that there is a definite separation between the penetrated and unpenetrated zones. This is typical of brick which are readily penetrated, but is much less apparent with 90 per cent alumina brick. It has also been observed that 90 per cent alumina bricks which have been reacted show little, if any, change in volume.

Actually, this separation, or crack, may be a fallacy of the cup test and does not necessarily occur in service, since these types of brick have frequently given years of service with no evidence of such damage. Ninety-nine per cent $Al_2O_3$ bricks of excellent

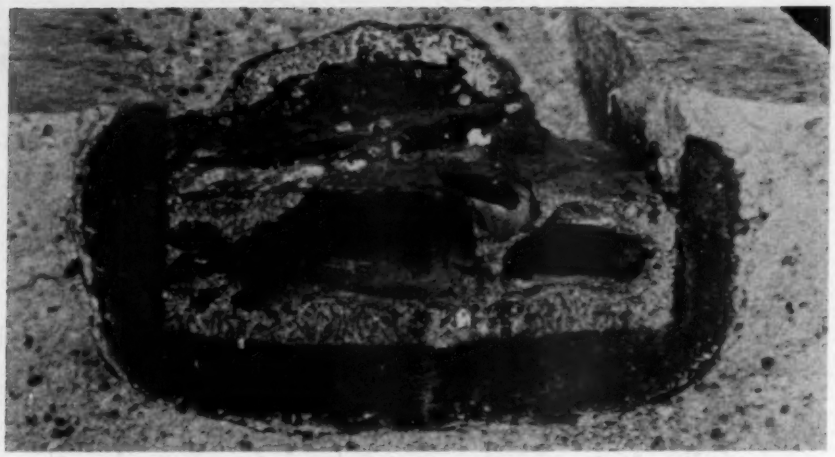


Fig. 4 — Cup test showing aluminum penetration in fireclay brick.

quality are produced, but due to cost their use has been largely confined to operations for which extreme purity is paramount.

### *85 Per Cent Alumina Brick*

Within the past several years 85 per cent alumina bricks made with special bonding agents or additives to increase resistance to metal penetration have been developed. Most prominent have been the phosphate-bonded brick which are manufactured both unburned and high fired.

The unburned type is cured at approximately 500 F (260 C) to 600 F (316 C) to stabilize the aluminum-phosphate bond. These bricks have excellent resistance to penetration, but generally are not as strong as the high-fired types. The modulus of rupture of the unburned brick ranges from 1100 to 2200 psi, as compared to over 3000 psi or higher for the fired brick. Unburned phosphate-bonded brick with alumina content as low as 50 per cent have been developed, and are best adapted principally for lining furnaces operated under relatively moderate conditions.

The high-fired 85 per cent alumina phosphate-bonded brick have unusual strength and also excellent resistance to reaction. They are used successfully in furnaces where contamination from the refractory would be objectionable, and they also contribute to easier cleaning of the furnace walls.

More recently, bricks of 85 per cent alumina content or higher bonded by special chemical additives have been developed. Preliminary trials indicate that their resistance to penetration or reaction by molten aluminum will be superior to that of other high-alumina refractories. This is illustrated by Fig. 5, in which the new type refractory is compared with a 90 per cent alumina brick after immersion in molten aluminum for three days. Figure 6 illustrates the same two brands of brick after three weeks in molten aluminum.

Hard-burned chrome and chrome-magnesite bricks have excellent resistance to wetting by molten aluminum, but their spalling and impact resistance is lower than that of most high-alumina brick. They are particularly adapted for use in holding furnaces where charging practices are not so severe.

### *Silicon Carbide Brick*

Silcon carbide bricks, have good strength, and since dross does not readily adhere to them cleaning is simplified. Formation of aluminum carbide will result in disintegration of the brick during shutdown periods, but special compositions have been developed to retard this reaction. Silicon nitride bonded silicon carbide bricks are used principally for tap-out shapes, pyrometer tubes and pump parts. High cost is a limiting factor to the wide use of these refractories.

Zircon bricks are not penetrated by molten aluminum at normal operating temperatures, but they do not withstand the mechanical abuse or impact of hard charging practices. Zircon, the silicate of zirconium, is an acid refractory and may be damaged by basic fluxes at high temperatures. Zircon bricks also are of relatively high cost, and they are used principally in furnaces where metal purity is paramount.

The bonding mortar used with any type of brick is highly important, since weak or readily penetrated joints are vulnerable to attack. It is generally recognized that mortars bonded with sodium silicate react with aluminum, as illustrated in Fig. 7. This reaction is frequently accompanied by expansion of the joint, resulting in distortion of the wall. Phosphate-bonded zircon and high-alumina mortars are now available which are resistant to penetration. Figure 8 illustrates this clearly. The bricks are of the 90 per cent alumina class, and the mortar is a phosphate-bonded 85 per cent alumina material.

Plastics, castables and ramming mixes have been used with varying results. Most of these materials are not greatly reacted upon but lack adequate strength at operating temperatures, which largely restricts their use to furnaces employing moderate charging procedures. However, monoliths have been entirely adequate in many applications.

Chrome base castables especially are nonwetting, and when installed by ramming, rather than casting, have given satisfactory service. Chrome castables also show promise used as a maintenance material when

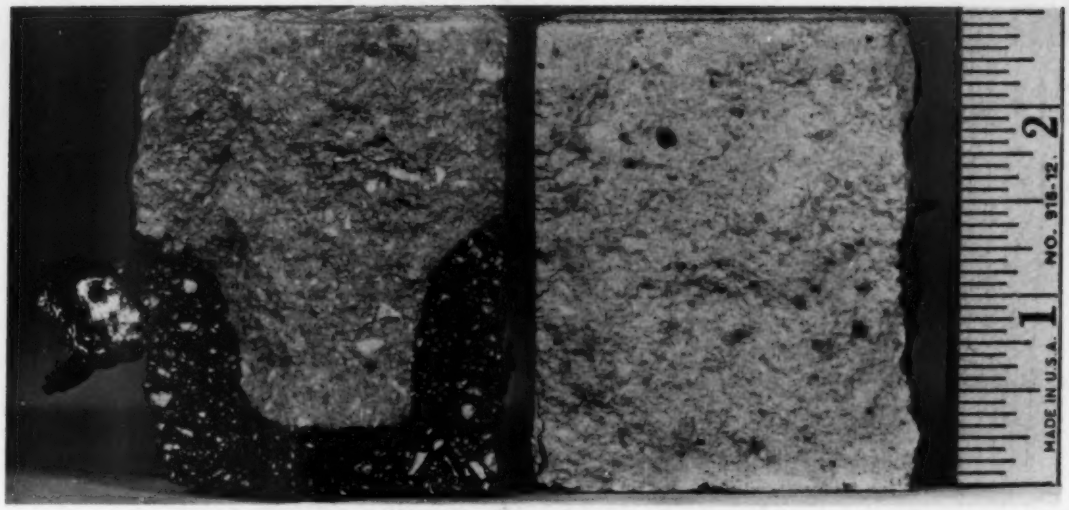



Fig. 5 — Comparison between 90 per cent alumina brick. *(left)* and special bonded 85 per cent alumina brick *(right)* after three days in molten aluminum.

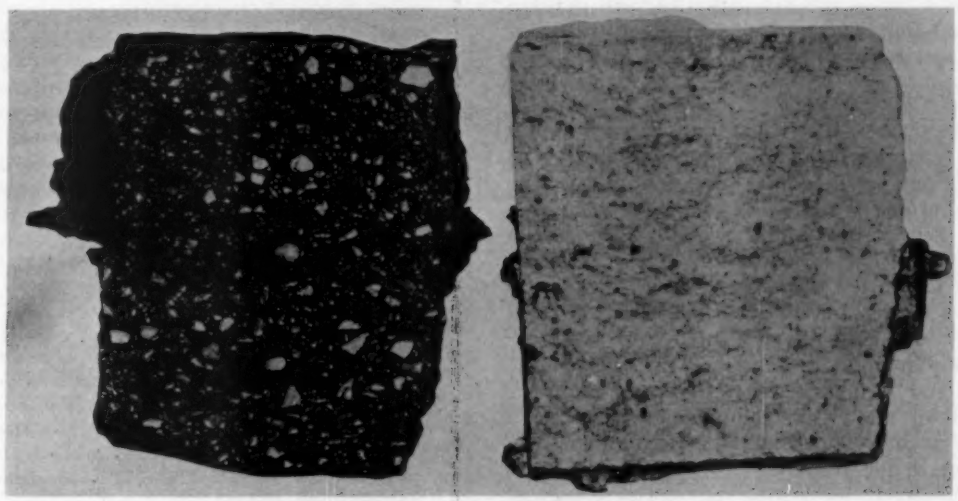


Fig. 6 — Same as Fig. 5, after 3 weeks.

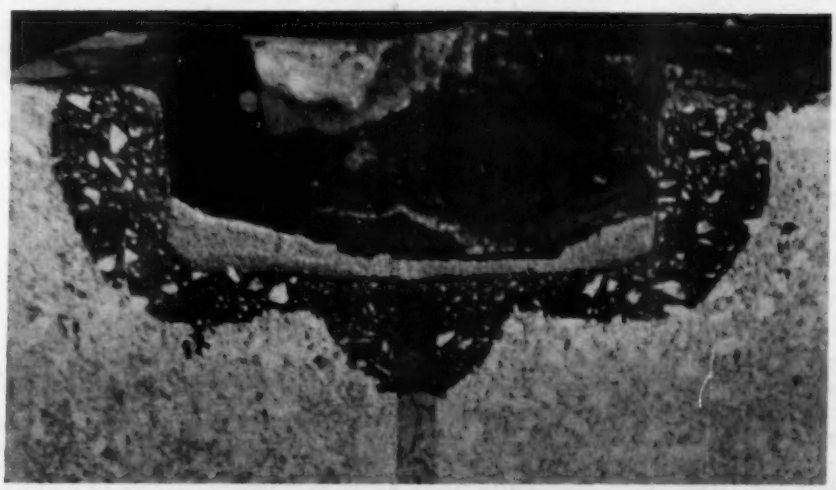


Fig. 7 — Attack of molten aluminum on mortar containing silicate of soda.



Fig. 8 — Resistance to attack by molten aluminum of phosphate-bonded high-alumina mortar.

applied by the air-placement method. Phosphate-bonded high-alumina ramming mixes are also being used with satisfactory results. This particular type of ramming mix, when properly installed, will develop good strength that is maintained throughout the entire range of operating temperatures.

### *Furnace Lining Impregnation*

Impregnation of aluminum melting furnace linings with a mixture of 80 per cent sodium chloride (NaCl) and 20 per cent cryolite ($Na_3AlF_6$) apparently originated in Europe, possibly due to the limited availability of refractories which would retard silicon pickup. Impregnation is accomplished by melting the mixture of salt in the furnace and slowly raising the level of the molten salt by additions of hot or cold metal. The melting point of this mixture is 1463 F (795 C).[3] The molten salt penetrates the brick, filling the pores, and to a degree is effective in retarding penetration by molten metal. Although there are a number of such installations on this continent, cost information is not available.

Various materials for coating refractories to prevent penetration have been proposed and tried over the years. An early reference suggested "aluminum bronze varnish," ordinary graphite blacking as used in the foundry and finely ground dead-burned magnesite.[4] Another reference proposed various mixtures such as calcined kyanite and calcium carbonate in 2 to 1 ratio, ball milled with sodium silicate.[5] More recently, a coating compound composed principally of boric acid has been developed.[6] Coating the brick with phosphate-bonded high-alumina mortar is also recommended. Coating materials do have some immediate effect in retarding penetration, but this is generally of only a temporary nature.

## SUMMARY

The refractory most suitable for lining the hearths of aluminum melting furnaces should possess several superior properties. Resistance to penetration and reaction by molten aluminum and its numerous alloys is most important. This property retards contamination, reduces metal loss in the refractory, makes for easier cleaning and obviously extends service life of the furnace. Almost equally important is sufficient strength at operating temperatures to withstand mechanical abuse.

Other desirable characteristics include resistance to thermal shock, adequate refractoriness, dimensional accuracy and, of course, cost commensurate with service performance. Several refractories developed specifically for this application closely approximate the optimum in this combination of desirable properties. However, because of the many variables involved in melting aluminum, selection of the refractories for each individual unit must be made on the basis of operating conditions and economic considerations.

Average service life for a particular refractory is difficult to evaluate, since so much depends on melting rates, alloys melted, size and design of furnaces and other pertinent factors. Probably the best estimate would be on the basis of pounds of aluminum melted, and some authorities feel that 35 million to 50 million lb normally is satisfactory furnace life.

## CONCLUSION

Research on refractories for aluminum melting is continuing, and new products still in the development stage or in trial installations should contribute to increased economies. As new products are developed, the cooperation of operators in arranging trial installations is essential, since laboratory equipment can simulate actual furnace conditions only to a limited extent.

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# CAST NICKEL CONTAINING ALUMINUM BRONZE PROPERTIES AND MICROSTRUCTURE

by E. Belkin


## ABSTRACT

A study has been made of the properties and microstructure of aluminum bronze alloys containing 9.5 to 11.5 per cent aluminum, 3 to 6 per cent iron and 3 to 6 per cent nickel.

Mechanical properties and microstructures were examined for as-cast alloys to determine the separate effects of aluminum, iron and nickel It was found that aluminum has a greater effect on properties and structure than do the other two elements. Nickel causes an increase in ductility at the highest aluminum content by reducing the amount of beta phase and increasing the amounts of alpha phase. Iron adds to the hardness, but variations in iron have no other significant effect on the properties measured. No attempt was made to measure the effect of iron on grain size.

Microstructure, hardness and tensile properties were determined after various heat treatments of one alloy. The effect of quenching and air cooling from temperatures ranging from 1000 C (1832 F) to 500 C (932 F) was studied by hardness measurements and by microscopic examination. Tensile properties were studied after various quenching and tempering combinations, which included quenching temperatures of 900 C (1652 F), 850 C (1562 F) and 800 C (1472 F); tempering temperatures of 700 C (1292 F) and 600 C (1112 F); and air cooling and quenching from the tempering temperatures.


## INTRODUCTION

Aluminum bronze alloys are one of the two groups of copper-base casting alloys which are commonly used for high-strength structural applications. Of these, the alloy used for the most severe structural applications is the nickel containing variety identified as A.S.T.M. B148-52, Alloy 9D. With increasingly severe service applications there is the corresponding increase in the demand for higher strength materials, and, thus, the aluminum bronze alloys can be expected to find wider application.

Because published data are scarce concerning the effect on mechanical properties of variations in composition and of various heat treatments for the cast nickel containing aluminum bronze alloys, an investigation is being conducted related to the highest strength alloys. Without such information, the foundryman and designer are at a loss to choose a composition which best fits the application, or to explain some of the service or processing difficulties relating to composition or heat treatment. There is, however, considerable information regarding wrought aluminum bronze alloys. Especially noteworthy is the work by Cook, et al.[1]

Results are described on the hardness, tensile properties and microstructure of as-cast and heat-treated material within the composition ranges of 9 to 12 per cent aluminum and 3 to 6 per cent each of iron and nickel. Most of the data are based on small heats made under laboratory conditions, and are thus preliminary to additional studies which should be made with material produced in a commercial foundry.

## PROCEDURES AND RESULTS

All heats were made using O.F.H.C. copper in combination with master alloys of copper-aluminum, copper-nickel, copper-iron and copper-manganese. All the heats except one were melted in a high frequency induction furnace of 60 lb capacity and tapped into two core sand Y-block molds. In Fig. 1 are shown a typical mold and casting used for these studies. The melting procedure used for the laboratory heats is:

1. 50-50 copper-aluminum, 90-10 copper-iron, 50-50 copper-nickel and O.F.H.C. copper charged into furnace.
2. Molten at about 1200 C (2192 F).
3. 70-30 copper-manganese added, stirred.
4. Held for 5 min.
5. Power turned off, metal frozen to reduce gas content.
6. Remelted.
7. Added 0.1 per cent cryolite, stirred, skimmed.
8. Tapped into tundish.
9. Metal held till temperature dropped to approximately 1175 C (2147 F).

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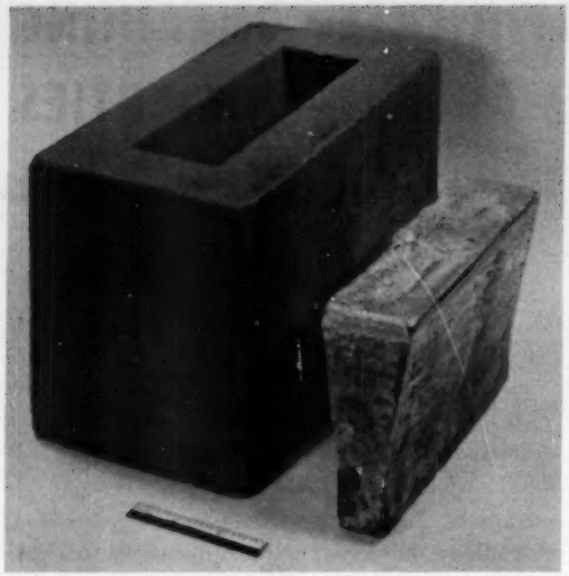


Fig. 1 — Mold and ingot made in the laboratory heats.

10. Removed stopper and poured into two Y-bar molds in succession. Each ingot was allowed to cool to about room temperature before removal from the mold.

The first phase consisted of making a number of heats to explore the effect of variations in aluminum, iron and nickel on as-cast tensile properties.

Two 7/8-in. thick slices were cut longitudinally from each ingot and later machined into 0.505 in. diameter tensile bars. The ends of each slice were sectioned for microscopic examination and chemical analysis.

### *Mechanical Properties of As-Cast Alloys*

Table 1 lists the composition, as-cast room temperature tensile properties and hardness for each of the alloys studied in this phase. Each value represents the average of two tests. The data for the sand cast material are arranged so that each group represents one aluminum level. Within each group the alloys are arranged in order of increasing iron content.

It can be seen that the major effects of increasing aluminum content from 9.5 to 10 per cent are to reduce the percentage elongation and increase the hardness. Lesser effects are noted in the increase in yield strength and ultimate strength. With an increase to 10.5 per cent aluminum, the decrease in elongation and increase in hardness is less pronounced for the group, but the increase in yield strength and ultimate strength is similar to that noted for an increase from 9.5-10 per cent aluminum. The pattern of increase in yield strength and hardness continues with increase in aluminum to 11.5 per cent, but elongation

TABLE 1 — COMPOSITION AND MECHANICAL PROPERTIES OF SAND CAST ALLOYS IN AS-CAST CONDITION

| | Intended Analysis, % | | | | Actual analysis, % | | | | | | | |
|---|---|---|---|---|---|---|---|---|---|---|---|---|
| Heat No. | Al | Fe | Ni | Mn | Al | Fe | Ni | Mn | Yield Str., 0.2% Offset, psi | Ultimate Str., psi | Elong., % in 2 in. | Bhn, 3000 Kg |
| 1 | 9.5 | 3 | 3 | 0.5 | 9.37 | 3.16 | 3.10 | 0.60 | 34,500 | 88,000 | 25 | 143 |
| 2 | 9.5 | 3 | 5 | 0.5 | 9.4 | 2.74 | 5.23 | 0.55 | 37,000 | 86,000 | 20 | 149 |
| 3 | 9.5 | 4 | 4 | 0.5 | 9.4 | 4.14 | 3.80 | 0.49 | 32,500 | 92,000 | 23 | 152 |
| 4 | 9.5 | 4.5 | 3.5 | 0.5 | 9.4 | 4.61 | 3.68 | 0.54 | 34,500 | 94,500 | 25 | 156 |
| 5 | 9.5 | 5 | 5 | 0.5 | 9.42 | 4.80 | 5.11 | 0.60 | 40,000 | 93,500 | 19 | 163 |
| 6 | 10 | 3 | 6 | 0.5 | 10.2 | 2.86 | 6.26 | 0.54 | 46,000 | 91,000 | 10 | 170 |
| 7 | 10 | 4 | 4 | 0.5 | 9.8 | 4.25 | 3.88 | 0.31 | 37,000 | 93,000 | 15 | 167 |
| 8 | 10 | 5 | 3 | 0.5 | 10.2 | 5.15 | 3.27 | 0.54 | 40,000 | 94,500 | 11 | 179 |
| 9 | 10 | 5 | 5 | 0.5 | 9.9 | 4.76 | 5.08 | 0.50 | 38,000 | 95,000 | 15 | 170 |
| 10 | 10 | 6 | 6 | 0.5 | 9.8 | 6.05 | 6.03 | 0.48 | 40,000 | 91,000 | 10 | 181 |
| 11 | 10.5 | 3 | 3 | 0.5 | 10.3 | 2.84 | 3.07 | 0.55 | 36,000 | 94,000 | 11 | 170 |
| 12 | 10.5 | 4.5 | 6 | 0.5 | 10.4 | 4.30 | 5.88 | 0.54 | 42,000 | 100,000 | 14 | 187 |
| 13 | 10.5 | 5 | 5 | 0.5 | 10.7 | 4.97 | 4.97 | 0.45 | 45,500 | 101,000 | 9 | 197 |
| 14 | 10.5 | 5.5 | 5.5 | 0.5 | 10.4 | 5.69 | 5.39 | 0.52 | 41,500 | 102,000 | 15 | 187 |
| 15 | 10.5 | 6 | 6 | 0.5 | 10.5 | 5.80 | 6.13 | 0.51 | 45,000 | 98,000 | 8 | 193 |
| 16 | 11 | 3 | 3 | 0.5 | 11.0 | 3.06 | 3.15 | 0.60 | 50,000 | 100,000 | 6 | 207 |
| 17 | 11 | 3 | 5 | 0.5 | 10.8 | 2.89 | 5.26 | 0.55 | 51,000 | 106,000 | 10 | 197 |
| 18 | 11 | 3 | 6 | 0.5 | 11.1 | 2.86 | 6.10 | 0.55 | 49,000 | 100,500 | 9 | 201 |
| 19 | 11 | 4.5 | 3 | 0.5 | 10.9 | 4.47 | 3.17 | 0.55 | 45,500 | 98,500 | 7 | 207 |
| 20 | 11 | 5 | 5 | 0.5 | 11.0 | 4.80 | 5.30 | 0.60 | 55,000 | 107,500 | 10 | 197 |
| 21 | 11 | 5.5 | 5 | 0.5 | 11.2 | 5.63 | 5.05 | 0.54 | 48,000 | 107,000 | 10 | 207 |
| 22 | 11.5 | 3 | 3 | 0.5 | 11.6 | 3.11 | 3.09 | 0.63 | 53,000 | 103,000 | 4 | 235 |
| 23 | 11.5 | 3 | 5 | 0.5 | 11.6 | 2.85 | 5.35 | 0.54 | 52,000 | 109,000 | 8 | 217 |
| 24 | 11.5 | 5 | 5 | 0.5 | 11.4 | 4.54 | 5.31 | 0.61 | 59,000 | 108,000 | 7 | 217 |
| 25 | 11.5 | 5.5 | 5.5 | 0.5 | 11.8 | 5.65 | 5.58 | 0.54 | 53,500 | 111,500 | 8 | 223 |

and ultimate strength show almost no change upon increasing the aluminum content from 11 to 11.5 per cent.

To study the effect of aluminum more closely, it is desirable to compare compositions in which only the aluminum content varies. Tensile and hardness data in the form of curves are shown in Fig. 2 for alloys which have approximately 3 per cent iron and 3 per cent nickel but different amounts of aluminum, and for a second group with 5 per cent each iron and nickel. Depicted in this fashion, it can be seen that changes in aluminum content between 9.5 and 11.5 per cent strongly affect tensile and hardness properties. The property changes are similar in degree for both the 3 per cent nickel-iron alloys and the 5 per cent nickel-iron alloys.

There is no significant influence on properties as a result of increasing iron from 3 to 6 per cent at any of the aluminum values, except the slight increase in hardness at 9.5 per cent aluminum. Similarly, variations in nickel between 3 and 6 per cent have little effect on properties, except that at 11 and 11.5 per cent aluminum the elongation increases when nickel is increased. At the highest aluminum level, an increase in nickel from 3 to 5.5 per cent causes the hardness to drop slightly.

The data in Table 2 and Fig. 2 confirm that aluminum has a greater effect on properties than do nickel or iron.

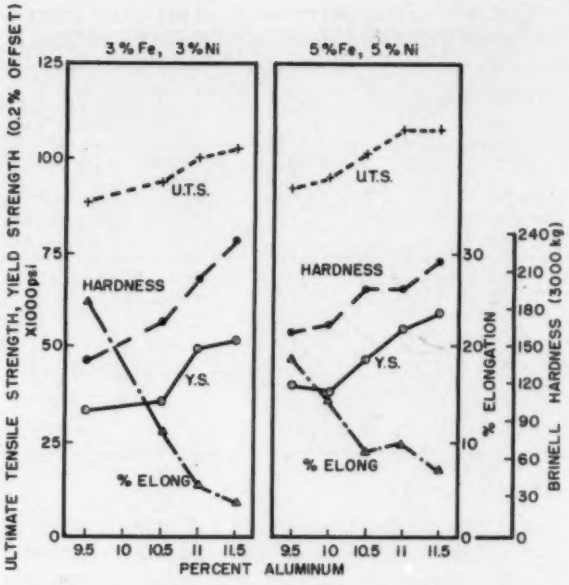



Fig. 2 — Relationship between aluminum content and mechanical properties of cast alloys.

### *Microstructures of Sand Cast As-Cast Alloys*

Increases in aluminum content result in an increase in the amount of hard, brittle eutectoid phase and beta phase at the expense of the soft, ductile alpha phase. Photomicrographs in Fig. 3 show microstructures of four heats, 1, 11, 16 and 22 in which the iron and nickel contents are approximately 3 per cent. Tensile properties and compositions of the respective heats are shown adjacent to each photomicrograph.

TABLE 2 — MECHANICAL PROPERTIES OF SAND CAST ALLOYS IN HEAT TREATED CONDITION

| Heat No. | Intended Analysis, % Al | Fe | Ni | Mn | Yield Str., 0.2% Offset, psi | Ultimate Str., psi | Elong., % in 2 in. | Bhn, 3000 Kg |
|---|---|---|---|---|---|---|---|---|
| 1 | 9.5 | 3 | 3 | 0.5 | 40,000 | 93,000 | 25 | 163 |
| 2 | 9.5 | 3 | 5 | 0.5 | 49,000 | 101,000 | 14 | 197 |
| 3 | 9.5 | 4 | 4 | 0.5 | 44,500 | 97,000 | 17 | 179 |
| 4 | 9.5 | 4.5 | 3.5 | 0.5 | 43,000 | 99,000 | 23 | 179 |
| 5 | 9.5 | 5 | 5 | 0.5 | 44,000 | 96,500 | 13 | 185 |
| 6 | 10 | 3 | 6 | 0.5 | 61,000 | 111,000 | 9 | 229 |
| 8 | 10 | 5 | 3 | 0.5 | 61,500 | 103,000 | 6 | 201 |
| 9 | 10 | 5 | 5 | 0.5 | 52,000 | 102,000 | 10 | 201 |
| 12 | 10.5 | 4.5 | 6 | 0.5 | 55,000 | 108,500 | 11 | 217 |
| 14 | 10.5 | 5.5 | 5.5 | 0.5 | 56,500 | 107,000 | 10 | 209 |
| 16 | 11 | 3 | 3 | 0.5 | 57,000 | 111,000 | 8 | 220 |
| 17 | 11 | 3 | 5 | 0.5 | 79,500 | 118,500 | 8 | 241 |
| 18 | 11 | 3 | 6 | 0.5 | 76,500 | 116,500 | 6 | 241 |
| 19 | 11 | 4.5 | 3 | 0.5 | 63,000 | 116,000 | 10 | 229 |
| 20 | 11 | 5 | 5 | 0.5 | 82,500 | 119,600 | 7 | 246 |
| 21 | 11 | 5.5 | 5 | 0.5 | 73,500 | 116,000 | 8 | 229 |
| 22 | 11.5 | 3 | 3 | 0.5 | 51,500 | 104,500 | 5 | 229 |
| 23 | 11.5 | 3 | 5 | 0.5 | 86,500 | 121,000 | 6 | 251 |
| 24 | 11.5 | 5 | 5 | 0.5 | 83,500 | 121,000 | 6 | 251 |
| 25 | 11.5 | 5.5 | 5.5 | 0.5 | 88,000 | 124,500 | 6 | 255 |

Treatment: 850 C (1562 F) — one hr, W.Q., 700 C (1292 F) — 2 hr, A.C.

TABLE 3 — COMPOSITION OF SAND CAST KEEL BARS USED FOR HEAT TREATMENT STUDIES

| Heat No. | Chemical Analysis, % Cu | Al | Fe | Ni | Mn |
|---|---|---|---|---|---|
| 61 | 80.96 | 10.6 | 3.92 | 4.08 | 0.56 |

| Qualitative Spectrographic Analysis | Estimated Concentration, % |
|---|---|
| Pb | 0.03 |
| Ag | 0.015 |
| Sn | 0.015 |
| Mg | 0.003 |
| Si | 0.003 |
| Co | 0.003 |

The photomicrographs in Fig. 4 of alloys 5, 9, 13, 20 and 24 show the effect of increasing the aluminum content from 9.5 to 11.5 per cent when the iron and nickel contents are approximately 5 per cent. In these structures, the result of higher aluminum content can be seen as an increase in the amount of eutectoid, a decrease in the amount of alpha and an increase in the amount of beta. The increase in eutectoid content accounts for the decrease in ductility and the increase in hardness.

As can be seen by comparing the photomicrographs in Figs. 3 and 4, the effect of increasing the nickel content from 3 to 5 per cent is to cause the eutectoid to be less massive, the alpha phase to be present in smaller, less rounded form and to reduce the amount of beta phase. In addition, these two groups of photomicrographs reveal that increasing the iron content from 3 to 5 per cent causes greater amounts of the spheroidal aluminum-rich iron particles, called kappa phase, to be present.[2-3] For equal variations, aluminum is obviously more effective than nickel or iron in causing changes in properties and microstructures.

TABLE 4 — HARDNESS OF AN ALUMINUM BRONZE ALLOY* AS-QUENCHED AND AS-AIR COOLED FROM ELEVATED TEMPERATURES

| Heat Treatment Temp., C (F) | Bhn, 3000 Kg, Cooling Rate: Water Quench | Air Cooled |
|---|---|---|
| 1000 (1832) | 309** | 241 |
| 950 (1742) | 340** | 232 |
| 900 (1652) | 313** | 232 |
| 850 (1562) | 323** | 226 |
| 800 (1472) | 255 | 217 |
| 750 (1382) | 214 | 197 |
| 600 (1112) | — | 181 |
| 500 (932) | — | 189 |

NOTES:

*Heat 61 — see Table 3 for composition.

**Converted from D.P.H. (10 Kg) numbers. As Received hardness — Bhn 185

### *Mechanical Properties of Heat Treated Alloys*

Sections of the alloys cast in sand molds were heated to 850 C (1562 F) for one hr quenched into water, heated to 700 C (1292 F) for 2 hr and air cooled to room temperature. The temperatures were maintained within ± 5 degrees C. These bars were then machined into standard 0.505 in. diameter tensile bars and tested at room temperature, with the results shown in Table 2. Although the tensile results for the heat-treated alloys are discussed in this part of the report, the heat treatment was selected on the basis of studies described in a later portion. Not all of the tensile results are included because examination of the fractured bars revealed that some bars were unsound.

To evaluate the result of heat treatment of these alloys, it is necessary to compare the tensile results in Table 2 with the as-cast properties listed in Table 1 for the same alloys. At 9.5 per cent aluminum, the heat-treated alloys show a relatively small increase in yield strength, ultimate strength and hardness accompanied by a slight decrease in elongation. At 10 per cent aluminum, the increase in strength and hardness is somewhat greater and the loss in ductility is more severe. For the 10.5 to 11 per cent aluminum alloys, heat treatment causes the largest increase in yield strength compared to the other aluminum ranges. Ultimate strength and hardness also increase on heat treating, and the elongation decreases in about the same degree as in the lower aluminum range alloys.

A comparison of alloys in which the percentage of aluminum and iron is constant, but the nickel differs, shows that at 9.5, 11 and 11.5 per cent aluminum the effect of higher nickel is to cause a considerable increase in yield strength in the heat-treated alloys. This increase is accompanied by loss in ductility at 9.5 per cent aluminum, but no change at other aluminum levels. Note the properties listed in Table 2 for alloys 1 and 2 with 3 and 5 per cent nickel, respectively; alloys 16, 17 and 18 with 3, 5 and 6 per cent nickel, respectively; and alloys 22 and 23 with 3 and 5 per cent nickel, respectively. The lack of compositions with precisely the same aluminum and iron content prevents the analysis of the effect of nickel for the 10 and 10.5 per cent aluminum ranges.

### *Various Cooling Rates From Quenching Temperatures Effect*

To study the effect of various heat treatments, 12 keel bars made from one 300 lb heat were obtained from a commercial foundry. The composition of this heat, assigned number 61, is shown in Table 3. Several bars were cut into slices approximately ½-in. by one in. by one in. for heat treatment. After one hr at one of the following temperatures: 1000 C (1832 F), 950 C (1742 F), 900 C (1652 F), 850 C (1562 F), 800 C (1472 F), 750 C (1382 F), 600 C (1112 F) and 500 C (932 F), one piece from each temperature was cooled in air to room temperature. In addition, one piece from each temperature, except the last two, was quenched into water. Hardness measurements made

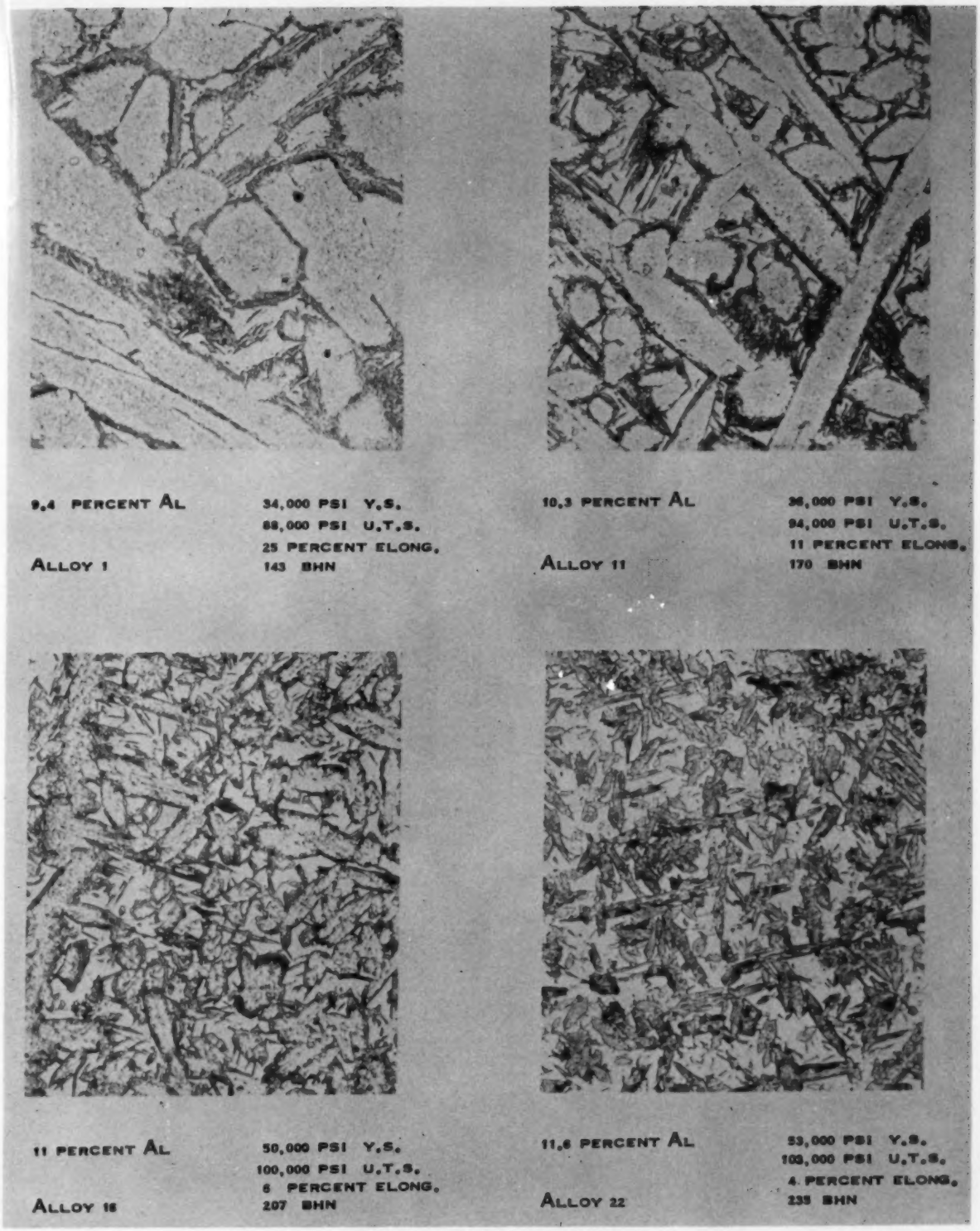



Fig. 3 — Microstructures and associated properties related to aluminum content for as-cast aluminum bronzes. Nominal composition for each alloy — 3 per cent Fe, 3 per cent Ni, 0.5 per cent Mn, Al as shown, balance Cu. One per cent chromic acid etch. 250 ×.

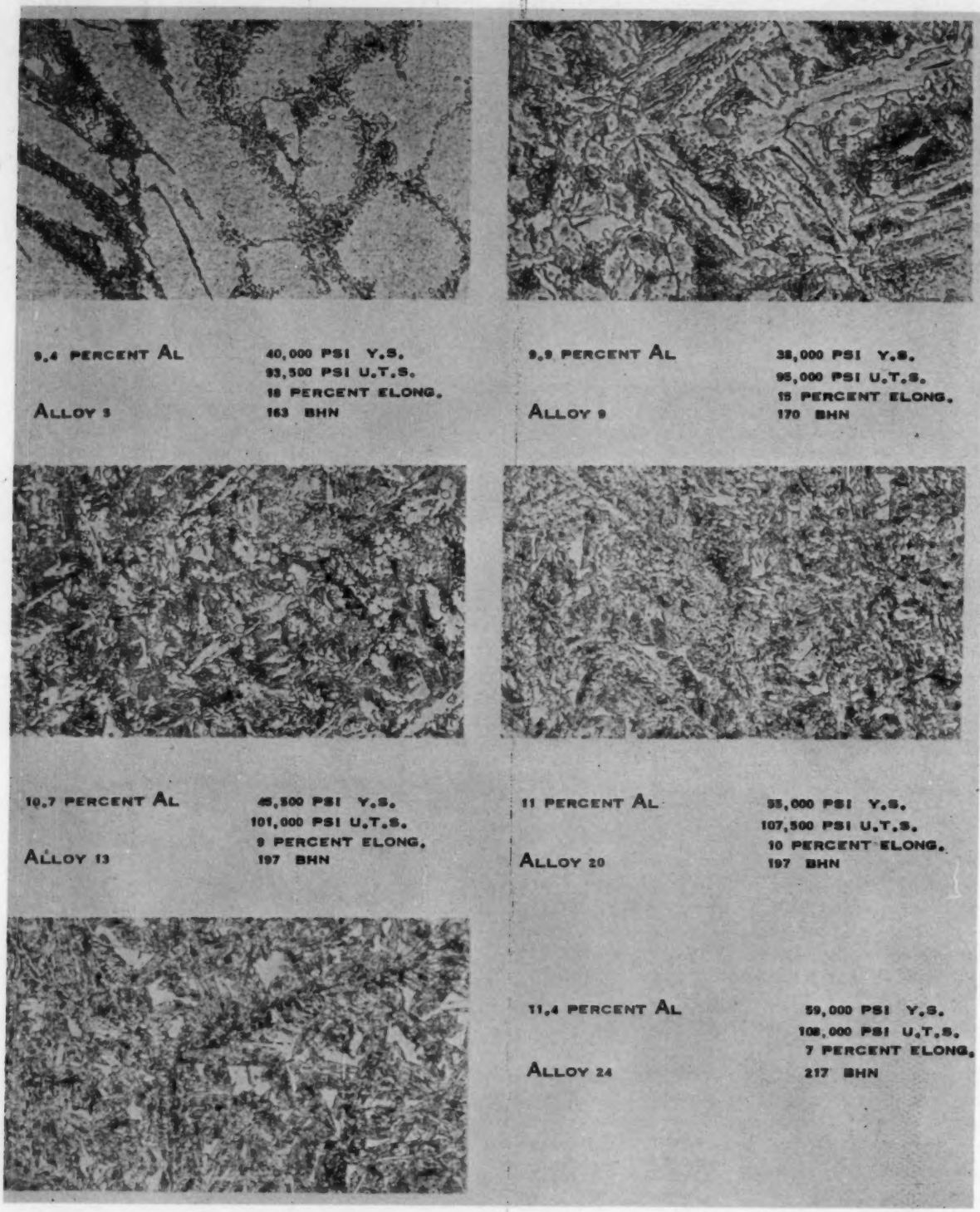



Fig. 4 — Microstructures and properties related to aluminum content for as-cast aluminum bronzes. Nominal composition for each alloy — 5 per cent Fe, 5 per cent Ni, 0.5 per cent Mn, Al as shown, balance Cu. One per cent chromic acid etch. 250 ×.

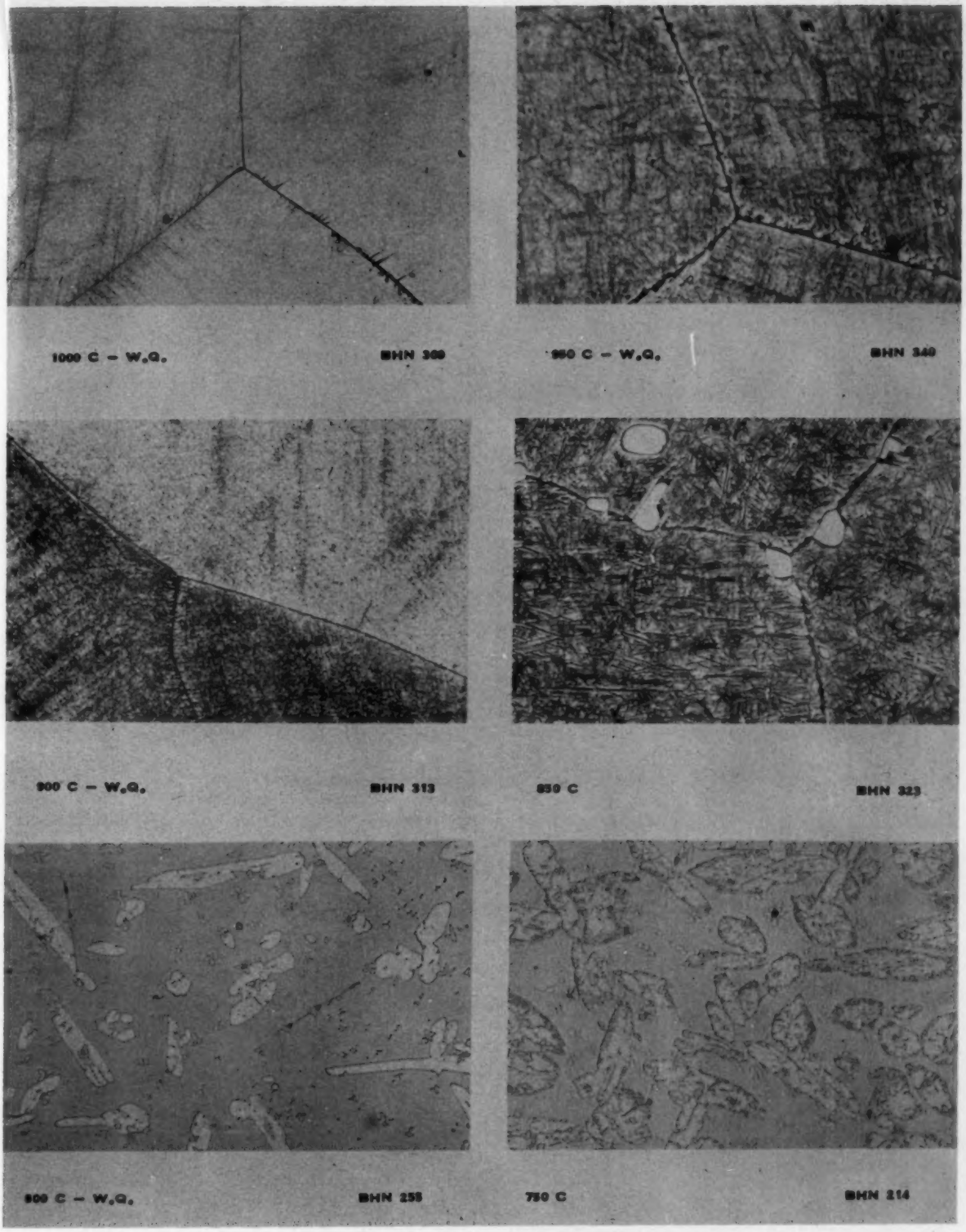



Fig. 5 — Microstructures of cast aluminum bronze from heat 61 on quenching into water from indicated temperatures. One per cent chromic acid etch. 250 ×.

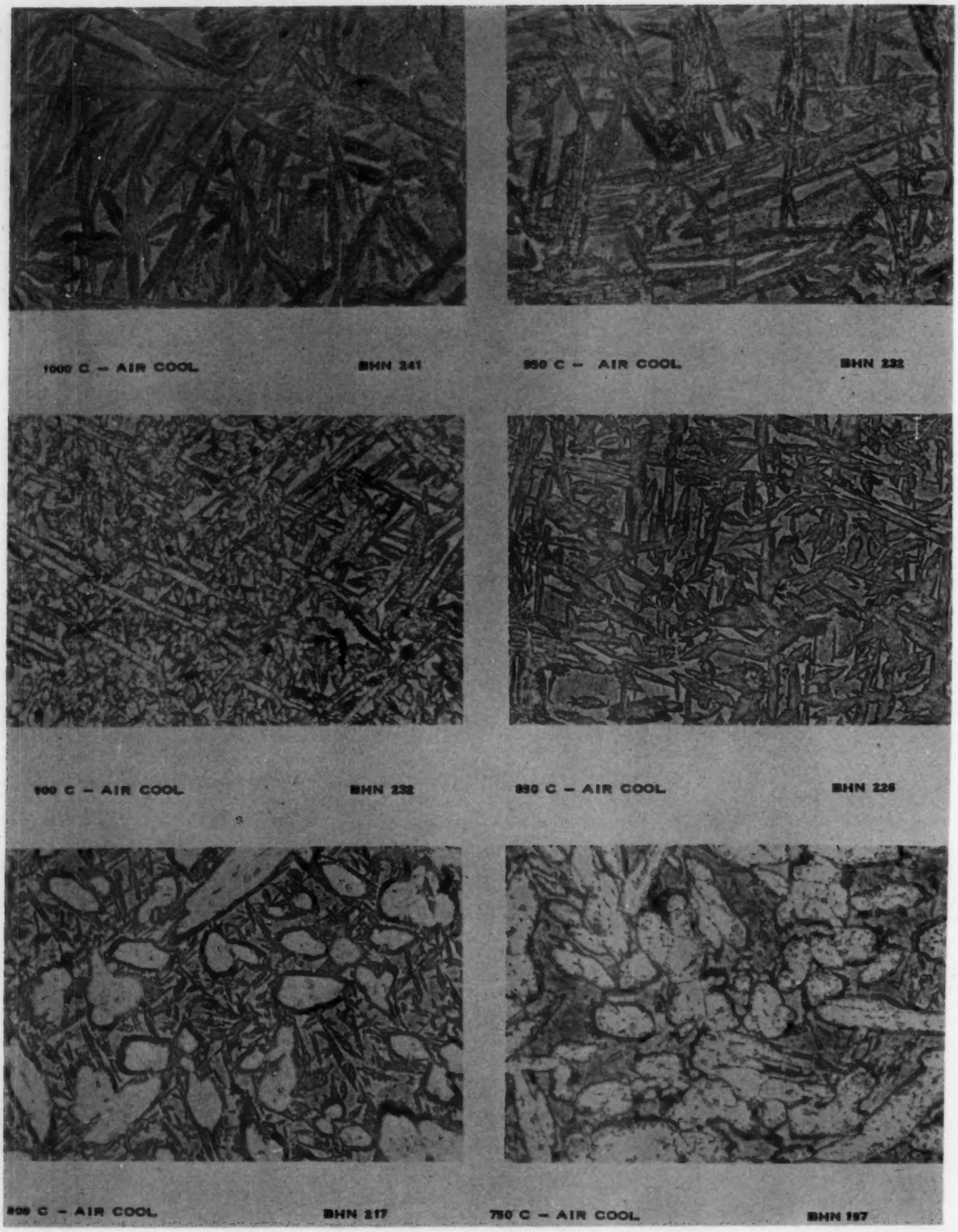



Fig. 6 — Microstructures of cast aluminum bronze from heat 61 on cooling in air from indicated temperatures. One per cent chromic acid etch. 250 ×.

on each piece after quenching or cooling in air are listed in Table 4.

Samples quenched from 850 C (1562 F), 900 C (1652 F) or 950 C (1742 F) have approximately the same hardness, whereas the as-quenched hardness is lower when the quenching temperature is reduced to 800 C (1472 F) and the hardness is further reduced on quenching from 750 C (1382 F). Air-cooled samples attained maximum hardness upon cooling from 1000 C (1832 F). With lower heating temperatures, the hardness after cooling to room temperature was less, except that the piece cooled from 500 C (932 F) was harder than the 600 C (1112 F) sample.

In every case, the quenched samples were harder than those air cooled from the same temperature. This difference is an expected result of the formation of larger amounts of the martensitic phase on rapid quenching.[4] An explanation of the increase in hardness at 500 C (932 F) is that some of the beta phase in the as-cast material decomposed to the harder eutectoid phase.

### *Microstructures of Material As-Quenched and As-Air-Cooled From Elevated Temperatures*

Photomicrographs in Fig. 5 trace the transformation which takes place upon quenching from temperatures ranging from 1000 C (1832 F) to 750 C (1382 F). On quenching from the highest temperature, the structure is almost completely martensitic. The 850 C (1562 F) sample contains alpha phase in the form of needles, as spheroidal particles at the grain boundaries and as spheroids within the grains. The structure of the 800 C (1472 F) material consists of a beta matrix and elongated particles of alpha.

In this structure, the aluminum-rich iron particles (kappa phase) are clearly visible in the form of small rosettes or spheroids. At still lower quenching temperature of 750 C (1382 F) some eutectoid has formed, and is seen as the dark shading around the edge of the alpha particles. It is noticeable also that the alpha particles are less elongated. The matrix phase in this structure is beta.

The microstructures, shown in Figs. 6 and 7 of the samples air cooled from 1000 C (1832 F), 950 C (1742 F), 900 C (1652 F) and 850 C (1562 F), are quite similar in that the matrix is a mixture of beta and elongated particles of alpha. There is fine eutectoid within the alpha particles for each of the above microstructures but there is somewhat more alpha at the lowest temperature 850 C (1562 F). After cooling in air from 800 C (1472 F) the alpha phase appears in a more rounded form, the eutectoid is slightly coarser and some eutectoid is seen scattered in the beta matrix.

In this structure, the kappa is visible as rosettes and spheroids. The samples cooled from 750 C (1382 F) and 600 C (1112 F) are quite similar, except that the lower temperature structure has more beta, the remainder of structure being composed of alpha and eutectoid. Kappa phase is visible in both structures. As a matter of reference, the microstructure of the material in the as-received condition is shown.

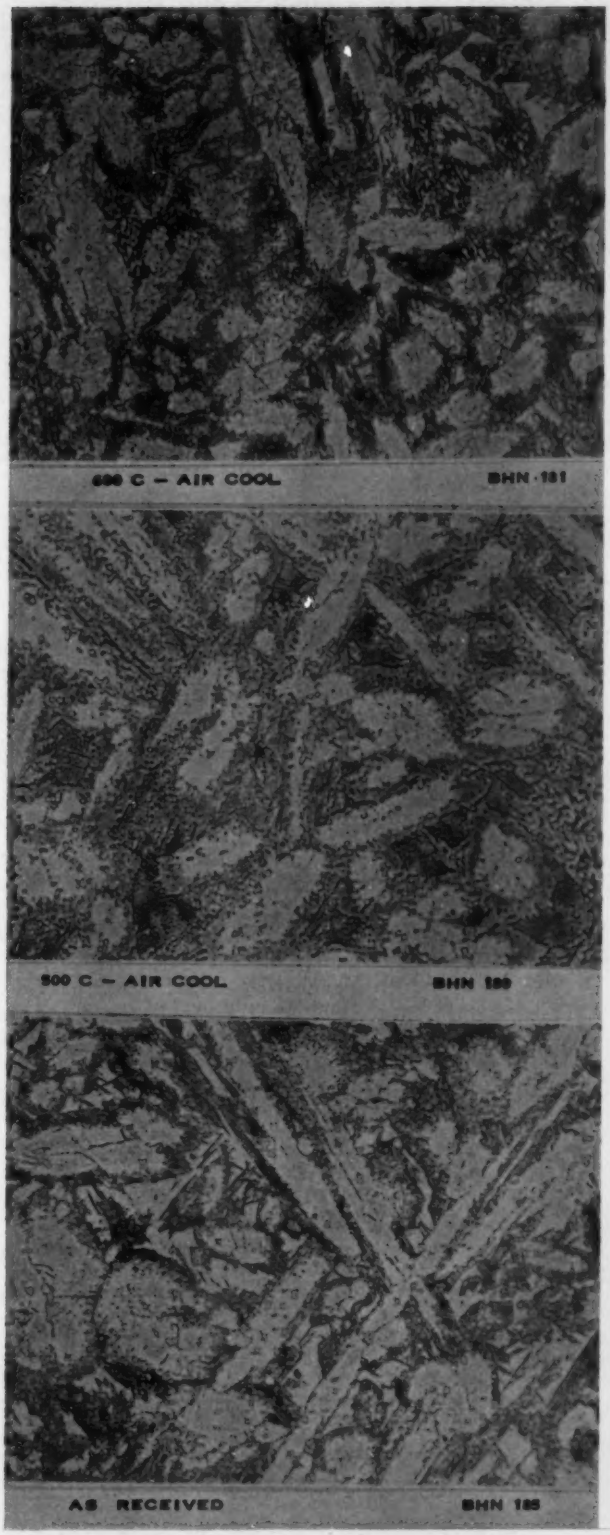



Fig. 7 — Microstructures of cast aluminum bronze of heat 61 on cooling in air from indicated temperatures. Alloy in as-cast condition is shown for comparison. One per cent chronic acid etch. 250 ×.

TABLE 5 — HARDNESS OF AN ALUMINUM BRONZE ALLOY* AFTER QUENCHING AND TEMPERING

| Heat Treatment | | Bhn, 3000 Kg, After Tempering | |
|---|---|---|---|
| Quenching Temp., C (F) | Tempering Temp., C (F) | Water Quenched From Tempering Temp. | Air Cooled From Tempering Temp. |
| 900 (1652) | 400 (752) | 380** | |
| (1652) | (752) | | 380 |
| (1652) | 500 (932) | 302 | |
| (1652) | (932) | | 302 |
| (1652) | 600 (1112) | 255 | |
| (1652) | (1112) | | 255 |
| (1652) | 700 (1292) | 229 | |
| (1652) | (1292) | | 217 |
| 800 (1472) | 400 (752) | 269 | |
| (1472) | (752) | | 265 |
| (1472) | 500 (932) | 229 | |
| (1472) | (932) | | 229 |
| (1472) | 600 (1112) | 201 | |
| (1472) | (1112) | | 197 |
| (1472) | 700 (1292) | 197 | |
| (1472) | (1292) | | 189 |

NOTES: *Heat 61 — See Table 3 for composition.
**Converted from D.P.H. 10 Kg numbers.

This structure is composed of fine eutectoid and alpha with kappa phase scattered throughout.

### *Tempering After Quenching Effect*

Based on the hardness data for as-quenched samples, two quenching temperatures, 900 C (1652 F) and 800 C (1472 F), were selected for quenching and tempering experiments. Pieces ½-in. by one in. by one in., cut from heat 61 were heated for one hr at 900 C (1652 F). An equal number of pieces were heated for one hour at 800 C (1472 F). After heating, all of the samples were quenched into water. Two pieces from each quenching temperature were then tempered for one hr at 400 C (752 F), two at 500 C (932 F), two at 600 C (1112 F) and two at 700 C (1292 F). After tempering was completed, one of the samples from each tempering temperature was quenched into water and the other piece cooled in air.

TABLE 6 — HEAT TREATMENTS USED FOR CAST ALUMINUM BRONZE ALLOY*

| Quenching Treatment | Tempering Treatment |
|---|---|
| 900 C (1652 F)—one hr—W.Q. | 600 C (1112 F)—2 hr—air cooled |
| 900 C (1652 F)—one hr—W.Q. | 700 C (1292 F)—2 hr—air cooled |
| 900 C (1652 F)—one hr—W.Q. | 700 C (1292 F)—2 hr—W.Q. |
| 850 C (1562 F)—one hr—W.Q. | 600 C (1112 F)—2 hr—air cooled |
| 850 C (1562 F)—one hr—W.Q. | 700 C (1292 F)—2 hr—air cooled |
| 850 C (1562 F)—one hr—W.Q. | 700 C (1292 F)—2 hr—W.Q. |
| 800 C (1472 F)—one hr—W.Q. | 600 C (1112 F)—2 hr—air cooled |
| 800 C (1472 F)—one hr—W.Q. | 700 C (1292 F)—2 hr—air cooled |
| 800 C (1472 F)—one hr—W.Q. | 700 C (1292 F)—2 hr—W.Q. |

NOTE: *Heat 61 — see Table 3 for composition.
W.Q. — Water Quench (tap water)

Hardness measurements made on each piece are listed in Table 5. Maximum hardness resulted from tempering at 400 C (752 F) for both 800 C (1472 F) and 900 C (1652 F) quenching temperatures. With increasing tempering temperatures, the pieces showed lower hardness values. For the same tempering temperatures, the pieces quenched from 900 C (1652 F) had higher hardness. For both quenching temperatures, only a slight difference was noted between the pieces which were air cooled and those quenched from the tempering temperatures.

### *Mechanical Properties After Various Heat Treatments*

To extend the information on mechanical properties after quenching and tempering, tensile properties and hardness were determined after various heat treatments on keel bars of heat 61. A description of the heat treatments is listed in Table 6, and the tensile and hardness values are shown in the form of curves in Fig. 8. Tempering was conducted only at the higher temperatures because it was believed that the extreme hardness which resulted on tempering at 400 C (752 F) and 500 C (932 F) would be accompanied by brittleness.

Consistent with the hardness data after quenching, the pieces quenched from the highest temperature and tempered at the lowest temperature had the highest yield strength, highest ultimate tensile strength and highest hardness, but the lowest percentage elongation. This same pattern applied for each tempering temperature; that is, the strength is greatest but the ductility lowest for the highest quenching temperature. There is some overlapping in values, however, if a comparison is made between bars which were quenched and tempered at different temperatures. No significant difference was noted between the material quenched or air cooled from the tempering temperature.

Based on the tensile test results shown in Fig. 8, it was decided to use the following heat treatment for the alloys described in Table 1 and discussed earlier: 850 C (1562 F) for one hr, water quench, 700 C (1292 F) for 2 hr, air cool. This treatment seems to provide the combination of high yield strength and elongation. The choice of air cooling from the tempering temperature was made on the basis that air cooling would produce less residual stress than quenching.

## SUMMARY

The effect of mechanical properties and microstructure was explored for three elements—aluminum, iron and nickel. Of the three elements, aluminum is the most effective in altering microstructure and properties. With increase in aluminum from 9.5 to 10 per cent, there is a marked decrease in elongation, but a sizable increase in strength and hardness.

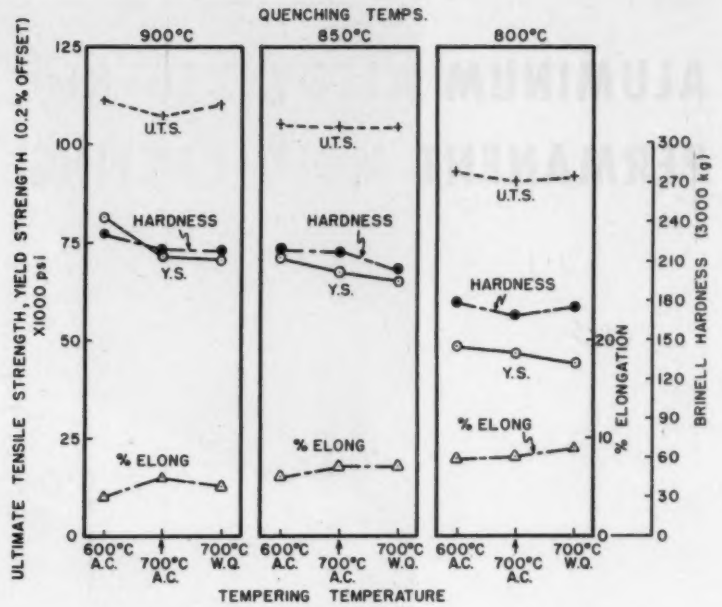



Fig. 8 — Mechanical properties after various heat treatments.

Further increases in aluminum causes the strength and hardness to decrease, but the reduction in ductility is not as great. Increase in strength and hardness with increase in aluminum content are associated with the reduction in the amount of the soft, ductile alpha phase and increase in the strong beta phase and brittle eutectoid phase.

For the range of 3 to 6 per cent iron, the only contribution noted is the increase in hardness upon increasing the iron content. The increase in kappa phase appears to explain the increase in hardness.

Nickel causes a change in microstructure which is analagous to reducing the aluminum content; that is, increasing nickel contents produce a reduction in the amount of beta phase and an increase in the amount of alpha phase. Further, nickel tends to refine the eutectoid phase and thus the tendency to brittleness normally associated with the eutectoid phase is reduced. A combination of high nickel content (5 to 6 per cent) plus high aluminum content results, on heat treating, in an alloy of high strength with usable ductility so far as structural applications are concerned. Similarly, an alloy with lower aluminum content, 9.5 to 10.5 per cent, has higher yield strength with good ductility if the nickel content is increased to a range of 5 to 6 per cent.

All the alloys studied, except alloy 22, responded to heat treatment; that is, yield strength, tensile strength and hardness were increased and elongation was decreased. From an examination of the microstructure after quenching, it can be reasoned that the amount of martensitic phase produced on quenching is the most important cause for changes in properties on heat treatment.

By various combinations of quenching and tempering temperatures, it is possible to produce different strength-ductility combinations although one must be sacrificed to obtain the other. For a given quenching temperature, the choice of tempering temperature will govern the properties. The higher the temperature the lower will be the hardness and strength and the greater will be the elongation.

Two considerations govern the choice of quenching temperature:

1) The possibility of severe grain growth.
2) The resultant ductility-strength combination.

The first item was not explored in this study. Higher quenching temperatures produced higher strength but lower ductility for the same tempering temperature. It is believed that the higher ductility is a result of the greater amount of alpha which exists on quenching from lower temperatures. For the heat treatments chosen, there is no significant difference whether the material is quenched or air cooled from the tempering temperatures. One can select air cooling to avoid unnecessary residual stress. A study of the microstructures of air cooled material reveals why it is necessary to avoid heating to within the eutectoid range, since the samples air cooled from 600 C (1112 F) and 500 C (932 F) showed almost continuous areas of eutectoid phase.

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# ALUMINUM ALLOY 356+BE PERMANENT MOLD CASTING

## *Centrifugal force effect on mechanical properties*

by A. J. Iler


### ABSTRACT

**A study was made, in connection with an Air Material Command R & D contract, conducted by the author's company (Contract AF 33(600)-38450), to determine the effect of forces applied by the permanent mold centrifuge on mechanical properties of aluminum alloy 356 + Be. Using a special mold castings were produced employing various centrifugal pressures. Coupons were cut from these castings from areas related to the feed and chill. These were radiographed, photomicrographed and tensile tested.**

**This paper discusses an analysis, concluding that superior properties obtainable by use of the centrifuge significantly exist only when coupled with configurations difficult to achieve by other methods without practices deleterious to casting integrity.**


### INTRODUCTION

Considerable interest has been focused on high tensile properties produced in castings made by the permanent mold centrifuge method. Fins for various missiles are currently being made by this method with tensile properties consistently higher than could normally be obtained by other means. Why there should be higher properties has been a subject of conjecture and speculation.

The centrifuge has been a necessary implement for producing this hardware. Some of these missile fins taper on the trailing edge to 0.050 in. thickness, which would be difficult to achieve by any other means and still produce the high tensile properties required. It is obvious that the mechanics of this method perform certain functions known to enhance elevated properties in castings. These include a faster chill, higher thermal gradients in the solidifying casting and cooler pouring temperatures.

These factors can easily be induced in statically poured permanent mold castings with various expedients used by the founder by application of chills or heat to specific areas of a mold, but the hydrostatic pressures of a static mold are not great enough to fill large areas of thin section.

It is the opinion of many foundrymen that these factors are solely responsible for higher properties achieved. However, there has remained the question of whether the centrifugal force actually had a direct influence on grain structure or homogeneity.

To investigate this possibility, a mold was made for the centrifuge which would produce castings whereby this subject could be studied without any appreciable influence of other solidification factors.

### EXPERIMENTAL PROCEDURE

When a casting cavity is filled with molten metal, the influence of pressure upon it is modified by casting configuration, the rate of chill and whether solidification is progressive toward the riser or sprue. To eliminate these factors a mold was developed (Fig. 1) which produced cast slabs one in. thick by 5 in. wide by 7 in. long. The metal was poured directly into an end riser. The opposite end (the farthest from the axis) was chilled with removable end chills.

A typical composition of the aluminum alloy used was:

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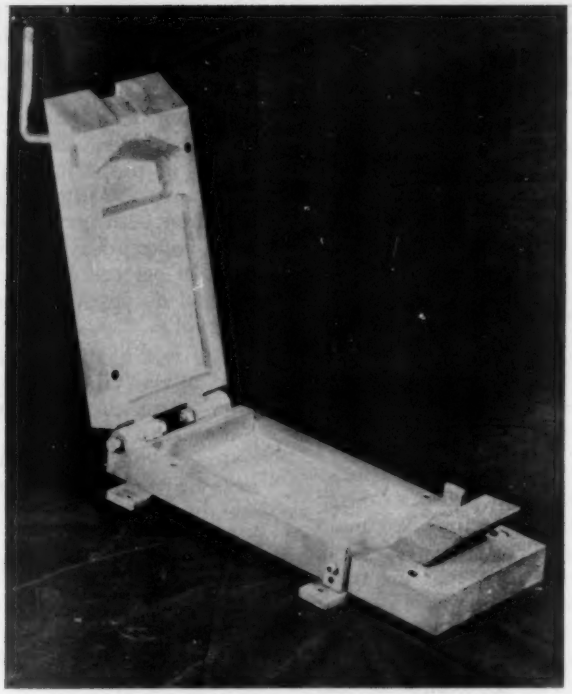


Fig. 1 — Mold developed which produced cast slabs one in. thick × 5 in. wide × 7 in. long.

| Zn | Ti | Mn | Si | Mg | Fe | Cu | Be |
|---|---|---|---|---|---|---|---|
| 0.10 | 0.16 | 0.04 | 6.70 | 0.71 | 0.13 | 0.14 | 0.09 |

The castings were solution heat treated at 1010 F (542 C) for 14 hr, quenched in 140 F (60 C) water with maximum delay of 22 sec, held at room temperature for 26 hr, then artificially aged at 340 F (171 C) for 6 hr.

The castings produced were thought to have enough volume to allow a free liquid movement during various applications of force and stages of solidification. Also, the solidification nucleated at the mold casting interfaces on the flat sides of the castings would not reduce the effect of centrifugal pressure in any area until the mass solidified.

The pressures exerted by centrifugal force were constant against the solidifying metal. Induced chilling started solidification at the chill end, and the applied force diminished more in proportion to the diminishing radius as solidification progressed toward the turning axis than by cross-sectional reduction.

The formula used to calculate the applied pressure at any given point did not take into consideration solidification nucleated at the mold sides nor the centrifugal arc, but was based upon a noncompressible fluid in a container being constant in cross-sectional area. The formula:

$$p = \rho\,\omega^2 x\left(r + \frac{x}{g}\right)$$

Where:

$\rho$ = (density in lb/in.$^3$).
$\omega$ = (angular velocity in radians/sec).
$r$ = (distance in in., axis to casting).
$x$ = (distance in in., casting).
$g$ = (acceleration of gravity in in./sec).

The highest pressure applied (at 275 rpm) amounted to 17.3 psi at the chilled end of the casting. Figure 2 plots pressures at indicated distances

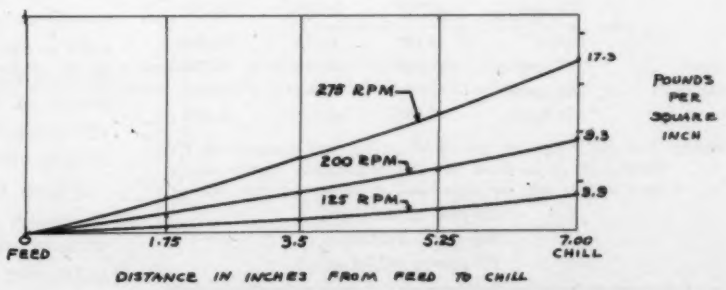



Fig. 2 — Pressure developed by centrifugal force on 1 × 5 × 7 in. slab casting. Calculations based upon fluid mass.

TABLE 1 — TENSILE PROPERTIES (AVG.) VS. CENTRIFUGE SPEEDS

| Static, psi | 125 rpm, psi | 200 rpm, psi | 275 rpm, psi |
|---|---|---|---|
| Tensile Strength at Chill End | | | |
| 52,625 | 53,610 | 53,125 | 52,225 |
| Tensile Strength at Center | | | |
| 48,975 | 50,200 | 48,525 | 48,275 |
| Tensile Strength at Feed End | | | |
| 48,850 | 49,900 | 50,125 | 49,000 |
| Yield Strength at Chill End | | | |
| 46,400 | 46,900 | 47,000 | 45,850 |
| Yield Strength at Center | | | |
| 45,000 | 45,830 | 46,200 | 44,900 |
| Yield Strength at Feed End | | | |
| 44,500 | 44,750 | 44,975 | 42,750 |

from the feed end of the casting which was considered zero pressure, ignoring the slight pressure exerted by the gated metal in the pouring cup. The mold was mounted on a centrifuge in a tilting position, about 20 degrees from horizontal, so that the mold would fill for a statically poured casting.

The chill end was placed 15 in. from axis of rotation. The chills were of iron, like the rest of the mold, but were removable so that they could be held at room temperature until the pouring operation took place. The mold temperature at the time of pouring ranged from 440 to 500 F (227 to 260 C) and the pouring temperature was 1360 F (738 C). Castings were produced under these conditions in sequence at centrifugal speeds of static, 125, 200 and 275 rpm. The castings were heat treated to a T-6 condition.

## RADIOGRAPHIC ANALYSIS

X-rays were taken of each casting on fine grained film to show as much detail as possible, but did not reveal an unacceptable quality condition for a normal Class 1A casting. A segregation was noted, however, which may have some significance—the mode of solidification of cast aluminum alloys is the source of dispersed microporosity, they freeze in a mushy state with nucleation occurring nearly simultaneously throughout the mass with solid and liquid existing in intimate contact over most of the casting.

The radiographs indicated that this microporosity dispersion was uniformly sporadic throughout the statically poured casting. However, it rose toward the header in proportion to the amount of centrifugal force until the discernible microporosity in the casting, spun with greatest speed (275 rpm), indicated a small concentration near the feed end of not more than 10 per cent of the mass. The rest of the casting had no apparent discontinuity as revealed by radiography.

The radiographs taken of these castings (Figs. 3a and 3b) give credence to the opinion that centrifugal force contributes to denser castings. This, however, is not necessarily true on typical thin section air frame castings, where solidification is more rapid and the feed by centrifugal force is cut off by freezing of members adjacent to the header (Fig. 3c).

TABLE 2 — TENSILE PROPERTIES PER DISTANCE FROM CHILL (AVG.)

| | Centrifugal Speed | Feed End | Center | Chill End |
|---|---|---|---|---|
| | Static | 48,800 | 48,700 | 52,400 |
| Tensile | 125 rpm | 49,800 | 50,300 | 53,500 |
| Strength | 200 rpm | 50,000 | 48,900 | 53,200 |
| | 275 rpm | 48,900 | 48,200 | 52,700 |
| | Static | 43,100 | 44,200 | 45,100 |
| Yield | 125 rpm | 44,600 | 45,600 | 46,900 |
| Strength | 200 rpm | 46,000 | 45,900 | 46,900 |
| | 275 rpm | 42,500 | 44,500 | 45,100 |

NOTE: The chill end of the mold cavity was located on the centrifuge 15 in. from the axis of rotation; therefore, the rpm speeds may be expressed in surface ft/min as:

125 rpm = 668 sfpm
200 rpm = 1070 sfpm
275 rpm = 1471 sfpm

## TENSILE PROPERTY ANALYSIS

Standard 0.505 in. round test bars were machined from three positions in each casting. All were cut transverse to the direction of solidification, and were taken from the chill ends, centers and feed ends.

Figure 4 indicates that highest tensile properties were obtained at the lowest speed used above static (125 rpm), and lowest properties were about equal for both static and the highest speed (275 rpm). However, the scatter band was narrow (not over 3000 psi difference), indicating little improvement in mechanical properties due to increased pressure. Figure 5 shows tensile properties as related to distance from chill.

## SURFACE CONDITIONS

The statically poured casting, with negligible amount of hydrostatic pressure, did not reproduce the mold face except in general shape. It was noted, however, that detail of brush marks on the refractory mold was reproduced in greater acuity in proportion to the rate of rotation. At 275 rpm the finest detail completely reproduced on the casting. This could explain why the statically poured casting had slightly lower mechanical properties, since a shrinkage gap between the casting and the mold would affect the rate of heat transfer.

## METALLOGRAPHIC ANALYSIS

Metallographic mounts were made of the test bars removed from chill and feed ends of castings produced at four speeds. The coupons were taken adjacent to the fracture so that sub-surface structure could be studied near the center of the gage length, or the center of the casting. Thus, as much as possible the influence of grain refining by quicker chilling at the casting-mold interface was eliminated.

Figure 6 illustrates grain size and the distribution of silicon precipitates. It will be noted that a comparison of photomicrographs shows a similarity in grain size regardless of the rotational speed, whereas

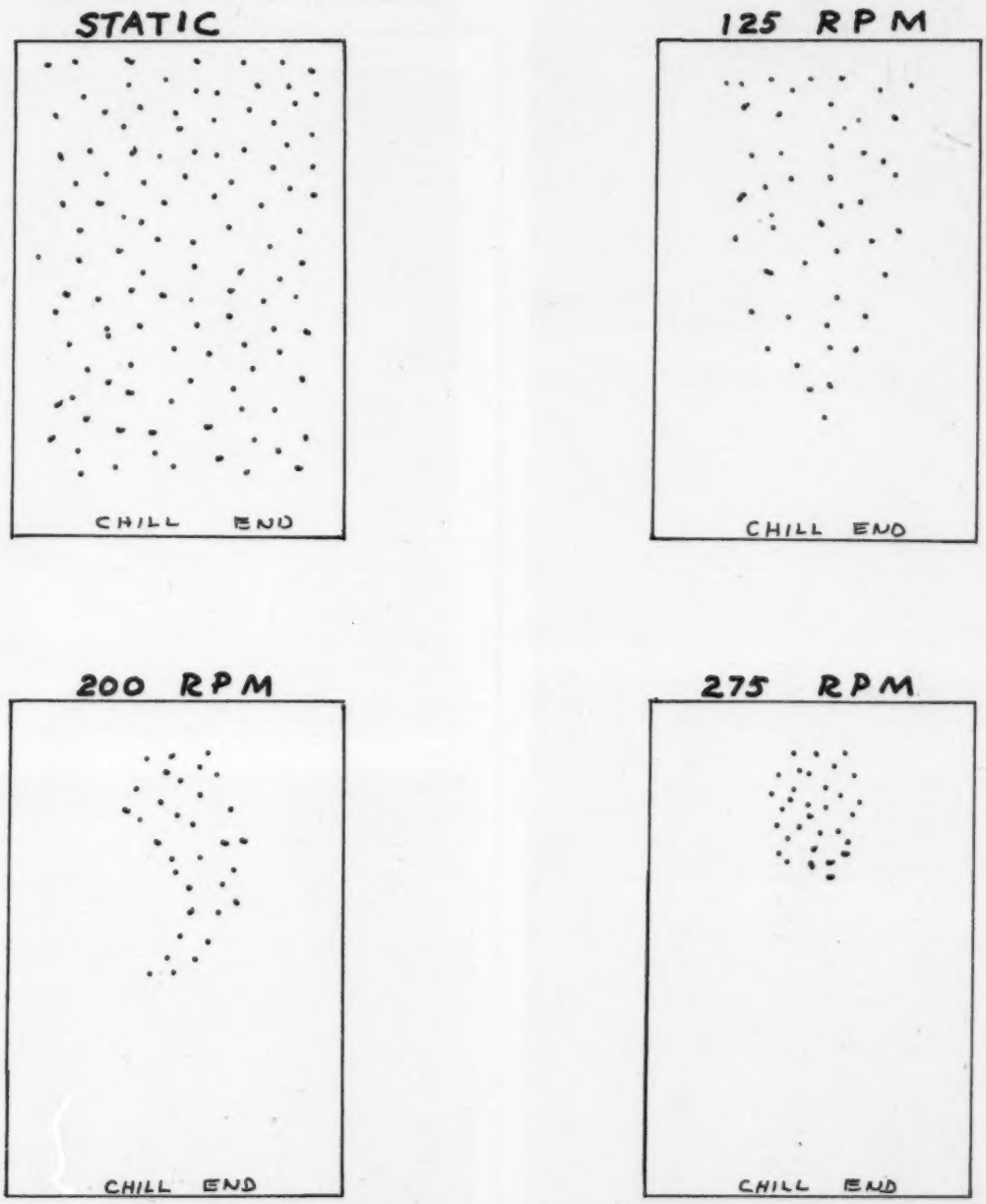
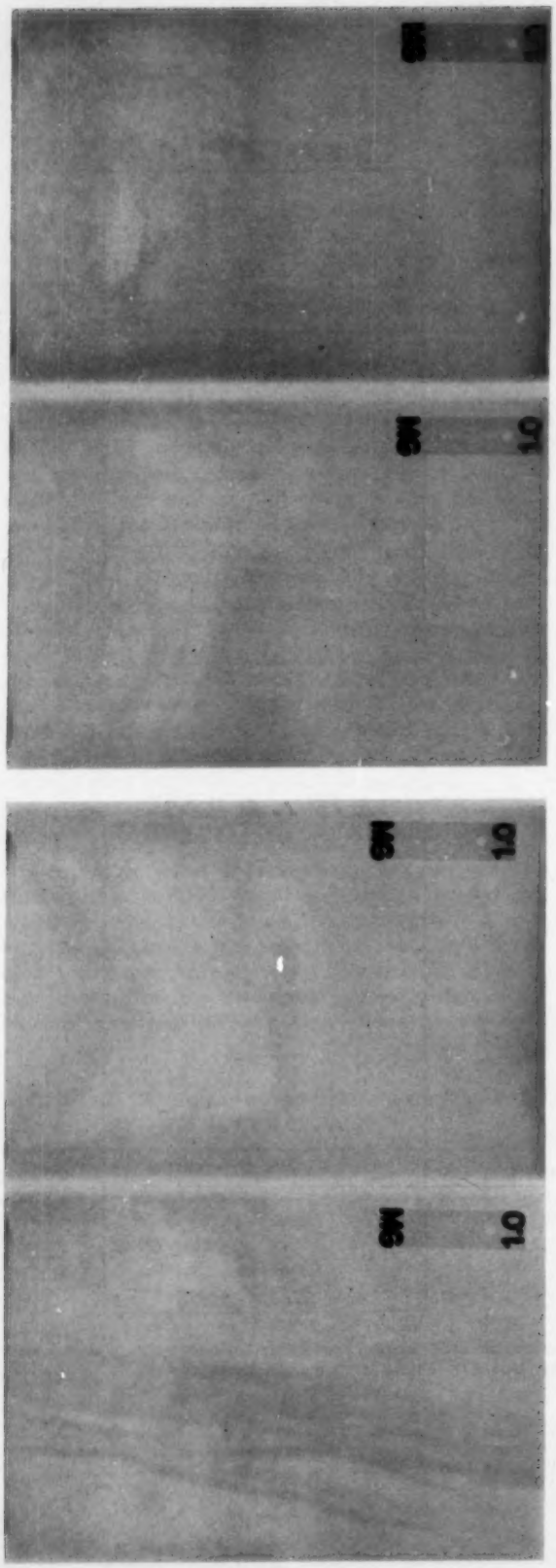



Fig. 3a — Radiographic definition of distribution of solidification voids for centrifugally cast slab of 356 + Be aluminum alloy (see Fig. 3b for actual radiographs).

Fig. 3b(1) — X-ray film, taken of one in. thick slabs 5 in. wide × 7 in. long cast in centrifugal permanent mold at indicated speeds, of magnesium alloy AZ-91. Light areas have least density. *Bottom to top* — no. 16 — static, no. 10 — 50 rpm (950 sfpm), no. 11 — 75 rpm (1425 sfpm) and no. 12 — 105 rpm (1995 sfpm). See also Fig. 3b(2).



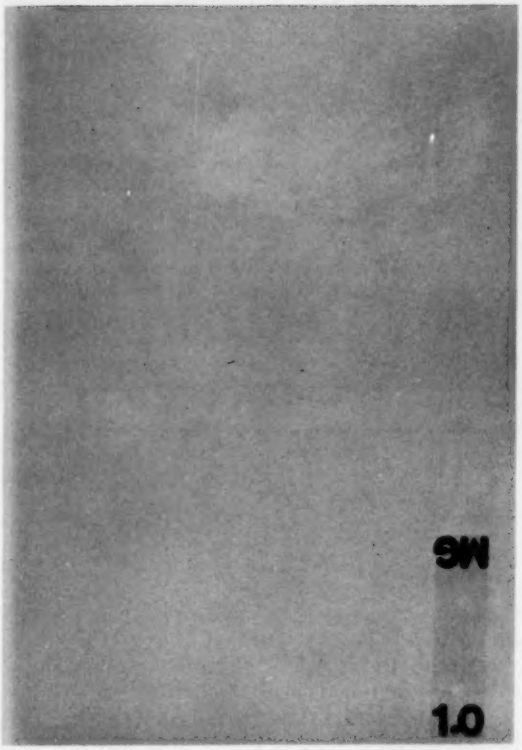


Fig. 3b(2) — X-ray film, taken of one in. thick slabs 5 in. wide × 7 in. long cast in centrifugal permanent mold at indicated speeds, of magnesium alloy AZ-91. Light areas have least density. *Top left* — no. 13 — 140 rpm (2660 sfpm), *bottom, left* no. 14 — 185 rpm (3515 sfpm) and *above* no. 15 — 220 rpm (4180 sfpm).

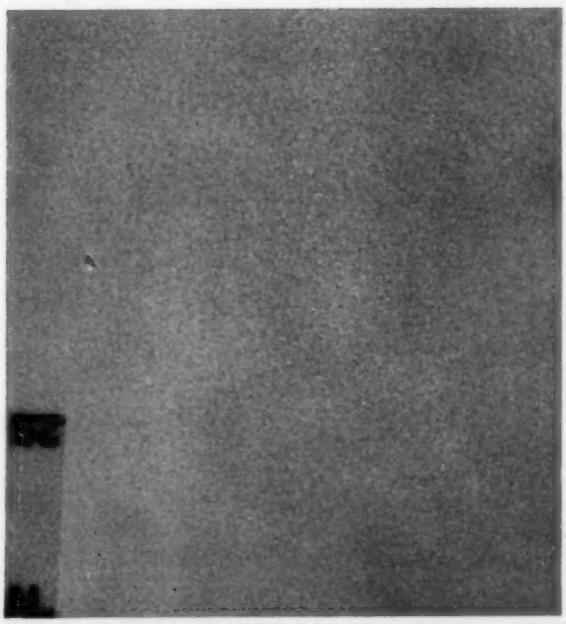
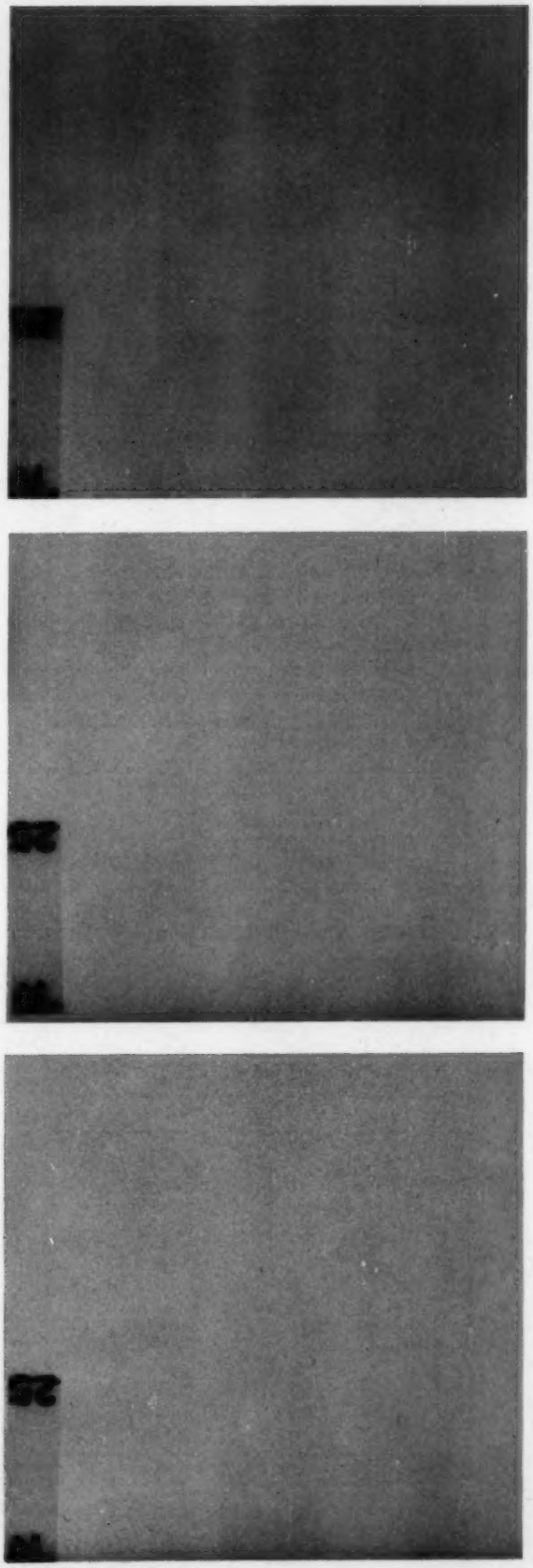


Fig. 3c — X-ray film of aluminum alloy 356 plus Be (compare with Figs. 3a and 3b). *Above* — static pour, feed end; *lower right* — 75 rpm, feed end; *center right* — 140 rpm, feed end; *upper right* — 220 rpm, feed end.



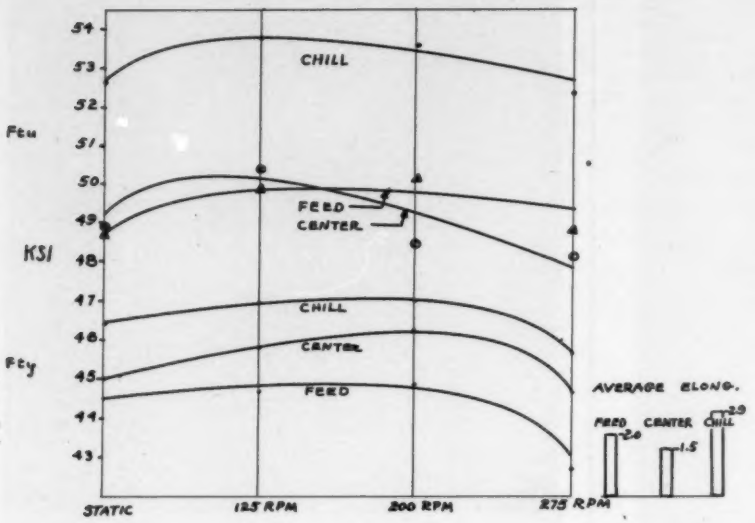



Fig. 4 — Tensile properties vs. centrifuge speeds. Ftu-tensile strength, Fty-yield strength, K51-psi × 1000.

the significant difference is found in relation to the chill or riser ends. The grain is fine and well dispersed at the chill end of all castings, and the grain is coarse at all of the riser ends. There appears to be a slight grain refinement at the chill end of the casting produced at highest speed over that produced statically, but negligible differences are noted in tensile specimens pulled.

## CONCLUSION

Any force sufficient to fill the mold rapidly and maintain pressure of the solidifying mass to the mold-casting interface bears an influence on tensile properties as satisfactorily as higher pressures. This is usually done better by conventional usage of gating, the exception being casting configurations with webs too thin for static pouring.

Preferential dispersion of microporosity and improvement in homogeneity noted in this study are greatly minimized in thin wall castings due to the rapidity of solidification. Since this was not significantly effective in tensile properties, it can be disregarded from the standpoint of property improvement.

The advantages gained by use of the centrifuge are to be expressed in terms of hardware rather than metallurgy. Castings are being produced by the industry in configurations impossible to achieve by other means. Thin sections can be filled with good, sound casting. Colder molds can be used, thus providing a more effective chill. Lower pouring temperatures can be used, which may sometimes be evaluated by less gas and more stable chemistry.

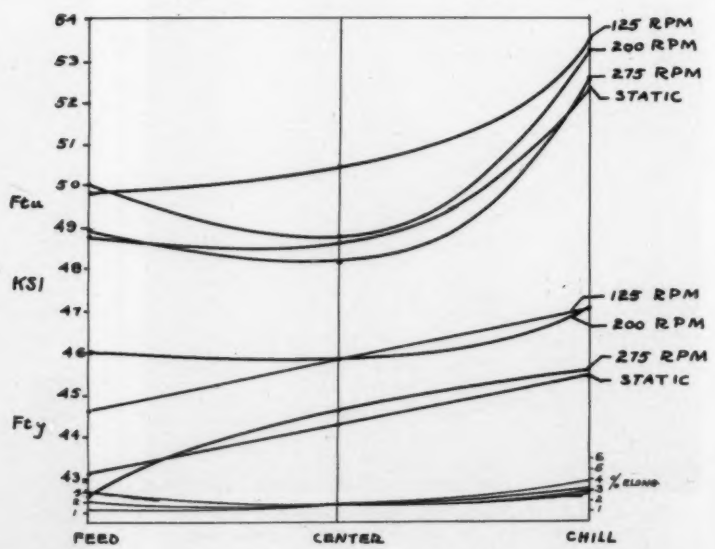



Fig. 5 — Tensile properties per distance from chill. Ftu-tensile strength, Fty-yield strength, K51-psi × 1000.

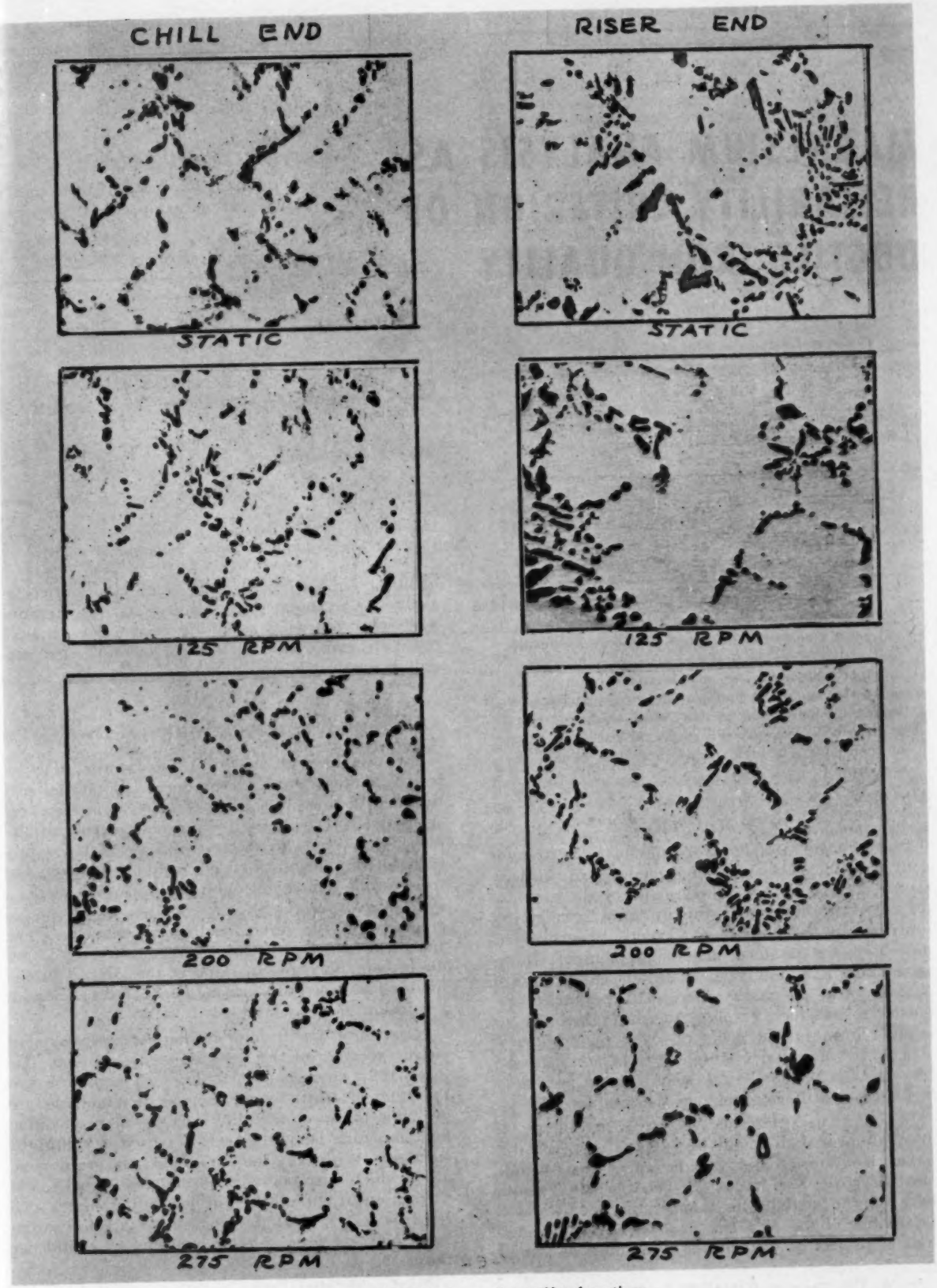



Fig. 6 — Microstructure of centrifugal castings.

# MAGNESIUM ANALYSIS AS RELIABILITY CRITERION OF DUCTILE IRON QUALITY

by AFS Ductile Iron Division
Research Committee (12-K)*

## ABSTRACT

**The intent of this project was to determine with what degree of reliability magnesium analysis could be employed as a control of inspection procedure for ductile iron quality. Nine identical samples were distributed to different laboratories for analysis. As a result it was found that agreement between laboratories on retained magnesium content is, in some cases, poor, and that, at present, the test coupon is a more satisfactory method of controlling quality in ductile iron than magnesium analysis.**

## INTRODUCTION

Ductile iron owes its properties of high strength with good ductility and a high modulus of elasticity to the nodular or spheroidal shape of the graphite present. Control of the graphite form is largely a function, in the absence of certain subversive elements, of the retained magnesium content. If retained magnesium is to be used for either control or inspection of quality, the small percentage required to produce graphite in the spheroidal or nodular form—on the order of a few hundredths of a per cent—necessitates a high level of accuracy and precision in the analytical technique as well as a high degree of homogeneity in the samples.

In order to determine (1) what agreement in magnesium analysis might be expected between different laboratories and (2) the effect that sample microstructure and possible heterogeneity would have on the analysis for magnesium, the Research Committee of the AFS Ductile Iron Division undertook a project of distributing a number of sets of identical samples to nine different laboratories for analysis. The intent of the project was to determine with what degree of reliability magnesium analysis could be employed as a control or inspection procedure for ductile iron quality.

## PROCEDURE

Two commercial analytical laboratories, one research laboratory and six producer laboratories participated in this project. All of the laboratories were accustomed to analyzing for magnesium and all employed some type of spectrochemical analytical procedure. The instrumentation included one type of spectroscope employing visual comparison with a standard; conventional spectrographs using photographic film and direct reading spectrometers. The analytical techniques employed consisted of either direct point to plane sparking of the samples or the sparking of electrodes impregnated or covered with a solution of the samples.

No attempt was made to evaluate or pass judgment on the accuracy of the reported analytical data. Unfortunately, no standard samples, such as those issued for other elements by the Bureau of Standards, are available for spectrochemical analysis of magnesium, and thus each laboratory employs its own calibrated standards. With only the samples cast in green sand, all of which were poured from one 300 lb ladle, was there a sufficient number of identical samples analyzed by each laboratory to permit a statistical evaluation of the degree of reproducibility or confidence limits.

Even with these samples, there could be some doubt concerning the validity of the statistical procedure, since there were only five samples involved

*H. W. RUF, Chairman, D. MATTER, Secretary, A. W. ANDERSON, D. L. CREWS, H. G. HAINES, A. H. RAUCH and W. M. SPEAR.

and since there is a possibility of difference in magnesium content due to segregation or a concentration gradient within the ladle. It was thought, however, that the results might be indicative of the reproducibility and agreement that might be expected.

## SAMPLE IDENTIFICATION

The identification and description of the samples employed in this program are shown in Table 1.

TABLE 1 — IDENTIFICATION OF SAMPLES

| Sample No. | Size, in. | Type of Mold | Samples per Ladle | Samples from each Ladle |
|---|---|---|---|---|
| 1-5, incl. | ⅜x1½x10 (poured horizontally) | Green Sand | 5 | 1-5, incl. |
| 6C-10C, incl. | 2 dia. x 1 | Copper Chill | 2 | 6C and 6X, 7C and 7X, 8C and 8X. |
| 6X-10X, incl. | 1 sq. x 6 (poured vertically) | Baked Core Sand | | 9C and 9X, 10C and 10X, 11C and 11X. |
| 11-15, incl. | 2 dia. x 1 | Baked Core Sand | 2 | 11 and 12, 13 and 14, and 15. |

NOTE: Samples 1-5 were cast from basic cupola melted iron while the remaining samples were cast from iron melted in an indirect rocking arc furnace.
Samples 11 and 13 were poured from the first part and samples 12 and 14 from the last part of each of two 800 lb ladles. Sample 15 was poured from the first part of a third 800 lb ladle. All of these samples were analyzed on both top and bottom faces and the surface analyzed is identified as *T* or *B*, respectively.

The magnesium contents reported for the various sets of samples are as shown in Tables 2, 3 and 4.

## DISCUSSION

The green sand molded samples indicate that the degree of reproducibility or coefficient of variation for some laboratories and procedures is good, but the agreement between the laboratories is, in some instances, quite poor. It would appear from this

TABLE 3 — CHILLED vs. CORE MOLDED SAMPLES

| Laboratory | Magnesium Content, % | | | | | | | | | |
|---|---|---|---|---|---|---|---|---|---|---|
| | 6C† | 6X‡ | 7C | 7X | 8C | 8X | 9C | 9X | 10C | 10X |
| A | — | 0.043 | — | 0.038 | — | 0.046 | — | 0.093 | — | 0.106 |
| C | 0.052 | 0.040 | 0.059 | 0.043 | 0.058 | 0.043 | 0.150 | 0.095 | 0.163 | 0.110 |
| D-1 | 0.076 | 0.049 | 0.048 | 0.050 | 0.056 | 0.045 | 0.093 | 0.060 | 0.095 | 0.076 |
| D-1a | 0.052 | 0.044 | 0.043 | 0.042 | 0.054 | 0.039 | — | — | — | — |
| D-2 | 0.038 | 0.072 | 0.045 | 0.048 | 0.047 | 0.052 | — | — | — | — |
| D-3 | — | — | — | — | — | — | 0.096 | 0.088 | 0.108 | 0.106 |
| E | 0.057 | 0.049 | 0.054 | 0.044 | 0.059 | 0.048 | — | — | — | — |
| G | 0.027 | 0.038 | 0.022 | 0.032 | 0.026 | 0.024 | — | — | — | — |
| H | 0.042 | 0.049 | 0.031 | 0.047 | 0.039 | 0.039 | — | — | — | — |
| I | 0.050 | 0.049 | 0.036 | 0.046 | 0.051 | 0.040 | — | — | — | — |
| Max. | 0.076 | 0.072 | 0.059 | 0.050 | 0.058 | 0.052 | 0.150 | 0.095 | 0.163 | 0.110 |
| Min. | 0.027 | 0.038 | 0.022 | 0.032 | 0.026 | 0.024 | 0.093 | 0.060 | 0.095 | 0.076 |
| Avg. | 0.049 | 0.048 | 0.042 | 0.043 | 0.049 | 0.042 | 0.113 | 0.084 | 0.122 | 0.100 |

†C = Chilled*
‡X = Core Molded**

*Analysis made on face of samples in contact with copper mold.

**Analysis made on a surface at right angle to the longitudinal axis.

NOTE: D-2 and D-3 represent analysis with the same equipment using two different techniques. Results for 9C, 10C, 9X and 10X represent average of analysis of top and bottom faces in each case.

TABLE 4 — COMPARISON OF ANALYSIS ON TOP AND BOTTOM SURFACE OF INDIVIDUAL CORE MOLDED SAMPLES*

| Laboratory | Magnesium Content, % | | | | | | | | | |
|---|---|---|---|---|---|---|---|---|---|---|
| | 11T | 11B | 12T | 12B | 13T | 13B | 14T | 14B | 15T | 15B |
| A | 0.100 | — | 0.064 | — | 0.100 | — | 0.075 | — | 0.110 | — |
| B | 0.094 | 0.093 | 0.070 | 0.083 | 0.110 | 0.111 | 0.093 | 0.093 | 0.105 | 0.107 |
| C | 0.070 | 0.084 | 0.053 | 0.054 | 0.087 | 0.089 | 0.075 | 0.078 | 0.103 | 0.098 |
| D-1 | 0.065 | 0.066 | 0.056 | 0.055 | 0.069 | 0.078 | 0.055 | 0.050 | 0.080 | 0.075 |
| D-2 | 0.082 | 0.085 | 0.064 | 0.073 | 0.096 | 0.092 | 0.076 | 0.068 | 0.096 | 0.100 |
| E | 0.074 | 0.095 | 0.062 | 0.077 | 0.095 | 0.078 | 0.091 | 0.077 | 0.103 | 0.111 |
| G | 0.066 | 0.068 | 0.054 | 0.054 | 0.070 | 0.070 | 0.059 | 0.058 | 0.068 | 0.068 |
| H-1 | 0.081 | 0.087 | 0.061 | 0.052 | 0.091 | 0.087 | 0.072 | 0.068 | 0.097 | 0.093 |
| H-2 | 0.088 | 0.090 | 0.068 | 0.066 | 0.090 | 0.091 | 0.075 | 0.072 | 0.100 | 0.110 |
| I | 0.093 | 0.090 | 0.066 | 0.064 | 0.092 | 0.095 | 0.073 | 0.068 | 0.104 | 0.110 |
| Max. | 0.100 | 0.095 | 0.070 | 0.083 | 0.110 | 0.111 | 0.093 | 0.093 | 0.110 | 0.111 |
| Min. | 0.065 | 0.066 | 0.053 | 0.052 | 0.069 | 0.070 | 0.055 | 0.050 | 0.068 | 0.068 |
| Avg. | 0.081 | 0.084 | 0.062 | 0.064 | 0.090 | 0.088 | 0.074 | 0.070 | 0.097 | 0.097 |

*2 in. dia. x one in. high Samples

NOTE: H-1 and H-2 represent analysis with the same equipment using two different techniques.

TABLE 2 — GREEN SAND MOLDED SAMPLES

| Laboratory | Magnesium Content, % | | | | | | | | Coeff. of Variation | Predicted Range 95% Confidence Limits, % |
|---|---|---|---|---|---|---|---|---|---|---|
| | 1 | 2 | 3 | 4 | 5 | Max. | Min. | Avg. | | |
| A | 0.047 | 0.045 | 0.047 | 0.046 | 0.045 | 0.047 | 0.045 | 0.046 | 2.17 | 0.044-0.048 |
| B | 0.051 | 0.050 | 0.050 | 0.054 | 0.050 | 0.054 | 0.050 | 0.051 | 3.39 | 0.048-0.054 |
| C | 0.044 | 0.042 | 0.044 | 0.042 | 0.045 | 0.045 | 0.042 | 0.043 | 3.09 | 0.040-0.046 |
| D-1 | 0.048 | 0.038 | 0.043 | 0.038 | 0.045 | 0.048 | 0.038 | 0.042 | 10.35 | 0.034-0.050 |
| D-1a | 0.036 | 0.040 | 0.039 | 0.040 | 0.040 | 0.040 | 0.036 | 0.039 | 4.44 | 0.036-0.042 |
| D-2 | 0.055 | 0.040 | 0.037 | 0.050 | 0.055 | 0.055 | 0.037 | 0.047 | 17.81 | 0.030-0.064 |
| E | 0.051 | 0.048 | 0.055 | 0.048 | 0.052 | 0.055 | 0.048 | 0.051 | 5.81 | 0.045-0.057 |
| F | 0.034 | 0.034 | 0.036 | 0.038 | 0.032 | 0.038 | 0.032 | 0.035 | 6.55 | 0.030-0.040 |
| G | 0.041 | 0.049 | 0.047 | 0.047 | 0.046 | 0.049 | 0.041 | 0.046 | 6.52 | 0.040-0.052 |
| H | 0.058 | 0.047 | 0.055 | 0.052 | 0.054 | 0.058 | 0.047 | 0.053 | 7.69 | 0.045-0.061 |
| I | 0.051 | 0.052 | 0.052 | 0.051 | 0.052 | 0.052 | 0.051 | 0.052 | 1.08 | 0.051-0.053 |
| Max. | 0.058 | 0.052 | 0.055 | 0.054 | 0.055 | | | | | |
| Min. | 0.034 | 0.034 | 0.036 | 0.038 | 0.032 | | | | | |
| Avg. | 0.047 | 0.044 | 0.046 | 0.046 | 0.047 | | | | | |

NOTE: D-1 and D-1a are the reported independent observations of two analysts using the same equipment; D-2 represents the results obtained with another type of equipment in the same laboratory.

phase of the program that the green sand molded specimen of this particular dimension shows good uniformity within each sample and among separate samples poured from the same ladle. However, these data represent only one set of samples and one magnesium level and should not be considered conclusive.

Inspection of the data for the chill molded versus core molded samples indicate not only poor agreement between laboratories, but also an inconsistency within a given laboratory on the part played by the molding practice and the microstructure. The comparison of the data for top and bottom of 2 in. dia. x one in. thick core molded samples indicate that some laboratories found no appreciable or significant difference between these two surfaces, while others reported a considerable difference. Again, the agreement between laboratories in some instances was poor.

Consideration of the information gathered in this program demonstrates that, with certain types of equipment and analytical techniques, a fairly good degree of reproducibility can be achieved within a given laboratory and, with a proper sampling procedure, magnesium analysis can be reliable for in-plant control. There remains to be established, however, a relationship between retained magnesium content, the degree of accuracy and precision of magnesium analysis required, and the graphite form before magnesium analysis can be employed to control or assess ductile iron quality.

Some idea of the relationship between retained magnesium content and graphite form can be obtained from a consideration of experience of one of the producers involved in this program. Shown in Table 5 is a breakdown of the results of magnesium analysis in terms of its relationship to the amount of vermicular or stubby graphite, as well as flake graphite. The data represent an evaluation of 964 samples taken from Y-block test bars cast over a period of seven months. These test bars were cast from the first and every tenth ladle following. The results are not used for routine process control—this producer examines a test coupon poured from each ladle treated for control purposes. This producer's specification, on the basis of experience, allows up to 10 per cent vermicular graphite.

TABLE 5 — MAGNESIUM CONTENT vs. GRAPHITE FORM

| Mg Range Group, % | No. of Test Bars | Vermicular Graphite | | Percentage of Range Group Above 10% Vermicular Graphite |
|---|---|---|---|---|
| | | Trace to 10% | Above 10% | |
| 0.010-0.020 | 4 | | 4* | 100.0 |
| 0.021-0.025 | 15 | 1 | 14** | 93.3 |
| 0.026-0.030 | 17 | 4 | 13 | 76.5 |
| 0.031-0.035 | 37 | 23 | 14 | 37.9 |
| 0.036-0.040 | 60 | 27 | 1 | 1.7 |
| 0.041-0.045 | 86 | 15 | 0 | 0 |
| Above 0.046 | 745 | 0 | 0 | 0 |

*Three samples exhibited entirely flake graphite.
**One sample exhibited entirely flake graphite.

The data in Table 5 should not be construed as setting minimum magnesium levels for obtaining essentially nodular graphite. Until such a time as a reliable magnesium standard is established, the magnesium content cannot be employed as a criterion. Furthermore, with the analytical technique used by this producer, the 95 per cent confidence limits for magnesium analysis are considered too wide to permit the use of the magnesium content as a reliable criterion of acceptability. The analysis is used as a check on magnesium recovery and effectiveness.

There is some reason to believe that the minimum amount of retained magnesium required to secure a satisfactory graphite condition may vary with the melting practice or initial sulfur content. There is also some evidence that a change in a given producer's practice could alter the retained magnesium content required and make the analysis for magnesium unreliable as a criterion.

## CONCLUSION

The agreement between various laboratories on retained magnesium content is, in some instances, quite poor. It is believed that with a suitable standard and certain techniques and instrumentation reliable and reproducible results may be achieved. However, even with a satisfactory procedure for magnesium analysis, there are other considerations which could affect the reliability of analysis for retained magnesium as a means of controlling and assessing graphite form or ductile iron quality. These other factors are segregation, the effect of sample microstructure and soundness and the possibility of a shift in magnesium level required to obtain nodular graphite.

A producer, by employing the proper procedure and instrumentation, and with consistent sampling, may establish a relationship between retained magnesium for in-plant control purposes and graphite form. A user, however, analyzing castings for magnesium could be misled by segregation in the samples, or by failing to know the minimum magnesium content required to secure the desired graphite form with a particular supplier's practice.

Many producers employ for daily process control the examination with the microscope of either the test coupon devised by the Ductile Iron Research Committee, described in the annual AFS TRANSACTIONS, *vol. 68* (1960) or some form of it. This procedure, when properly applied, is reliable since it reveals directly the graphite form, which is the condition that must be controlled for quality. Control employing magnesium analysis relies upon maintaining the accuracy and precision of an analytical procedure. This is particularly important since small variations in analysis assume great importance where an element is present initially in only small percentages. The equipment that will give the necessary accuracy and precision along with the speed required is elaborate and expensive.

The Committee feels that the examination of a test coupon is, for the present, a more satisfactory method of controlling the quality of ductile iron. The National Bureau of Standards is in the process of developing a standard sample, No. 342, with a low magnesium content as currently used for producing ductile iron. When this is available, magnesium analysis will become more reliable.

# HIGH STRENGTH STRUCTURAL STEEL CASTINGS FOR AEROSPACE APPLICATIONS

by W. R. Roser


## ABSTRACT

**Steel castings offer a definite advantage to the designer of aerospace vehicles in the form of high strength and attendant weight and volume savings. The author's company's designers have utilized these advantages in the design of a supersonic trainer. Steel castings used on the trainer have been a problem from the standpoint of mechanical properties, and the causes of these problems appear to originate in the foundry. Controls in the foundry are required and will be more important as the alloys become more complex and the casting requirements more stringent. Foundrymen need to re-evaluate their position and up-grade their product in order to get their fair share of business from the aerospace movement.**


## INTRODUCTION

In order to conquer one of the last and most challenging frontiers left to man – that of outer space, and to build aircraft to fly higher and faster than ever before, the Design Engineer has turned to numerous materials, processes and designs to lighten and improve the capability, performance and reliability of the numerous complex components making up the system, whether it be aircraft or missile. Materials engineers are working constantly to improve known materials and alloys, and to discover new materials which can fulfill these requirements.

A great deal of work and money is being expended on materials such as beryllium, which shows a great deal of promise because of its high modulus of elasticity and low density. However, the metal has a large number of problems to be solved before it can take its place as a common structural material.

## STEEL INVESTMENT CASTINGS

Steel castings, especially those produced by the investment process, can fulfill the requirements for lightweight components. Even though steel is a dense material, unnecessary material not removable in machining is eliminated. Parts entirely too complex to be "hogged-out" and remain light weight or produced without elaborate welding, brazing or some other means of joining, can be cast with comparative ease. Entire assemblies can be made one integral part. The possibilities are great. Steel also offers good strength-to-density ratios and good strength-to-volume ratios, as well as moderately high-temperature strength.

In short, theoretically at least, steel castings offer great promise for the aerospace industry. It might be well to interject the thought at this point that the production cost savings attendent to high volume production are not taken into consideration, due to the fact that high volume production will probably never be achieved in this industry. The design-structural-materials engineering team at the author's company made good use of 17-4PH investment castings in the design of the supersonic trainer currently in production for the Air Force.

The figures show some typical illustrations of the castings utilized in this trainer. Figure 1 is one of the cast aileron hinges. Figure 2 is a canopy down-lock housing. The structural use of these castings dictate the necessity for a high degree of reliability and strength. There are approximately 20 different

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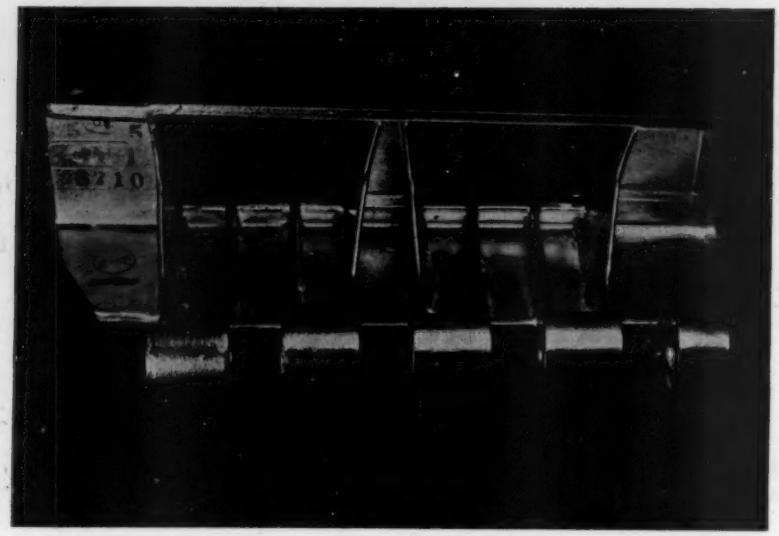


Fig. 1 — Cast aileron hinge.

17-4PH steel investment casting designs utilized in the aircraft.

It nearly became necessary, however, to abandon the entire 17-4PH steel castings program. Castings were tested early in the program and found to be lacking in ductility, and it became necessary for the company to investigate the causes. The table is an example of the typical mechanical properties obtained on several lots of castings from as many foundries. It should be noted that these were not only specimens taken from castings, but keel-gated, separately-cast, test bars.

| Test Bar | Yield Strength, psi | Ultimate Strength, psi | Elong., % in 1 in. | Red. of Area, % |
|---|---|---|---|---|
| 1 | 181,200 | 195,600 | 2.0 | 2.0 |
| 2 | 169,300 | 183,000 | 2.0 | —* |
| 3 | 175,500 | 195,800 | 2.0 | 3.0 |
| 4 | 168,300 | 201,200 | 1.0 | 2.0 |
| 5 | ——** | 180,400 | 0 | 0 |
| 6 | 167,400 | 177,100 | 0 | —* |

* Flat Bar from casting.
** Failed before 0.2% offset yield.
Remainder — round, keel-gated test bars.

## DUCTILITY PROBLEM

At this point it might be well to review the reasons believed to be responsible for lowered ductility. The first and most difficult to eliminate is the delta ferrite pattern, shown in Figure 3. This condition is believed to be caused by improper chemical balance or rate of solidification. The second is microshrinkage, as illustrated in Figure 4. These conditions do not noticeably detract from the strength of the casting unless present in large amounts, but severely lower the ductility. Another condition not shown which causes lowered strength is that of retained austenite. This is caused by out-of-specification chemistry, such as high nitrogen or carbon. This condition was rarely encountered in the castings purchased by the author's company.

In the face of the poor showing which the steel investment castings were making, it became necessary to rectify the situation to build airplanes. Had it not been for the fact that it would have been extremely

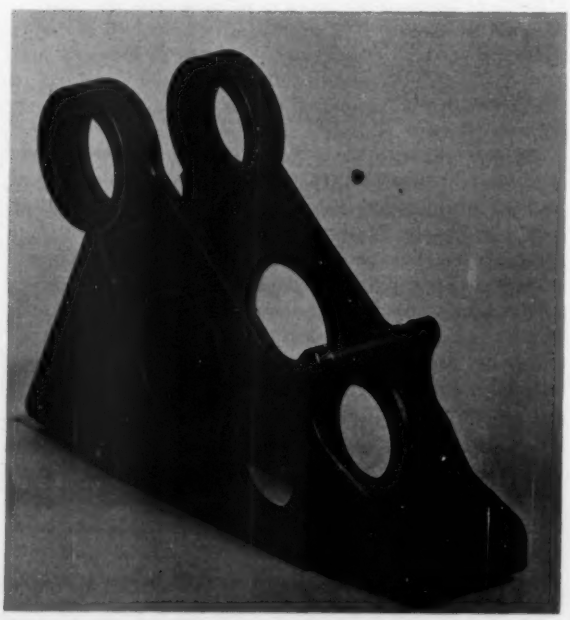


Fig. 2 — Canopy down lock housing.

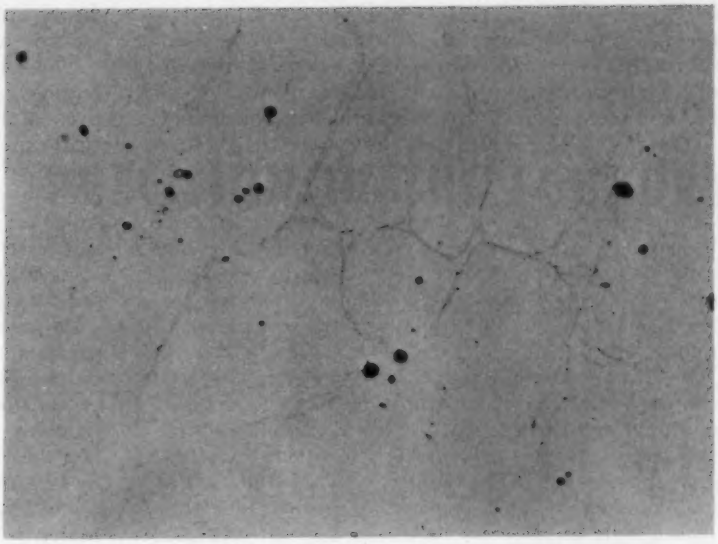


Fig. 3 — Delta ferrite pattern in 17-4 PH steel with elongation less than one per cent. This pattern is most difficult to eliminate. 500 ×.

costly to change to "hogged-out" parts (and disadvantageous from the weight penalty associated therein), the castings would have most certainly been abandoned. In order to assure that the castings were usable, it became necessary to issue a supplementary document to the specification.

This document provided a means of obtaining as much as possible an insight into the properties of the casting without imposing impossible controls on the foundry's operation. A provision was made whereby a foundry must furnish proof in the form of cast test bars from three foundry heats of material that it is able to handle the alloy. This seems like a simple requirement, but the results were surprising. There were foundries incapable of casting a passable test bar. There are also requirements for test bars cast within each flask of castings for purposes of verification of melt quality.

This is done principally because of the small foundry heats used in investment foundries, and the possible lack of uniformity from heat to heat. In some of these foundries, one heat is a single casting. The properties of 17-4PH appear to be rather sensitive to mold and pouring temperatures; and since it is difficult to control these exactly, the integrally cast bars provide a good indication as to whether the metal in the casting has been properly processed.

## TENSILE TESTING

At the outset of a foundry's production of a particular casting, one casting out of each production lot is cut up, and tensile tests made in order to assure that the properties in the casting are satisfactory as well as the correlation of the integrally cast bars to the casting properties. In addition, castings are peri-

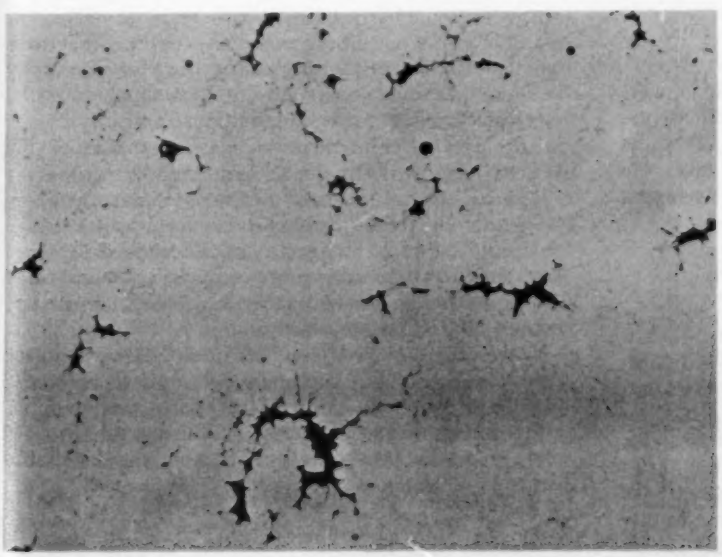


Fig. 4 — Microshrinkage in cast 17-4 PH steel with elongation less than one per cent. 100 ×.

odically cut up and tested. Needless to say, all of this testing is quite an expense. Nevertheless, it has been shown to be invaluable in preventing parts of inferior quality from finding their way into airplanes. When the failure of a single casting could cost the life of the pilot and/or the airplane, the comparative cost is infinitesimal.

This additional testing has made buying of castings more than a difficult job, due to the fact that although the strength requirements for these castings are not in excess of current specification requirements, the company insists that they are obtained and tests to make sure they are. The result is a higher than usual rejection rate.

As was previously stated, 17-4PH steel in the cast form is a rather difficult alloy in which to get the optimum properties. Chemical composition, pouring and mold temperatures and heat treatment are all of great importance to the properties obtainable in the final product. What is being done to solve the problems of the alloy?

## REVISED CHEMICAL COMPOSITION

The producers of master heat materials have worked to come up with a revised chemical composition to assure that castings achieve the required minimum strength and ductility. It is unfortunate that this is but half the answer. At least two major aircraft and missile producers have complained that the problem of low ductility still exists. Therefore, it must be supposed that the answer lies in a place other than the chemical composition. This leaves foundry practices.

However, let us not criticize only the foundry. It is the consumer that has fostered the "good enough" philosophy in the foundry by not testing adequately and forcing the quality up. "Good enough" as applied to defroster knobs cannot be applied to aeleron hinges. The uses to which steel castings have been put in the past probably did not warrant the high testing costs; but in the present, and certainly in the future, reliability is the foremost requirement, and the quality of steel castings must be up-graded.

Properties of random samples of 17-4PH steel castings tested at the author's company have far exceeded the specification, and were found to be as good if not better than the wrought material. Such an experience makes one wonder why this is not done more often, rather than have one lot fall into the high strength, high ductility category, and the next be a dismal failure in one or both respects.

## PROCESS CONTROL

The answer is control. Control upon control is required from virgin metal through the last heat treating and cleaning operation. Little can be left to chance. In conversation with foundrymen, they have admitted this readily but qualify it by saying, "The controls which are required to give you what you want are too expensive for our operation to make a profit." Unfortunately, they are right. Most foundries are producing castings for the aerospace industry only as a side line, and the mainstay of their business is commercial production.

The foundry industry is an extremely competitive business, and in order for a foundry to be competitive its production costs must be cut to the bone. Unfortunately, the answer to this problem of economics is beyond the comprehension of this engineer. It has served, however, to make competition stiffer for the foundries, since some companies in the aerospace industry have opened their own foundries so as to get what they require from castings. Others have designed away from castings.

Control of the process from pattern to cut-off is paramount in achieving the desired end result. It is quite probable that the cost of instituting these process controls might not be as high as one might think. In order to institute controls, it is necessary to isolate the variables and to determine which are major in influence on the end result and which have only a minor influence. Control those which have been found to have the major influence. Through the trial and error method only mediocre high-cost results will be realized. Professor Howard F. Taylor of Massachusetts Institute of Technology summarized it well by stating, "It is not heretical nor unfair to observe that metal casters, taken as a group, have clung far too long to the belief that knowhow and experience are superior to science and technology."*

Experience and knowhow are important, but the technical, engineering approach must be combined with them if the casting requirements of the aerospace age are to be met. It is impossible to talk in terms of aerospace requirements without advancing the foundry technology. Reliability is the paramount requirement for success in any of the present aerospace products, and will be more so in the future.

## CONCLUSION

Foundries are going to have to accept the fact that the day of mass production of aircraft is rapidly fading, and with it large orders of castings. The future orders will be small, and the requirements will be tough to meet. It will also be necessary to resell many people on castings due to experiences they have had in obtaining what was required and actually promised in their part. Many designs which could be castings are being made by other means.

The fact remains that castings offer one of the biggest advantages to the Design Engineer to get the shape he requires without extra poundage from unnecessary metal impossible to remove by machining. It should be noted that many of the highest strength, high temperature alloys can only be produced in a usable shape by casting, since they are so tough they cannot be forged or rolled.

However, they will not be used unless they are produced to exacting quality standards. It is up to the casting industry to provide the aerospace designer and builder one of the most promising tools to get off the ground with, and that is a sound reliable casting.

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*_Foundry_ (Sept. 1960).

# PRECISION IN THE FOUNDRY

by A. Short


## ABSTRACT

**This paper discusses what is meant by precision when related to castings. The author believes it is generally accepted as meaning dimensional accuracy, but it should be related to surface finish, definition of detail, metallurgical soundness and general section thickness, all of which vary greatly with the alloy being cast and the method used or the process available in the foundry. The paper discusses methods available for obtaining the above broad definition of precision in alloys of magnesium, aluminum, iron and steel and refers to sand casting, die casting, the lost-wax process, shell molding and the $CO_2$ process.**


## INTRODUCTION

In the author's opinion there is no such thing as a dimensionally accurate casting, and no foundryman would warrant that a casting was accurate to the drawing in all respects. Many castings are held to within a close tolerance, and some firms are able to advertise truthfully that they can make close-tolerance castings in various alloys.

During the last decade the impression has been given or assumed that the investment casting process using expendable patterns would make precision castings. Nothing is further from the truth. Such castings are no more accurate than sand castings.

The author would concede one exception among all the methods and processes used for making castings —pressure die casting. Unfortunately, this is restricted in the number of alloys that can be used, but it is possible to obtain castings on a repetition basis of a high order of accuracy.

The term precision, when related to a casting, comprehends good surface finish, fine definition of detail, intricacy of design and close tolerance limits. It could be extended to cover the achievement of thin sections in alloys which are difficult to cast, to pass x-ray examination as well as any of a number of non-destructive tests.

There is no single process capable of making castings to this standard in weights ranging from ounces to tons in the multitude of alloys specified. The foundryman is left to chose the right process for the job, having uppermost in his mind that it must be the cheapest process compatible with the customer's requirements, whether they are of soundness and strength for engineering applications or high quality surface finish for the domestic market.

## DIFFICULTIES IN OBTAINING DIMENSIONAL ACCURACY

Some difficulties in obtaining dimensionally accurate castings are:

1) It is virtually impossible to make calculations at the drawing board stage which will forecast the metal contraction in all directions of a cast component. Ultimate accuracy is obtained by painstaking modifications to pattern or die equipment in relation to dimensional proof reports. Having modified his equipment to give an accurate casting, the foundryman has then to contend with the day-to-day variables in materials and conditions which will give rise to dimensional variations. An obvious example is the time lapse between casting and knocking-out, which will give varying physical properties as well as varying dimensions for the same casting. Others are the amount of pattern rap, metal temperatures and sand control.
2) Many alloys are subject to high temperature heat treatment to obtain optimum physical properties and any accuracy maintained by the foundry in the as-cast state is lost through distortion, erosion or growth.
3) The achievement of accuracy is expensive and no casting should be specified to closer tolerances than are functionally necessary.

As the designer becomes more appreciative of foundry problems, the making of castings for aero engines, for example, is made less difficult by the specification of minimum dimensions and maximum weights, or in the case of development engines target dimensions and tolerances, the final tolerances being

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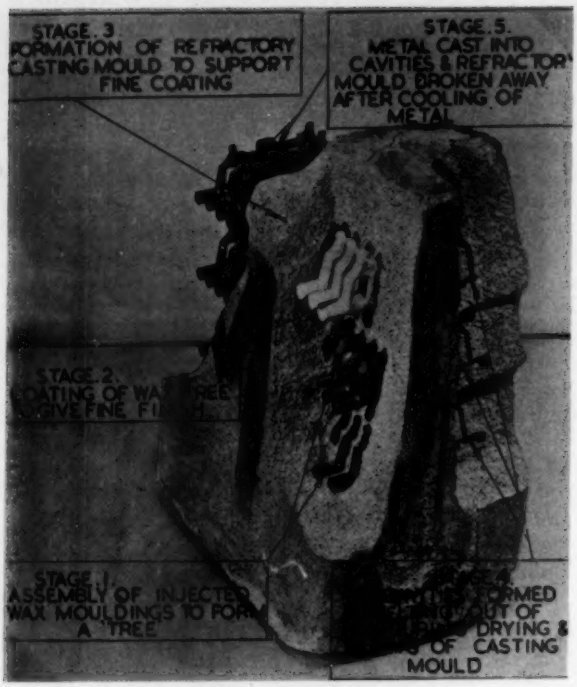



**Fig. 1 — Stages in the lost-wax process.**

specified for production only after foundry experience has been gained in casting the part.

It is not unusual to find that a particular part of a casting has to be thickened to make a component sound. The surplus material is then machined off to the dimension required by the customer. One wonders if it is an idle dream to hope that the foundryman will some day supply finished parts which would enable him to use his founding techniques to the best advantage and leave the machine shop to do the rest. Should this liberty be given to the foundryman he would probably consider the possibility of welding castings together to form an intricate design, or even coining or cold forming castings to obtain improved dimensional accuracy of specific dimensions and improved local properties. A new method available is spark erosion which will enable a component to be accepted for casting by ignoring a particularly difficult feature of the design in the cast state, and by subsequent processing to make it fully satisfy the customer's requirements.

## THE RIGHT PROCESS FOR THE JOB

Other factors inherent in a precision casting, namely, surface finish, definition of detail and soundness, are collectively associated with the alloy to be cast and the process selected for making the mold.

This paper refers to some of the newer methods of founding, particularly in relation to high temperature alloys.

A comment made by an American at a British conference last year was that sand casting had changed little since the days of ancient Egypt and would become obsolete in our lifetimes, and there are certainly many new techniques available to foundrymen developed since the latter part of the last war. An important point to bear in mind is that, although these new and sometimes unusual processes have opened up a new field for castings, they are, in all cases, much more costly than sand casting techniques and should never be used when a sand casting will suffice.

The new processes have been sold to purchase managers, production engineers and designers as precision processes, with dimensional accuracy as the predominant feature. In fact, castings by these processes are becoming accurate only by the pressure brought to bear on foundries by customers who demand what has been advertised as being reasonably possible. If the same effort is put in by sand foundrymen to match all the aspects of precision, then these new processes will eventually cater for a small proportion of casting requirements.

## INVESTMENT CASTING PROCESS

The lost-wax process has been recorded in many technical papers and it is not necessary to discuss it here. Figure 1 illustrates the steps in the process.

It is accepted that complicated, thin-section castings are in everyday production in alloys that are difficult to machine. Generally, such castings are made from a wax pattern formed in one piece from a collapsible metal die. Certain designs, however, possess a complicated internal shape, and even if a collapsible core can be designed the cost is sometimes prohibitive when related to the ultimate quantity required.

Three different methods are available to produce a wax pattern with a complicated internal configuration.

Figure 2 shows a simple example of the assembly of wax patterns. The waxes are sighted and fused together. By such jointing, however, accuracy and definition can be lost. Some firms use polystyrene injection moldings instead of wax which, because of their strength, permit interference-fit assembly on mating locations.

### *Split Die or Core Box*

A second method is by making a split die or core box of the internal shape and injecting a pattern in a suitably filled polyethylene glycol to arrive at a wax-like model or core of the internal shape. This core is then inserted into the die and the normal wax used in investment casting is injected. Polyethylene glycol is soluble in water (more readily if the filler is sodium carbonate and the water slightly acid), and by immersing the double-injected pattern in water the inner core is dissolved leaving a wax model having a complicated internal shape (Fig. 3).

A third method is to preform the core in a $CO_2$ bonded refractory material (Fig. 4). The core is inserted in the die, the wax injected, and, after assembling the wax runner system, investment is done in the usual manner. After de-waxing, the $CO_2$ core remains locked in the mold. The following $CO_2$ core mix is typical:



Fig. 2 — Assembled wax patterns.



Fig. 4 — $CO_2$ process cores.

| Material | Per Cent |
|---|---|
| Zircon granular | 80 |
| Zircon flour* | 16 |
| Sodium silicate | 4 |

*Mechanical analyses of these ingredients are:

| | BSS sieve | % retained |
|---|---|---|
| Zircon granular | 100 | 5-10 |
| | 150 | 70-80 |
| | 200 | 15-20 |
| Zircon flour | 200 | < 20 |

The bond of this core mixture starts to break down at temperatures over 1500 F. If it is desirable to raise the temperature to, say, 1800 F to enable particularly thin sections to be run, the core should be soaked in a solution of hydrolysed ethyl silicate or a solution of magnesium chloride and fired before placing in the wax pattern die. This will strengthen the silicate bond given by the $CO_2$ process and render the preformed core stable at temperatures up to 2000 F.

The above three methods are now in regular use. Variations in the refractory preformed core are being developed to obtain improved surface finish and better accuracy. These include slip cast aluminum silicate bonded with refractory cements and pressed, and sintered aluminum oxide.

A further development of the investment casting process has been made in the foundry with which the author is associated for the production of large thin section castings in highly alloyed steels for incorporation into fabrications. Faced with having to make costly metal dies with which to experiment, an alternative method of producing wax patterns cheaply and quickly was sought. It appeared possible to take an assembled sand mold from the sand foundry and to pour wax into it. The bond in the sand, however, prevented the removal of the wax pattern without damage, so a special $CO_2$ bonded sand containing sodium



Fig. 3 — Soluble wax cores.

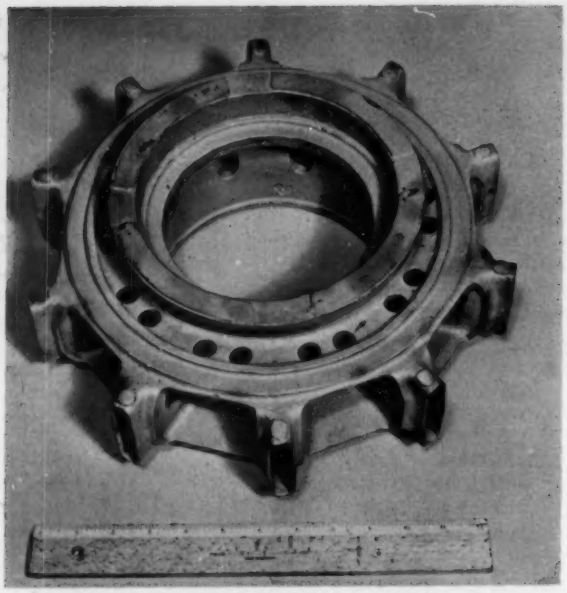


Fig. 5 — Casting produced from wax pattern made in a water dispersable sand mold.

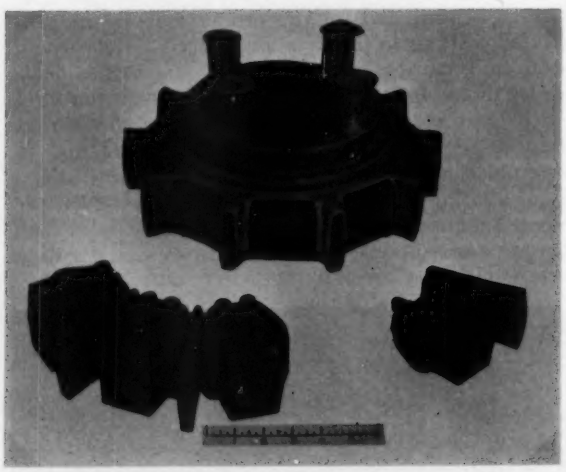


Fig. 6 — Wax pattern made from sand core assembly.

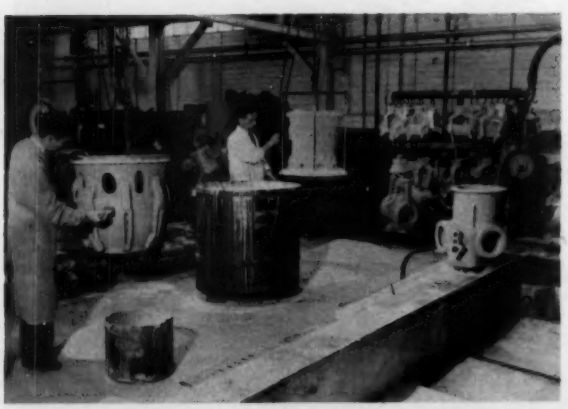


Fig. 7 — View of investment shell molding department.

carboxymethyl cellulose and sodium carbonate to make it water dispersible was developed. It was, therefore, possible to make wax patterns of virtually any size, the ultimate accuracy depending upon the skill of the patternmaker and the operator assembling the cores and molds.

A typical mix of water-dispersible sand is:

| Material | Per Cent |
|---|---|
| Redhill 'H' sand* | 91½ |
| Zircon flour | 1 |
| Sodium carbonate | 2½ |
| Sodium carboxymethyl cellulose | 5 |
| Sodium silicate | 8 |

*Mechanical analysis:

| BSS sieve | % retained |
|---|---|
| 100 | 10-15 |
| 150 | 25-35 |
| 200 | 25-35 |
| Pass 200 | 15-20 |

This method of making a wax pattern was the result of having a problem to be overcome, but the development opened up a commercial market for the supply of alloy steel castings which are large and heavy but thin.

### *Semi-Permanent Dies*

While discussing temporary means of obtaining quick wax patterns for trials, it is worth mentioning at this point that on many occasions normal $CO_2$ bonded sand molds or plaster dies have been made from wood patterns and used as semi-permanent dies for wax injection. This is a quick method of obtaining wax patterns for normal investing where the component is solid or has simple internal coring.

In handling such large wax patterns, it is not usual to make a solid investment mold from both the viewpoints of both weight and economy. It is preferable to build up a shell of investment by alternate dipping in a bonding solution and dusting with the finely divided refractory until five or six coats have been applied to give a shell thickness of ⅛-¼-in. (Fig. 7).

Many investment foundries now have a shell investment production line. It has the great advantage of overcoming the problems of bad packing associated with solid molds. These are vibrated to consolidate the investment around the pattern and result in bad packing on the underside of horizontal faces, while it uses less of the expensive mold refractories and binders. However, where components are relatively small with a small surface area/volume ratio conductive to close clustering on the runner system, the solid mold will be cheaper from the labor content aspect while consuming little more material.

Both solid and shell investments lend themselves to a high degree of mechanization and even near-automation. The investment casting process is probably the most universal inasmuch as almost any alloy can be successfully cast. The maximum weight is probably about 75 lb in the present state of the process, although castings have been made for development weighing well over 150 lb.

For general applications in commercial alloys, investment casting is not the best process. For example, it cannot compete in aluminum alloys with per-

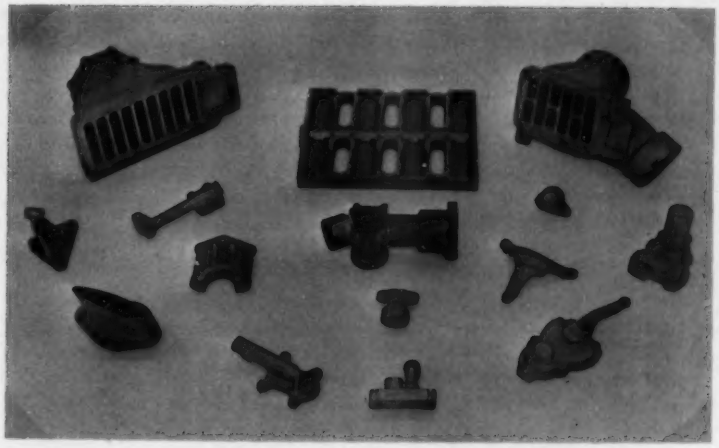


Fig. 8 — Steel castings made by $CO_2$ process.

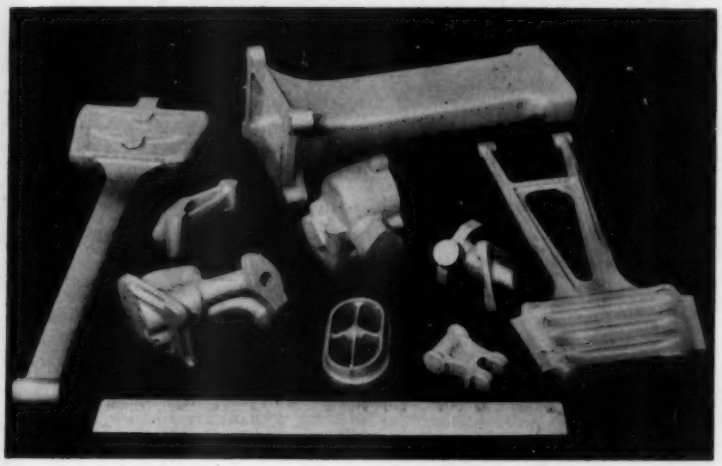


Fig. 9 — Steel castings made by Shaw process.

manent mold castings for price, strength of material or accuracy. In the production of castings for the space age, investment casting is ideal for alloys which have to be cast in vacuum or a controlled atmosphere, and where the mold and crucible must not contain any siliceous material.

The *Investment Casting, Engineering and Design Manual* published by the American Investment Casting Institute gives comprehensive data on tolerances for investment castings. It offers ± 0.005 in./in. on functional areas and requests from ± 1/64-in. on small dimensions to ± 1/16-in. on dimensions of over 6 in. on nonfunctional areas. This partly confirms the original comment that precision casting is no more accurate than sand casting.

## SHELL MOLDING, SHAW PROCESS AND THE $CO_2$ PROCESS

Three processes cater for a particular market requring castings which lie between precision investment castings and normal sand castings.

The $CO_2$ process, the Shaw process and shell molding are illustrated respectively in Figs. 8-10. The essential difference between these processes is the method and materials used in making the molds.

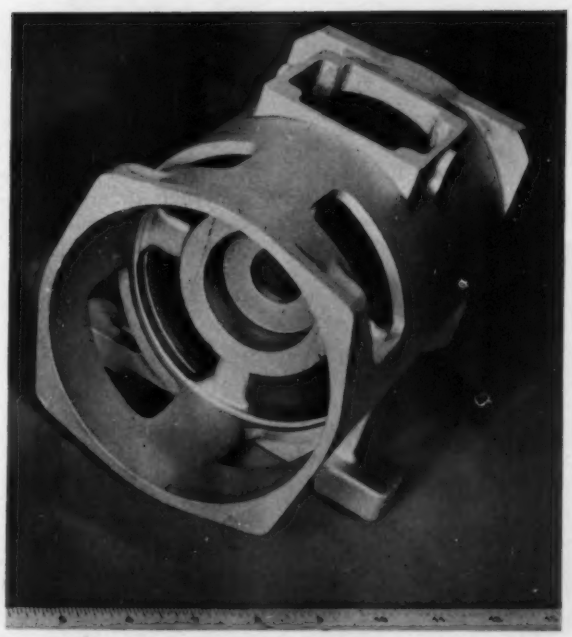


Fig. 10 — Aluminum casting made by shell ('C') process.

$CO_2$ sands are invariably synthetic—a clay-free sand is mixed with 2-4 per cent of sodium silicate and rammed into the mold or core box in the usual manner. Before removing the pattern or core box $CO_2$ is blown through the sand, and a chemical reaction occurs which precipitates a silica gel from the sodium silicate thereby bonding the sand. The application of carbon dioxide takes a few seconds at a low pressure, i.e., 5-15 lb./in.,$^2$ depending upon the size and shape of the mold or core. $CO_2$ is the preferred agent for hardening or solidifying the mold or cores, but the reaction can be accomplished by other agents in liquid or gaseous form.

They include ammonia, hydrochloric acid, methyl and ethyl alcohols, acetone, sulfur dioxide and nitrous oxide, but the use of these alternative materials involves certain hazards. After the gassing operation, the sand shape is immediately handleable and in most applications is ready for assembling and casting. The process can be used for casting most alloys. However, when casting nickel-base or other high temperature alloys, the silica sand is replaced by a finely divided refractory such as zircon.

The Shaw process uses similar materials to the investment technique, i.e., a refractory such as sillimanite with ethyl silicate-alcohol binder. The mixed slurry is poured over the pattern and fast, controlled gelation is affected by the addition of ammonium carbonate. The mold gels in a matter of seconds, and when stripped off the pattern it is ignited. When cold, the dried mold is assembled with cores made in the same manner.

### *Shell Molding*

Shell molding, sometimes called the 'C' process uses clay-free sand mixed with about 3 per cent of urea or phenol formaldehyde cured under heat while in contact with the pattern. In operation an all-metal pattern or core box is maintained at a temperature around 500 F. The dry mixed sand and binder is dumped on to the hot pattern. The mixture starts to cure from the surface of the pattern, and a predetermined time cycle allows a given thickness of shell to be built up before inverting the pattern to allow the surplus sand mix to drop off. Curing is then completed by putting the pattern and shell in an oven at 500 F. The time cycle for making a shell is 60-75 s. The resultant shell mold looks like a large biscuit with a smooth faithful definition of the pattern on one side and a rough finish on the other, with a thickness of 1/8-1/4-in.

The author purposely groups these techniques together as a number of foundries operate the three processes in the same department. Careful analysis is needed to determine why one of the processes should not be used to produce all types of castings in these foundries but, nevertheless, each has a certain advantage over the others.

The $CO_2$ and the Shaw processes are similar inasmuch as they can both use loose wood patterns. Whereas the $CO_2$ process is cheaper, the Shaw process gives a better surface finish and will produce thinner castings by virtue of the fact that the mold, when assembled, can be raised to a high temperature.

Shell molding is the most economical where quantities warrant all metal pattern equipment being made, bearing in mind that the resins used in this process break down at a much lower temperature than either the sodium silicate of the $CO_2$ process or the ethyl silicate of the Shaw process. Therefore, with certain high temperature alloys and stainless steels, mold metal reaction occurs which gives a rough surface, or local contamination of the alloy, or both.

These three processes assemble molds and cores made from pattern equipment as in normal sand-foundry practice, therefore they exhibit dimensional accuracy of the same order. The author's opinion is that these three processes are used mainly by sand foundries which do not employ lost-wax methods, as a natural extension of foundry sand methods.

No skilled labor is required in operating these processes, the technical knowledge required being provided by the same type of person as controls an investment casting foundry. It is therefore considered advisable for a foundry management to investigate these methods which as yet do not fall into a definite category, because they open up a new market in steel castings which are neither sand nor precision.

Of course, $CO_2$ and shell molding are now established production processes used in aluminum, magnesium, bronze and iron foundries either as complete molds and cores or as cores for assembly into standard sand molds.

## PLASTER MOLDING

Probably the best known form of plaster molding is the Antioch process, developed in the 1930's. Plaster of paris, plaster molding, gypsum molding and the Antioch process all use the same basic material, i.e., calcium sulfate.

In the research and development stages it was noted that, if the mold formed by the setting of the slurry were dehydrated and allowed to rehydrate without being disturbed, gypsum crystals would slowly reform throughout the mold mass as granules about the size of sand grains. These have sufficient intergranular strength at points of contact for the mass to retain its cohesion at all times and to acquire a high permeability to gases. The transformation does not take place on the surface of the mold because not enough water is present, so that the surface remains smooth while the remainder becomes granular and permeable.

The calcium sulfate is mixed with water to form a slurry which is poured over the pattern and sets solid in about 5 min. To have free water available for the hydration into granular form, it is advantageous to hydrate the green mold in an autoclave or an atmosphere of saturated steam. This can be done at about 15 lb/in.$^2$ in 6-8 hr for the average mold. The mold is allowed to stand at room temperature for about 14 hr, during which time the hemi-hydrate slowly rehydrates by combining with free water to become granular.

The mold is then dried at temperatures of about 450-475 F, and if the molds are allowed to reach oven temperature all the free water and most of the chemically combined water in the calcium sulfate is re-

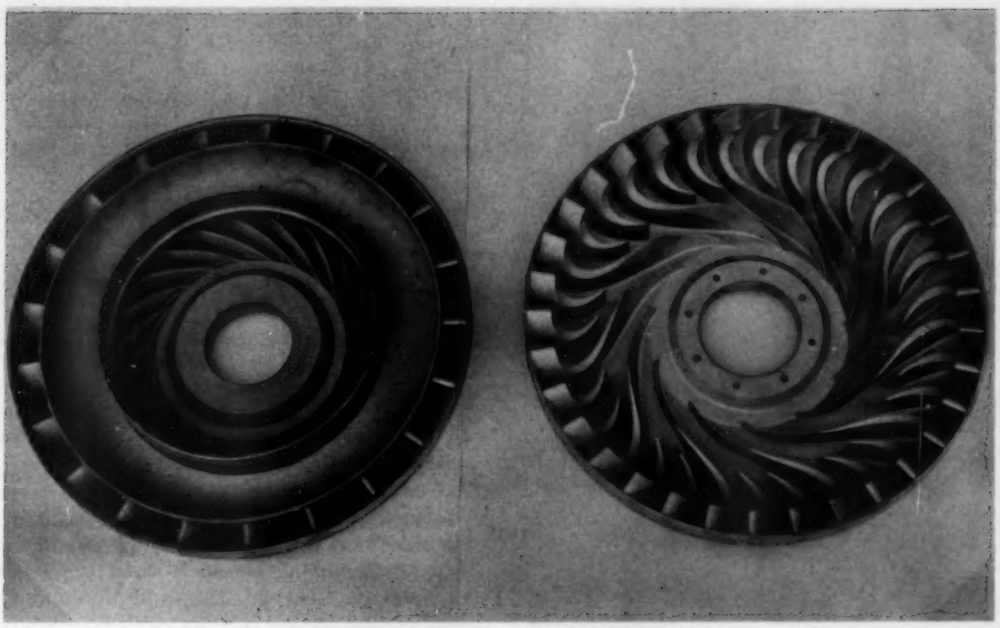


Fig. 11 — Aluminum castings by Antioch process.

moved. Up to 50 wt. per cent of finely divided sand can be added to the gypsum mix without interfering with the transformation of the mold structure into the permeable form.

During the process an expansion occurs which, if controlled, is in the order of one per cent or slightly less. This takes place while the mold is being dehydrated and rehydrated, and can be controlled by small additions of sodium silicate and Portland cement to give, in practice, between 0.1 and 0.25 per cent expansion.

The outstanding advantages of plaster molding are the surface finish and dimensional accuracy. The molding materials are, of course, more expensive than conventional sand casting materials, and where an ordinary sand casting will do the job it should always be used. In comparison with die casting, plaster molding is more versatile in the sizes and shapes which can be cast and, as the tooling costs are cheaper, the Antioch process will compete in price with die castings on medium quantity runs.

It does not overlap the investment casting or shell casting field to any great extent, and is restricted in the type of alloy which can be cast. For intricate shapes in aluminum-base alloys such as tyre molds, impellers, rotors, torque converters and wave guides, it is now well established (Fig. 11). One can expect a surface finish on aluminum and copper-base alloys by plaster molding of 30-40 $\mu$ in. (AA).

## THE McCANNACORE PROCESS

An interesting aspect of the art of making castings in aluminum and magnesium is the possibility of obtaining small intricate unlined cast holes (Figs. 12 and 13). For magnesium castings, the McCannacore process allows complete freedom of design, thereby saving weight, material and costs.

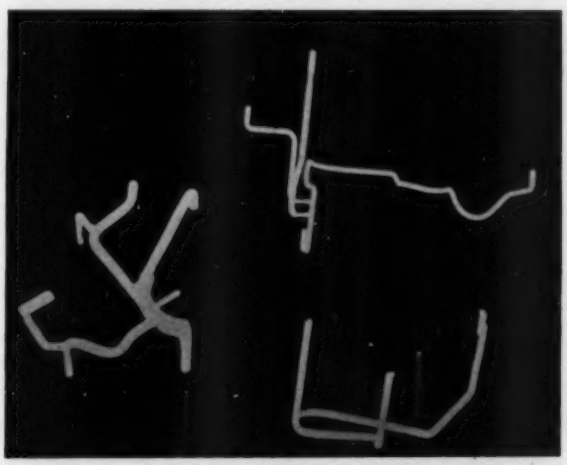


Fig. 12 — Intricate cores for McCannacore process.

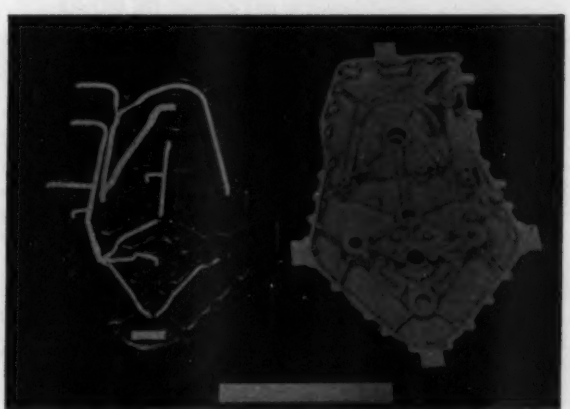


Fig. 13 — Intricate holes made by McCannacore process.



Fig. 14 — Iron die casting.

Briefly the process is to shape and bend heat resisting glass tubing to the desired configuration, and then to coat the assembly with a tungstic oxide wash, assemble in a sand mold and cast.

The completed casting is immersed in hydrofluoric acid to dissolve the glass tubing. To ensure that the passageway is properly cleaned out, the casting is x-rayed. As tungstic oxide is opaque to x-rays, its absence from the radiograph indicates that the glass has been completely dissolved.

A method for aluminum castings is the Dalton process which uses copper tubing bent to the required shape. The tube is then sheathed in fibreglass or braided steel wire. After casting, the copper is removed by reaction with nitric acid and the sheath pulled out, to leave a clean cast-in hole.

## DIE CASTING

In discussing this process, the great strides being made in this field must not be forgotten. Figures 14-18 show, respectively, die castings in iron, magnesium, aluminum (vacuum process), aluminum and bronze.

The American Die Casting Institute through its research foundation has completed the development of a flexible vacuum method for use in the die casting process. Its salient features are said to be:

1) Only the die cavity is evacuated.
2) Vacuum is maintained throughout the period of metal injection.
3) Pressures of 25 in. Hg, measured in the cavity, are consistently maintained without increasing cycle time.
4) The method can be conveniently adapted to provide vacuum feeding to the shot sleeve.

This is performed with a one hp pump supplying two average die casting machines, and with little modification to existing die equipment.

One of the greatest drawbacks to the use of pressure die castings in stressed applications has been the occluded and entrapped air in the casting which renders the component suspect from the viewpoint of reliable consistent physical properties, plus the restriction on the use of heat treatable alloys. One of the difficulties in having to heat treat a pressure die casting is that the temperature to which the casting has to be raised to obtain optimum physical properties causes the occluded air to expand and blister or even break through the surface of the casting. The new vacuum technique, however, will deserve a close examination to see how pressure die casting applications can be widened as more details of properties, etc., become available.

Slush casting is well known, being the method used to produce hollow castings where the internal finish or shape is not important. A domestic example is the aluminum teapot and kettle spout. An open metal die is filled with molten metal—the metal starts to solidify on the surface of the mold and, by experience, after a time lapse of a few seconds the mold is inverted and the still molten interior is allowed to run out. The section thickness is determined by the skill of the operator.

In recent months, an automated version of this has been announced, whereby the metal is injected into

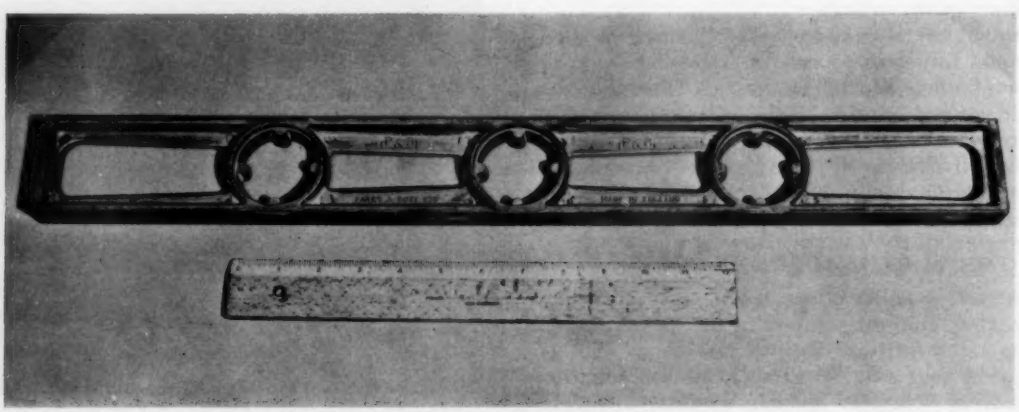


Fig. 15 — Magnesium pressure die casting.

the die and, after a predetermined time, the liquid interior is run off. Because this is all carried out on a controlled time cycle it is probable that this method may have engineering applications as it is said that consistent section thickness can be obtained.

## SAND CASTING

Having taken a brief look at these newer foundry techniques, it must not be forgotten that sand foundries with their semi-automatic and automatic plant still produce the cheapest castings. If the foundry maintains adequate controls and combines certain processes, such as $CO_2$ cores or shell cores in green sand molds, and even the use of the new oil-dispersable bentonites which permit green sand molding in the absence of water, there need be no fear of drastic competition from the newer processes.

The new processes, being expensive, must offer something special. The author hopes to prove it is not accuracy, but surface finish and section thickness, and it should not be beyond the ability of foundry technicians to introduce these two selling points into the sand foundry.

An example of a sand casting which comes within the definition of precision is shown in Fig. 19. This is the inner support for an early type of axial-flow jet engine in an acicular iron requiring minimum physical properties of 52,000 lb/in.$^2$ The two arms with feeders have a top and botton wall thickness of 0.400 in. with side walls of 0.250 in. The remaining six arms have a general thickness of 0.150 in. The circular internal wall has a thickness of 0.125 in. The casting must, of course, be free from defects after examination by x-ray and dye penetrant.

Another example is the air intake casing for a turboprop engine, shown in Fig. 20. This is cast in a magnesium-base alloy containing zirconium and rare-earth metals for high temperature application. The casting contains 43 cores, some of which are shown in Fig. 21. The lighter-colored cores are $CO_2$ process cores, and the darker ones are made with the traditional oil and cereal binder. Over 90 per cent of all coremaking in the author's magnesium foundry is by the $CO_2$ process.

Figure 22 shows an aluminum-base alloy containing 5 per cent copper, 1.5 per cent nickel and 0.2 per cent of zirconium. However, in certain higher temperature applications this component has to be fabricated in steel. This casing is shown in Fig. 23, which indicates the cast steel facings, etc., which are welded on to the sheet metal.

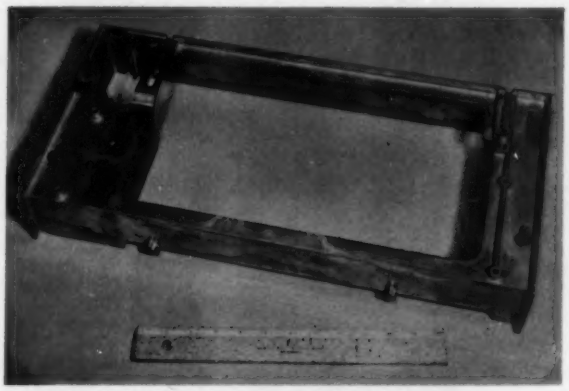


Fig. 17 — Aluminum pressure die casting.

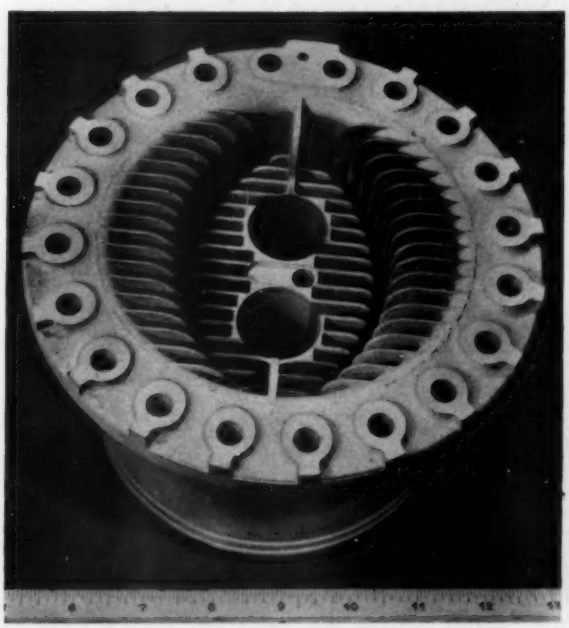


Fig. 16 — Aluminum vacuum die casting.

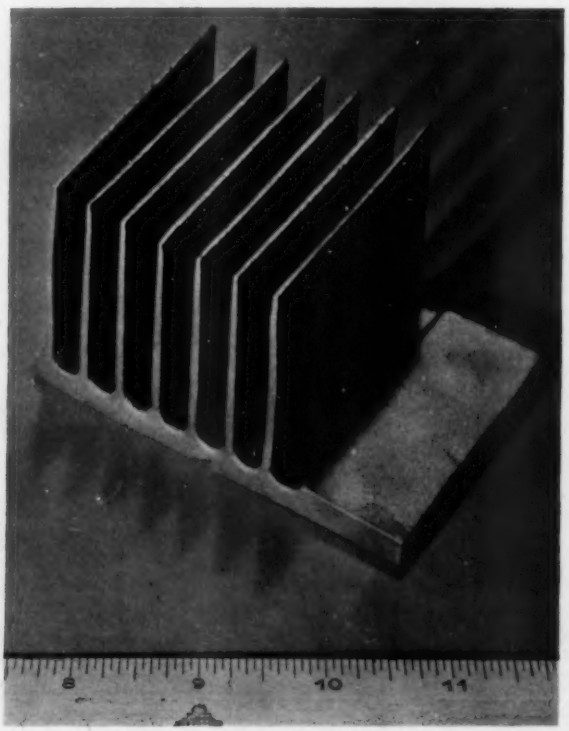


Fig. 18 — Bronze gravity die casting.

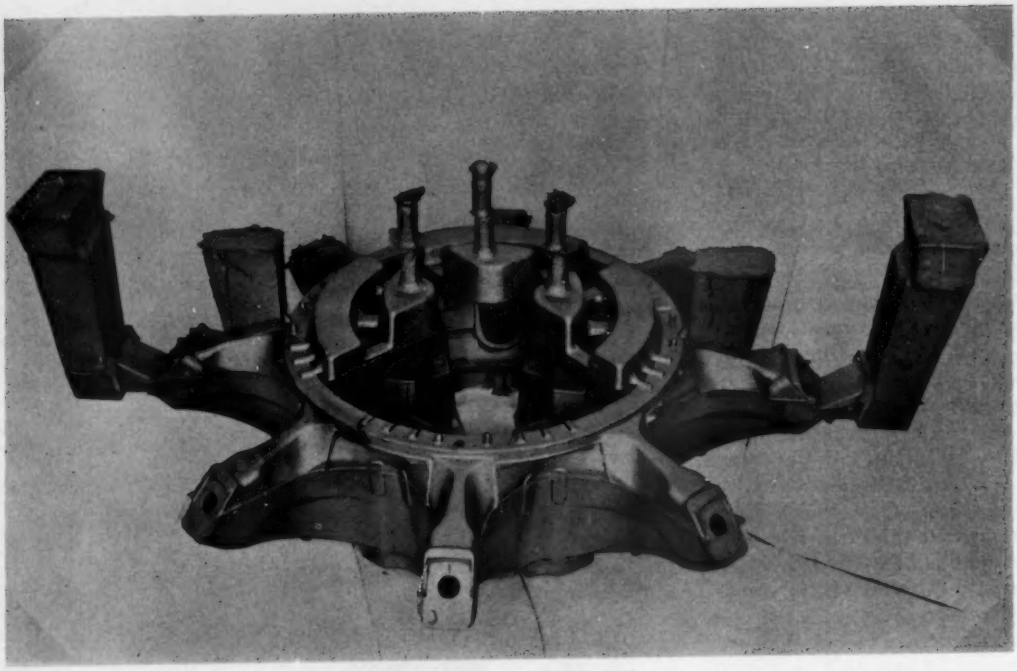


Fig. 19 — Support casing in iron.

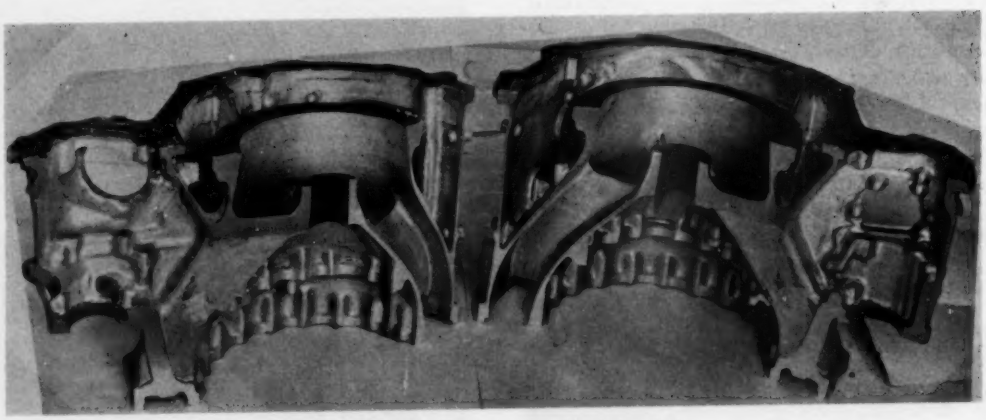


Fig. 20 — Air intake casing in magnesium.



Fig. 21 — Some of the 43 cores for casting shown in Fig. 20.

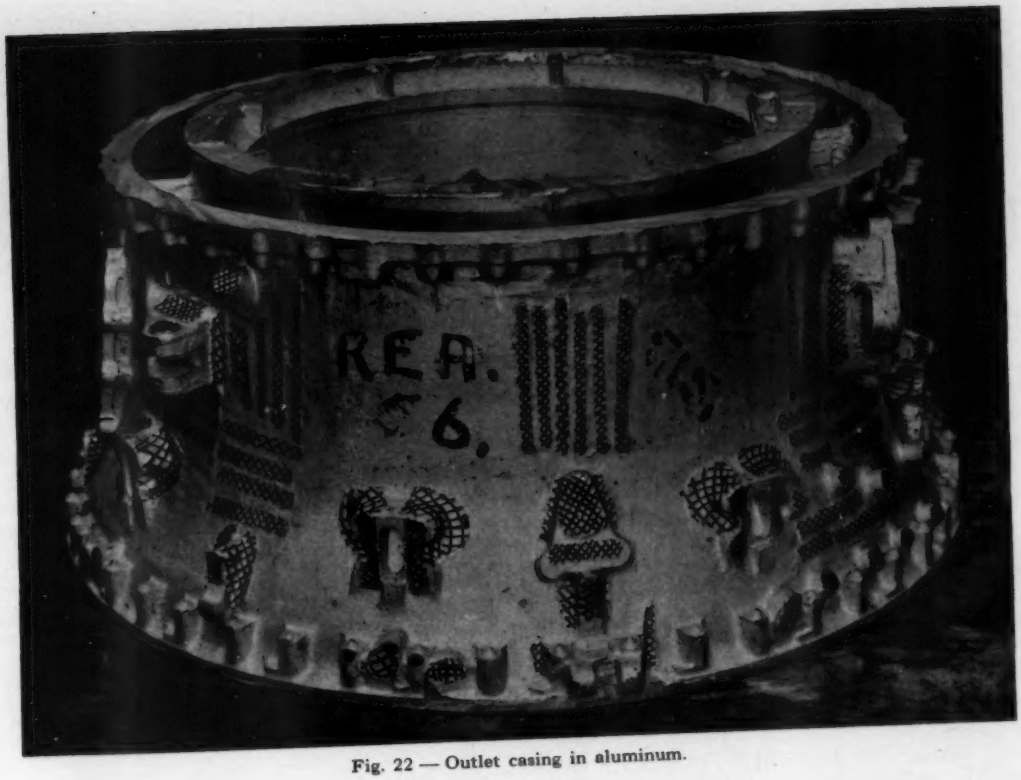



Fig. 22 — Outlet casing in aluminum.

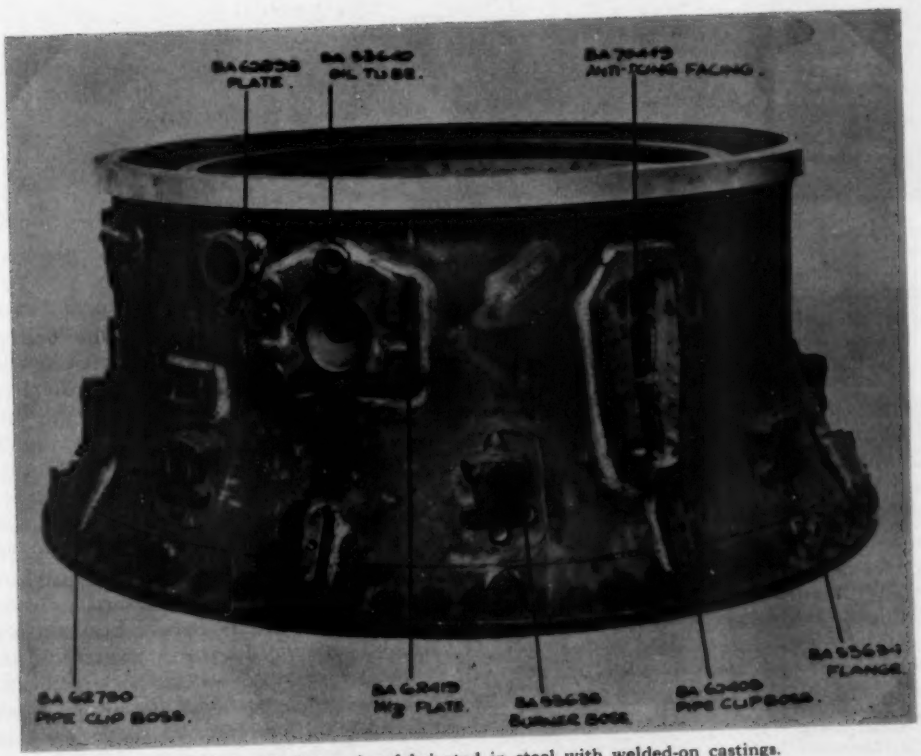



Fig. 23 — Outlet casing fabricated in steel with welded-on castings.



Fig. 24 — Epoxy resin dies.

## SURFACE FINISH

Almost every casting receives some form of surface finishing, even if only rough shot blasting to give a matt finish and camouflage defects. On the other hand many castings require an excellent finish for subsequent plating and polishing.

It is always desirable, therefore, to be prepared to go to some trouble in the foundry to obtain as good a finish as possible in the as-cast state to reduce subsequent finishing costs.

Some industries now specify surface finish in microinches. It may be quite foreign to the average foundryman to think of supplying castings to a surface roughness expressed in so many millionths of an inch, but this has to be accepted in the new age of founding. Numerical assessments of the average height of the irregularities constituting surface texture are expressed as the arithmetical average (AA). The British term for the same value is center-line average (CLA).

Taking some examples of castings by various processes some trials were made to find out what could reasonably be expected on as-cast surfaces, the results being as follows:

| Process | μ in. (AA) | Base Alloys |
|---|---|---|
| Plaster mold | 30-40 | Aluminum and copper |
| Investment | 40-50 | Steel |
| Permanent mold | 80 | Aluminum |
| Shaw | 80 | Steel |
| Shell | 100 | Magnesium and aluminum |
| Sand | 200-300 | Aluminum |
| | 150-300 | Magnesium |
| | 250-450 | Iron |

Naturally, the grain size of the refractory used makes a great difference to the as-cast finish, as also does the alloy to be cast, which, by mold metal reaction with the binders or refractories, can give a rough finish.

With an aluminum alloy containing 2 per cent silicon, 2 per cent copper, 0.1 per cent magnesium and 1 per cent nickel, the following maxima and minima on three castings were obtained.

| Mold | μ in (AA) |
|---|---|
| Natural sand (3% moisture) | 300-500 |
| Natural sand (Stove dried) | 290-420 |
| Silica sand ($CO_2$ bonded) | 150-320 |
| Silica sand (Fine grain, bonded with oil-dispersable bentonite) | 160-185 |

The foregoing figures are not comprehensive and, this could be a subject for further investigation.

## DIMENSIONAL ACCURACY

From published literature some companies offer the following dimensional accuracy for the newer processes on linear dimensions.

| | | |
|---|---|---|
| Shell ('C') | ± | 0.010 in. up to 6 in. max. |
| | ± | 0.015 in. between 6 in. and 20 in. |
| Shaw | ± | 0.005 in. up to 3 in. |
| plus | ± | 0.0015 in./in. additional. |
| Plaster | ± | 0.003 in. up to 3 in. |
| plus | ± | 0.001 in./in. additional. |
| Investment | ± | 0.005 in./in. |
| | ± | 0.062 in. over 6 in. |
| $CO_2$ | ± | 0.010 in. up to 6 in. |
| plus | ± | 0.0015 in./in. additional. |

It is not usual to find a sand casting varying more than 0.075 in. on the major aluminum and magnesium castings used in aero engines, but if this figure is converted over a 40 in. dimension, the tolerance is in the order of ± 0.001 in./in. which is better than tolerances offered by the newer processes.

## PATTERN AND DIE MAKING

In those sections of the industry making dies and patterns for use in the foundry, great strides are being made by the extended use of epoxy resins.

In the lost-wax process the epoxy resins have opened up a new field in competition with tin-bismuth alloy, as they can be cast from wood models. In the hands of a good patternmaker, the time to produce a given model and cast the epoxy die can be as little as one quarter of the time to cut the same die in steel in a toolroom (Fig. 24). Of course, in addition, drawing office time is completely eliminated as a patternmaker works from the component drawing.

An additional help in making epoxy dies from wood models is the introduction of accurate sheet wax having a self-adhesive backing (Fig. 25). When making a die having a general thickness of, say, 0.075 in. over a developed form which constitutes the bulk of the component, say, a pipe or casing, a model is made of the internal shape and core boxes are cast. Then, by wrapping the original model with self-adhesive wax sheet, a model is obtained of the exterior shape for casting the die, such features as

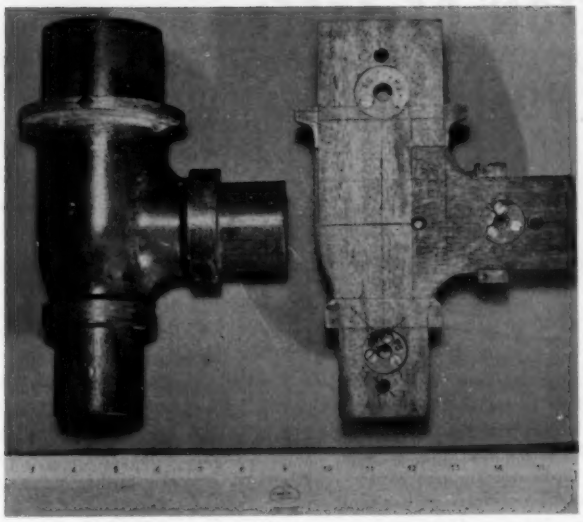


Fig. 25 — Use of sheet wax patternmaking.

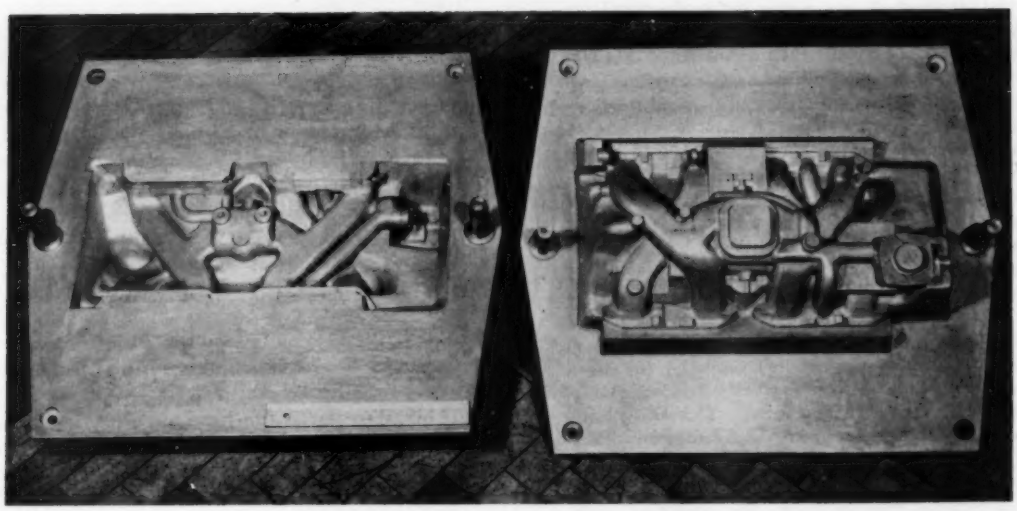


Fig. 26 — Fiberglass reinforced resin pattern.

exterior facings and bosses being added at the waxing-up stage. Similarly the patternmaker is increasingly using fiberglass-reinforced resin. Figure 26 shows an automobile induction manifold pattern made in aluminum powder filled epoxy resin reinforced with fiberglass.

## CONCLUSION

The introduction to this paper mentioned that enormous strides had been made in the foundry industry during the last decade. This development is hardly begun, since the space age requires the casting of metals such as molybdenum, tungsten, tantalum and niobium which will demand clinically clean foundries, with controlled atmospheres in areas where the castings are made, and high technological standards.

Nevertheless, precision as broadly defined already exists, particularly when one pauses to realize that castings are being made today that fulfill the following requirements:

1) Close tolerance with range specified. In some instances minimum dimensions are coupled with maximum weight.
2) Surface finish specified in micro-inches.
3) Pressure tightness in air pressure or hydraulic test.
4) Soundness by radiological examination.
5) Sub-surface soundness by penetrant dye test.
6) Physical properties by testing samples of each heat of metal.
7) Chemical analysis by sampling.

The author is conscious of the many omissions in this paper which go toward the making of precision castings such as inherent contraction characteristics of the metal, mold and core uniformity, mold pressures, gating and feeding practices, equipment and casting design. However, it is hoped that the claims made for new processes have been put in their correct perspective when related to sand castings, particularly from the viewpoint of precision.

# SOLIDIFICATION OF STEEL CASTINGS AND INGOTS

by M. C. Flemings, R. V. Barone, S. Z. Uram and H. F. Taylor

## ABSTRACT

This paper summarizes results of research conducted at Massachusetts Institute of Technology on solidification of high strength low alloy steel. The program was under sponsorship of Army Ordnance through Watertown Arsenal,[1,2,3] with its main emphasis on examining (1) different methods of obtaining directional solidification in cast steel and (2) effects of solidification variables on structure, segregation and properties of the cast steel.

A.I.S.I. 4330 and A.I.S.I. 4340 were studied. Electrolytic iron and other high purity charge materials were used to assure low sulfur content and freedom from impurities. Melting was by air induction. The majority of solidification experiments were on plate-like and cylindrical castings resembling ingots in shape. Solidification was controlled so the castings would freeze under a variety of thermal gradients. Control was obtained by using different molding materials and by using composite molds.

For example, plate castings were molded horizontally with an end riser in different types of molds including (1) zircon sand with a metal chill at the end of the plate, (2) silica sand chilled similarly, (3) silica sand without chill, (4) ethyl-silicate bonded mullite without chill, room temperature, (5) ethyl-silicate bonded mullite without chill, preheated to 1250 F (677 C) and (6) ethyl-silicate bonded mullite preheated to 1600 F (871 C).

The cylinder castings were also solidified under various thermal conditions, using mold materials similar to those that were used for the plates. They were cast (vertically) into different types of molds including (1) sand mold, (2) sand mold with bottom chill, (3) hot ethyl-silicate bonded mullite mold with bottom chill and (4) mold of exothermic moldable mixture with bottom chill. In the castings of item (4) above, essentially all heat was extracted through the chill and solidification under these conditions has been termed "unidirectional."

## INTRODUCTION

A major portion of this work was devoted specifically to the effect of solidification variables on microporosity. Microporosity (porosity so small as to be invisible to the naked eye except, perhaps, on a carefully polished surface) has been shown to be an important factor limiting properties of cast nonferrous alloys,[4,5] cast steel[6-9] and other metals. Low gas contents and steep thermal gradients during freezing are necessary to completely eliminate microporosity; in certain nonferrous alloys gradients greater than several hundred degrees Fahr./in. are required, and these are obtained only quite near a chill (or metallic mold wall).[4,5]

Microporosity in steel castings and ingots, when present, can usually be discerned by careful metallography. However, a more sensitive, and usually simpler, technique is by radiography of a thin section cut from the casting or ingot. The thin section may be as much as ⅛-in. in thickness, or it may be much thinner; better resolution is obtained with thinner samples. It is possible, conveniently, to mill rather extensive sections to approximately 0.020 in. in thickness and radiograph this section; several recent studies on steel have been carried out using this type of "semi-microradiography."[9,10]

Small pieces can be reduced to the order of 0.005 in. or less in thickness; this type of thin section radiography is usually termed "microradiography." Trillat[11] has reviewed the fundamental aspects of microradiography and described its application in many different types of investigation.

In this work, both semi-microradiography and microradiography were employed to determine location and amount of microporosity in castings produced. Factors other than microporosity were also examined and related to solidification behavior; these included grain size and type (columnar, equiaxed), dendrite spacing, inclusion size and distribution and macrosegregation. Mechanical properties (in the heat-treated condition) were obtained from the various castings produced. Details of experimental procedure not included herein may be found in earlier related papers and reports;[1-3,12,13] these publications also describe work on various solidification studies not reported here.

## FLAT PLATE CASTINGS

A series of test plates one in. thick was cast employing the mold design shown in Fig. 1. The distance from the end of the plate to the edge of the riser was slightly over 4 in. An exothermic sleeve was employed around the riser on all castings. Some castings were chilled as shown, and others were cast without a chill. Different types of mold materials were

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employed to obtain solidification under a variety of different thermal conditions (silica sand, zircon sand, ethyl-silicate bonded mullite preheated to several different temperatures).

### *Thermal Gradients and Microporosity*

Thermocouples were placed at the centerline of each plate casting one, 2, 3 and 4 in. from the end of the plate, and cooling curves taken during solidification. From these curves thermal gradients along the centerline of the plates could be plotted, Fig. 2. These are average gradients in the liquid-solid (mushy) region during solidification; details of the calculation of this curve have been given earlier.[1,12]

In all plates produced, a positive thermal gradient existed along the centerline of the plate. Thus, solidification in all plates was directional from the plate extremity towards the riser. However, directional solidification is not sufficient to assure a completely sound casting in most metals. Adequate thermal gradients must also be present during solidification. These gradients must be steep enough to maintain open feed channels so liquid metal from the riser can flow easily through the interstices of the growing dendrites to accommodate solidification shrinkage.

In each of the plate castings examined, gradients were highest near the end of the plate farthest from the riser and decreased towards the riser. Gradients probably increased again immediately adjacent the riser,[14] but in most castings not enough thermocouples were used in this location to detect such a change. Thermal gradients near the end of the plate increased with increasing ability of the mold to extract heat—for example, gradients were greatest in the casting poured in the zircon sand mold, end chilled; they were least in the casting poured in an ethyl-silicate bonded mullite mold, heated to 1600 F (871 C).

To determine if gross porosity was present in any of the plates cast, they were first radiographed by ordinary nondestructive techniques; no porosity was evident. Slices were then cut from selected plates, ground to ⅛-in. and radiographed; again no porosity was discerned. Finally, the slices were ground to 0.020 in., re-radiographed, and microporosity was found in all castings examined. Results of three of these radiographs are sketched in Fig. 3. In the sand castings (castings a, b and c), microporosity was most severe in cases where thermal gradients during solidification were low. Some porosity was evident in each of these castings (chilled and unchilled) in areas where thermal gradients along the centerline fell below about 150-200 F/in. Surprisingly, porosity in one casting examined that had been produced in an ethyl-silicate bonded mullite mold appeared to be less than that in the unchilled sand casting. There is no apparent reason for this, and further work is necessary to confirm the observation.

The plate castings described above were poured of 4330 alloy. Test coupons were cut from the plates parallel to the chill edge, heat-treated to the 200,000 psi ultimate strength level and tested. Ultimate tensile strengths and yield strengths obtained were not dependent on bar location or on type of mold employed; however, elongation and reduction in area were sensitive to these parameters (particularly to bar location). Figure 4 is a typical plot of elongation versus distance from the plate edge for plates cast in sand molds. Elongation in all plates was lowest in the region of 3-4 in. from the riser; it was in this region that microporosity was greatest.

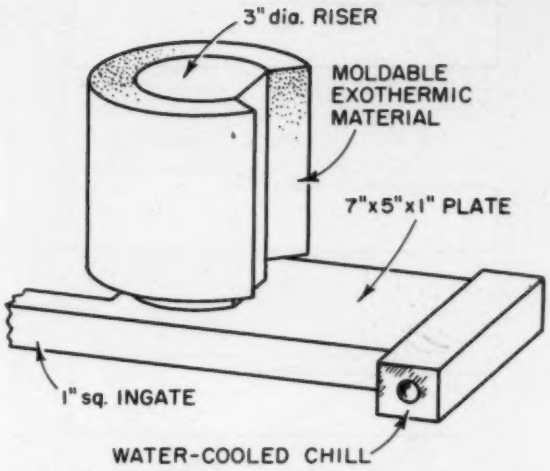



Fig. 1 — One in. plate pattern.

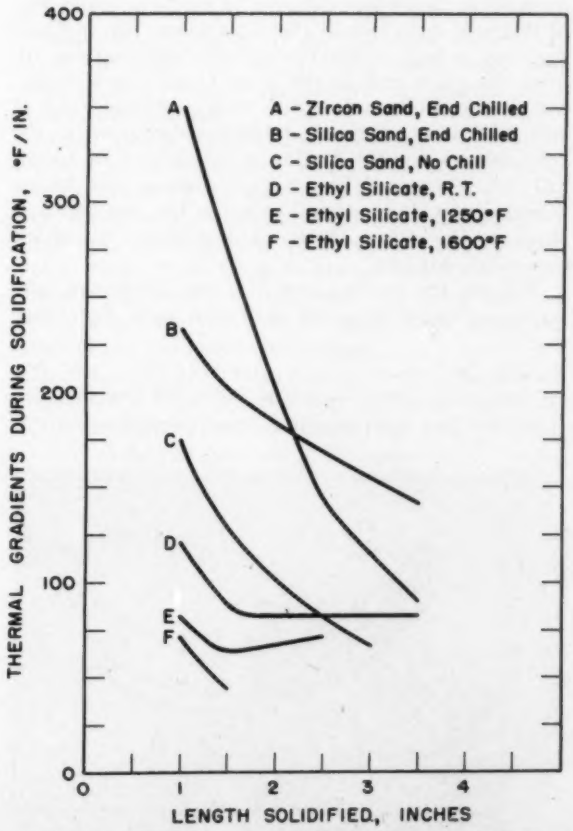



Fig. 2 — Average thermal gradients in the liquid-solid zone as a function of length of metal solidified for one in. thick plates (gradients measured between 2680 and 2610 F).

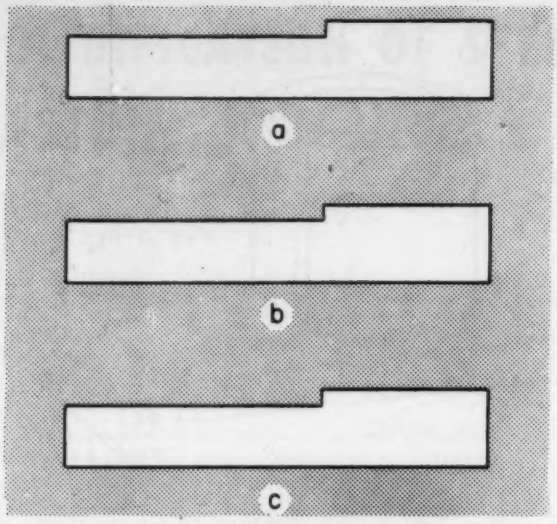



Fig. 3 — Semi-microradiographs of sections cut from cast one in. plates. (a) zircon sand, end chill, (b) silica sand, end chill, (c) silica sand, no chill.

### *Cooling Rate and Structure*

The different molding materials described had effects on solidification of the test plates in addition to altering amount and distribution of microporosity. As would be expected, the cooling rate (during solidification) was quite different in different plates and in different locations of the same plate, Fig. 5. Cooling rates as high as 300 F/min were obtained one in. from the plate end in the most rapidly cooled casting (zircon sand, end-chilled plate); cooling rates of only 20 F were obtained at the same location in the ethyl-silicate bonded mullite mold, heated to 1600 F (871 C). The curves for average cooling rate during solidification show generally similar but not identical shape to the curves for thermal gradients along the centerline, Fig. 2.

Altering the cooling rate of a cast metal can alter its metallurgical structure in several ways. For example, increasing cooling rate can (1) alter the grain structure, (2) increase the fineness of the dendritic structure (decrease dendrite arm spacing) and (3) decrease the size of inclusions. In this work the macrostructures of the plate castings were not appreciably affected by the varying thermal conditions. All castings were composed largely of coarse columnar grains, the only major differences between the castings being that those that were end chilled exhibited a marked fine columnar zone, which extended a maximum of about one in. from the chill face of the casting towards the riser.

Microstructures were, however, greatly affected; fast cooling rates produced an extremely fine structure within the austenite dendrites; slower cooling produced a substantially coarser structure, Fig. 6. Inclusion size and distribution in the cast plates were also dependent on mold material and on location within the plate. Fast cooling rates resulted in fine inclusions, evenly distributed; slower cooling rates produced coarser, more agglomerated inclusions. Each of the above effects (in addition to microporosity) would be expected to have some influence on mechanical properties.

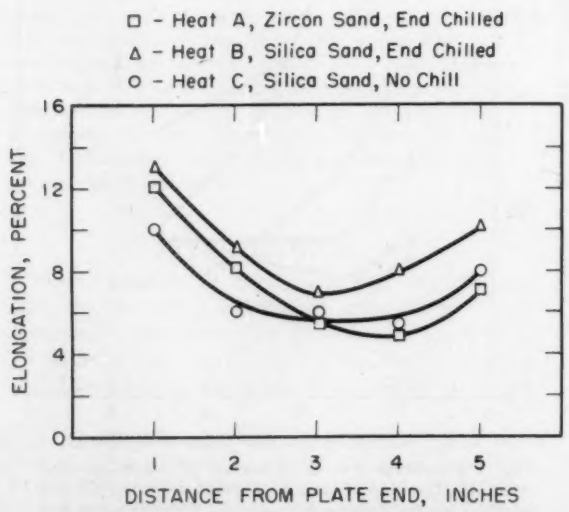



Fig. 4 — Elongation in one in. thick sand cast plate.

## ADDITIONAL PLATE CASTING STUDIES

In order to examine further the occurrence of microporosity in plate castings, a series of plates of various lengths was poured; the plates were ½-in. thick and 5 in. wide and were end risered (Fig. 7). Plates were cast of 5, 3 and one in. lengths. One series of plates was chilled at the end opposite the riser, where-

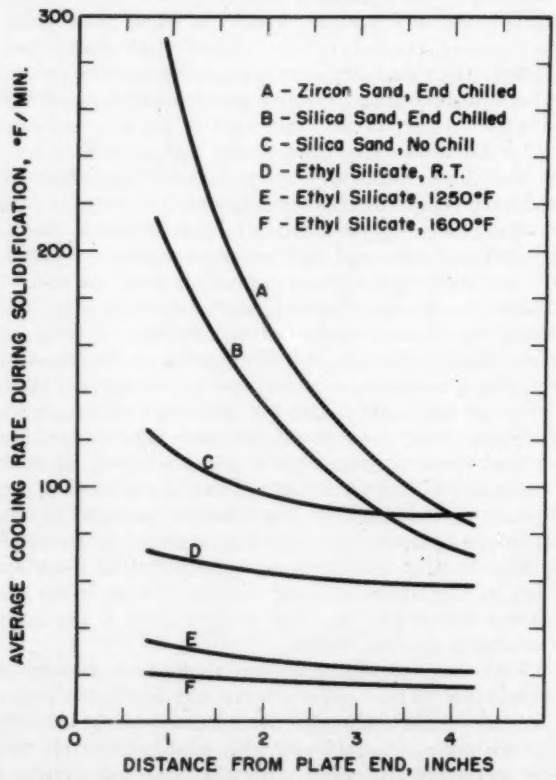



Fig. 5 — Average cooling rate during solidification from 2680 to 2610 F as a function of location in the cast plates.

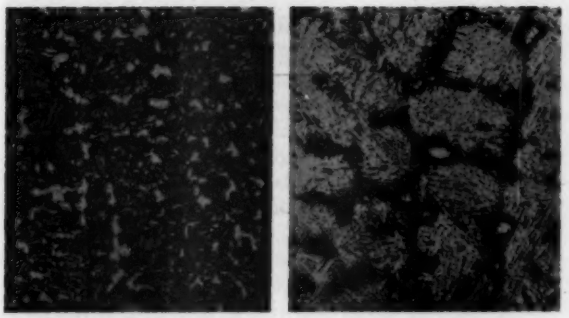


Fig. 6 — Microstructures from as-cast one in. thick plates. *Left* — silica sand mold, end chilled. *Right* — ethyl-silicate mold, no chill. 50 ×.

as a second comparison series was cast unchilled. The alloy studied was A.I.S.I. 4340.

All plates were apparently sound when examined by ordinary nondestructive radiography. This was expected from data on feeding distances in steel castings (except possibly for the unchilled 5 in. plate).[15] Nonetheless, microporosity was found to be present in all plates when examined by thin section radiography.

In this and subsequent portions of the investigation, microporosity was determined by radiographing sections approximately 0.006 in. thick (i.e., by microradiography). Several methods have been suggested for preparation of thin section for microradiography; the following procedure was employed. Specimens were mounted to a steel block with an epoxy resin mixture. They were ground to a thickness of about 0.010 in., removed and remounted on a special sample holder similar to that suggested by Sharpe.[16] After polishing one surface to a metallographic finish, the specimens were removed and the process repeated on the opposite surface; final sample thickness was approximately 0.006 in. X-raying was done using a standard x-ray diffraction unit.

Radiographs obtained were enlarged approximately 12× and the amount of microporosity then evaluated by placing a grid on the enlarged radiograph and counting the number of squares in which micropores were found. The amount of microporosity was determined as:

$$\text{Amount} = \frac{\text{number of squares containing micropores} \times 100}{\text{total number of squares}}$$

This method of rating microporosity results in a per cent microporosity that is magnified with respect to true volume per cent. The magnification comes from the fact that the method, in effect, measures the relative area of microporosity, but does so on a specimen of finite thickness. Actual magnification probably depends somewhat on pore size and shape as well as on experimental technique. The procedure has the advantages of sensitivity and a yielding semiquantitative data as to amount of porosity. Figure 8 illustrates a typical microradiograph of a fairly large amount of porosity (original magnification 50×); microradiographs showing three different levels of microporosity are in Fig. 9 (original magnification 12×).

As a result of examination of the amount of microporosity present along the lengths of the plates and

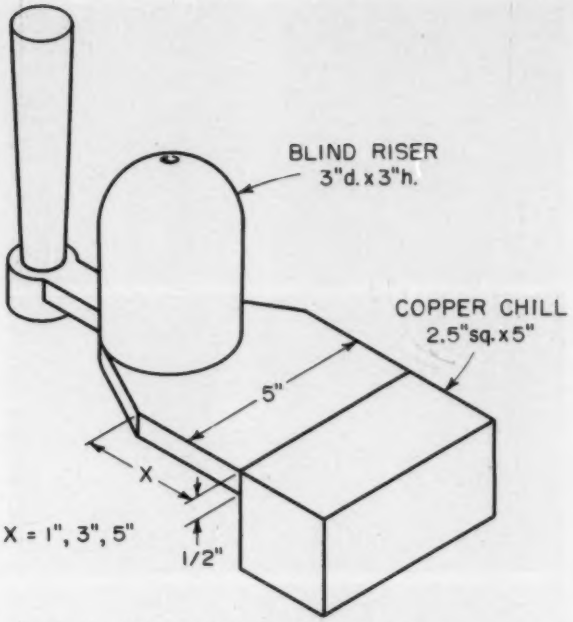



Fig. 7 — Design of ½-in. thick plate casting.

correlation of this porosity with location in the plate and with mechanical properties, Figs. 10 and 11 were prepared; more complete results are presented in Table 1. Major conclusions from these data are (1) a correlation exists between microporosity and ductility in the plate castings and (2) tensile and yield strengths are relatively unaffected by microporosity (in amounts studied).

In all plates, chilled and unchilled, only small amounts of microporosity were present in the vicinity of the end of the plate opposite the riser. However, this porosity increased with increasing distance along the plate. Figure 10 presents data for two typical plates (3 in. long). In the chilled plate, microporosity was essentially zero near the plate end but increased to one per cent at about 2 in. from the end, and to 3 per cent near the riser.

In the unchilled plate, microporosity was also essentially zero near the end, but increased rapidly at locations closer to the riser. Elongation and reduction

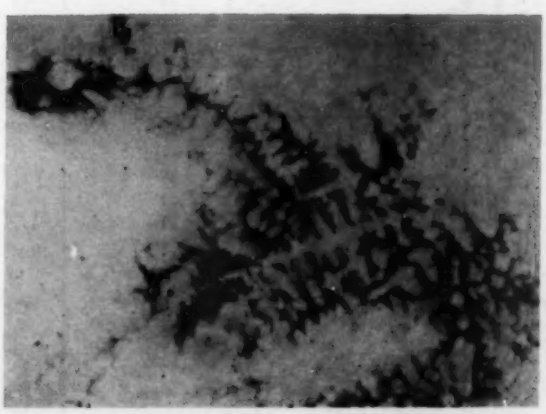


Fig. 8 — Microradiograph showing interdendritic nature of microporosity. 50 ×.

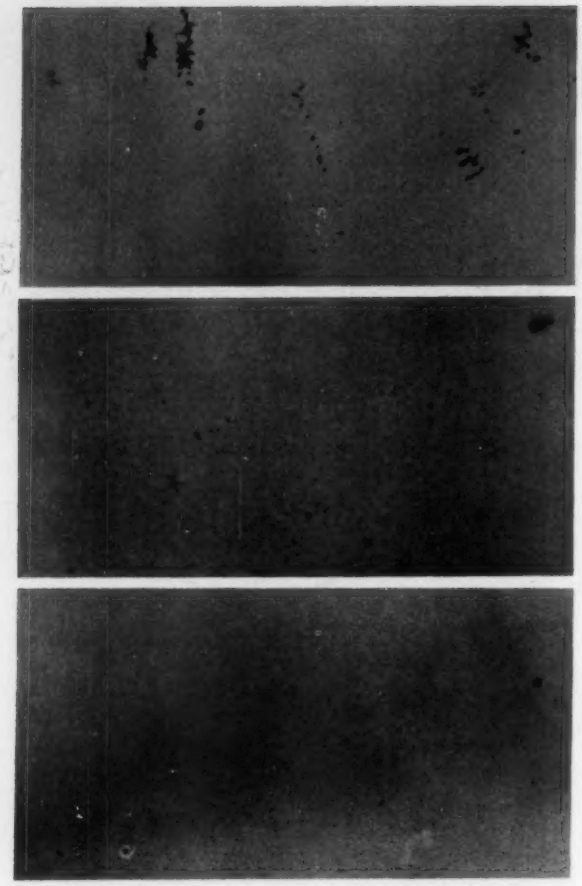
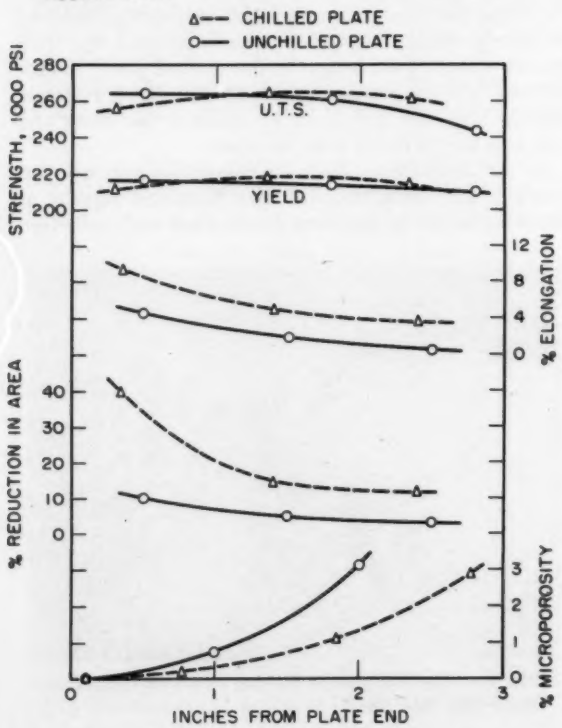


Fig. 9 — Typical microradiographs showing microporosity (black areas). *Top* — 3.50 per cent, *center* — 0.85 per cent and *bottom* — 0.20 per cent microporosity Approx. 12 ×.




in area generally decreased with increasing microporosity while tensile and yield strengths were relatively unaffected. Values of reduction in area as high as 40 per cent were obtained near the end chill.

Figure 11 summarizes microporosity data obtained from the 3 and 5 in. plates (chilled and unchilled); each of the points on this figure is the average of data from two to four different microradiographs. It should be remembered that per cent microporosity in Figs. 10 and 11 is a magnified quantity with respect to true volume per cent, due to the experimental method employed (point counting of microradiographs).

Figures 10 and 11 illustrate that it is extremely difficult to feed a steel casting to complete soundness. As thermal gradients are decreased from high values (several hundred degrees Fahr./in.) the amount of fine microporosity increases gradually. The type of microporosity that is detectable by microradiography is absent only in regions of a casting (1) close to a casting edge or (2) in the vicinity of a chill. Thermal gradients near a riser reduce porosity but do not eliminate it, as shown in the case of the 5 in. plate, Fig. 11. This fine microporosity, while probably of little significance in steel castings at low strength levels, is of major importance in determining ductility at the high strengths studied herein. There is some evidence that the porosity may influence properties of wrought steels as well.

### *Feeding to Soundness*

It is of interest to compare results obtained from various investigations on the length of uniform sec-

Fig. 10 — *Bottom — left* — Mechanical properties and microporosity in 3 in. plate castings.

Fig. 11 — *Below* — Summary of microporosity date for ½-in. plate castings. Each point represents an average of 2 to 4 different castings.

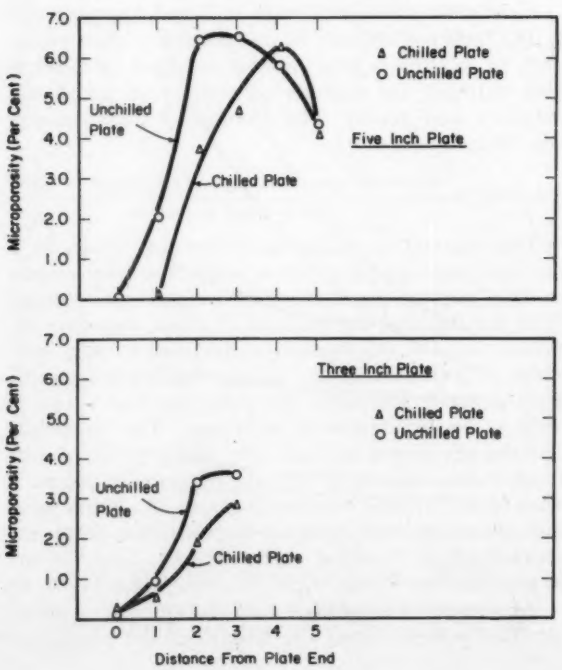


TABLE 1 — PROPERTIES OF ½-IN. THICK PLATE CASTINGS

| Description | Plate Length, in. | Location (in. from plate end) | Tensile Str., psi | Yield Str. 0.2 per cent offset, psi | Elong., % | Red. in area, % | Microporosity, % |
|---|---|---|---|---|---|---|---|
| unchilled | 1 | 0.5 | 262,000 | 216,000 | 5.7 | 15.6 | 0.90 |
| chilled | 1 | 0.5 | 264,000 | 216,000 | 7.9 | 27.0 | 0.10 |
| unchilled | 3 | 0.5 | 266,500 | 218,000 | 5.0 | 10.9 | 0.00 |
| unchilled | 3 | 1.5 | 262,000 | 216,000 | 2.1 | 5.5 | 0.71 |
| unchilled | 3 | 2.5 | 246,000 | 211,000 | 0.7 | 3.8 | 3.22 |
| chilled | 3 | 0.5 | 257,500 | 211,000 | 9.3 | 38.9 | 0.14 |
| chilled | 3 | 1.5 | 264,500 | 216,000 | 5.0 | 14.0 | 1.05 |
| chilled | 3 | 2.5 | 262,000 | 215,000 | 3.6 | 11.9 | 2.89 |
| unchilled | 5 | 0.5 | 265,000 | 216,000 | 2.1 | 3.8 | 1.05 |
| unchilled | 5 | 1.5 | —— | —— | — | — | 6.35 |
| unchilled | 5 | 2.5 | 242,000 | 213,000 | 0.7 | 3.3 | 3.54 |
| unchilled | 5 | 3.5 | 261,000 | 214,000 | 3.6 | 6.0 | 6.94 |
| unchilled | 5 | 4.5 | 258,000 | 216,000 | 1.4 | 4.4 | 4.33 |

tion castings that can be fed to soundness.[9,15,17] Table 2 lists data for chilled and unchilled plates, ½-in. thick. For castings of commercial quality (adequately sound to pass A.S.T.M. E-71, Class 2 standards), a plate 6 in. long can be produced with no chill and with one end riser; if a chill is employed at the extremity opposite the riser, an 8 in. long plate can be cast.

If apparent soundness is required when the plate is examined by nondestructive radiography, the maximum length that can be cast is somewhat shorter. If semi-microradiography is the inspection criterion the maximum plate length is again shorter, and if microradiography is the standard the maximum plate length is still shorter. The data in Table 2 have been assembled from widely different sources and apply to steels of slightly different compositions; however, the results almost certainly apply nearly quantitatively to 4330 or 4340 cast steel and illustrate how difficult it is to feed steel to complete soundness.

TABLE 2 — FEEDING DISTANCE IN END RISERED ½-IN. THICK PLATE CASTINGS

| Standard for Soundness | Feeding Distance, in. Unchilled Plate | Plate Chilled on end | Reference |
|---|---|---|---|
| ASTM E-71, Class 2 | 6 | 8 | 17 |
| Radiographically sound (nondestructive, 1½ percent definition) | 4 | 6 | 15 |
| Trace of porosity by semi-microradiography | 2 | 3 | 9 |
| Trace of porosity (approx. 1 per cent) by microradiography | 1 | 1.5 | This work |

## CYLINDER CASTINGS

In order to study solidification in shapes that would more closely simulate heavy castings or ingots, a series of cylinders were cast, 4 in. in diameter at the base, with a top riser 5 in. in diameter, Fig. 12. The casting was 9 in. high end poured weight was approximately 40 lb. Castings were made in different type molds, including (1) sand, (2) sand with bottom chill, (3) hot ethyl-silicate bonded mullite with a bottom chill and (4) moldable exothermic material with a bottom chill.

When a casting was made in sand (with a top insulating cover of rice hulls) gross shrinkage was present, but was generally confined to the riser section, Fig. 13. The structure of the zone free of macroshrinkage (bottom half of the casting) was usually composed of a short zone of columnar crystallization with the remainder of the casting equiaxed. In castings poured at low temperatures, the columnar zone was absent (Fig. 14); in castings poured at higher temperatures the columnar zone was up to ½ the radius in length.

When a large metallic chill was placed on the bottom face of the mold cavity, the shrinkage cavity was moved upwards in the casting, and a zone of fine columnar crystals grew from the chill face 1.5 to 2 in. upwards; otherwise the chill had little effect on the macrostructure (size of the equiaxed grains and length of the columnar grains on the side walls were not significantly altered).

To determine mechanical properties in the cylinder castings, coupons were cut from locations sketched

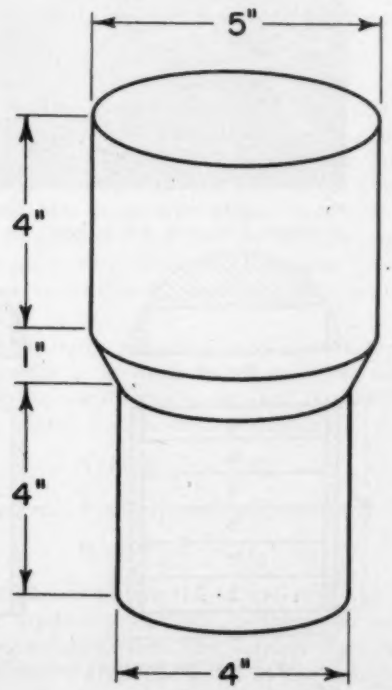



Fig. 12 — Nine in. high cylinder.

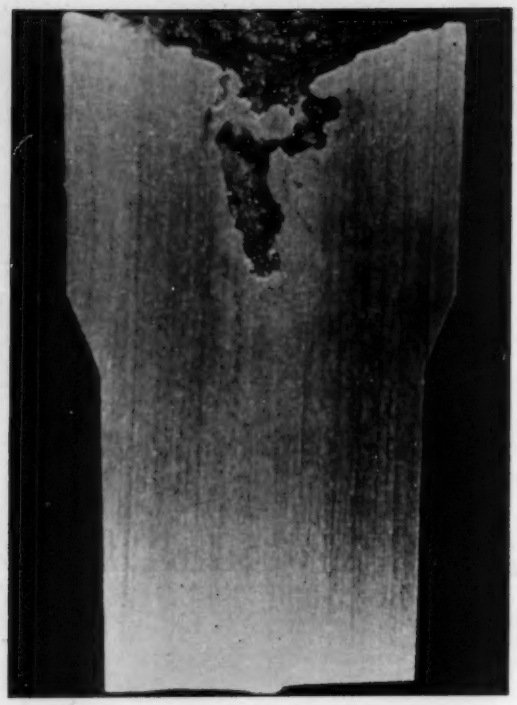


Fig. 13 — Section from 9 in. high cylinder casting.

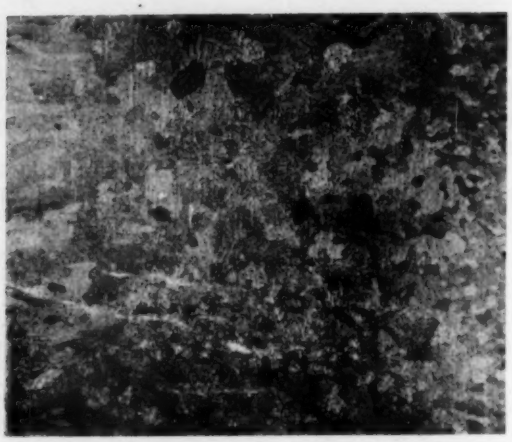


Fig. 14 — Macrostructure of sand cast 9 in. cylinder of section 4 in. wide x 4 in. high.

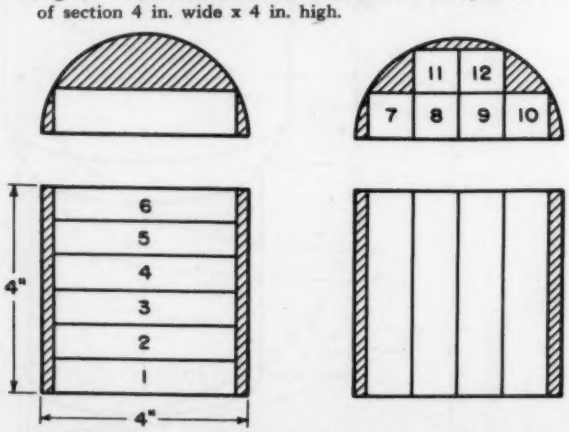



Fig. 15 — Test bar locations in 9 in. high cylinder casting.

in Fig. 15, heat-treated and machined into standard tensile specimens. Specimens for microradiography were also taken in the locations shown (along the centerline of the ingot). Ductility in sand cast cylinders was generally lower than that in the plates (Table 3), although in the bottom chilled cylinder ductility near the chill was quite high (Table 4). One reason for the generally low properties in cylinders was the presence of substantial microporosity (Table 5), but other factors related to the slower cooling of the sand cast ingot may also have been influential.

### *Unidirectional Solidification*

In a first attempt to solidify the ingot under steeper temperature gradients, a hot ethyl-silicate bonded mold preheated to approximately 1850 F (1010 C) was substituted in place of the sand mold. A water cooled copper chill was used on the bottom face, and nonmetal producing exothermic material was placed on top of the riser. Extensive piping still occurred in the riser, and the casting was composed largely of equiaxed grains.

The mold design, illustrated in Fig. 16, was then constructed so as to achieve essentially unidirectional heat flow during solidification. Sleeves of preformed moldable exothermic material were set in a sand mold to form the side walls of the mold cavity, a water cooled chill formed the base and a layer of exothermic material placed on top of the casting (after pouring) formed the cope. A somewhat similar mold arrangement has been described.[18] The exothermic material employed burns at a temperature slightly in excess of the temperature at which steel solidifies. As a result, if any adequately thick layer of exother-

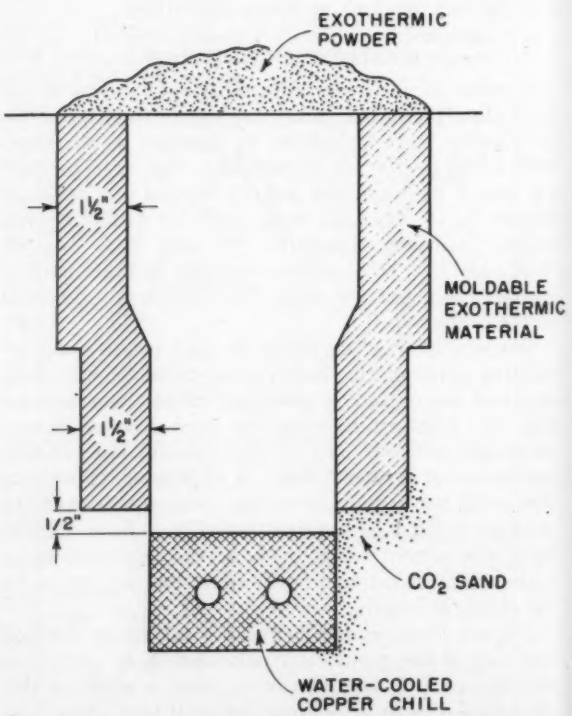



Fig. 16 — *Left — top* — Mold for unidirectionally solidified 9 in. high cylinder.

mic material is employed there can be no transfer of heat from the casting to the mold (except downward to the water cooled chill). There may, on the other hand, be some heat supplied from the mold walls to the casting.

In this casting, the sleeves were approximately 1½ in. thick. In early tests, they were pre-ignited and allowed to burn approximately 4 min before pouring the metal. In later tests the sleeves were not pre-ignited, and little difference was observed in macrostructures obtained. A total of five castings were poured which had structures similar to that of Fig. 17. They were columnar through most of their height, with the columnar grains extending vertically from the chill face upwards. In each of the castings the top inch or two contained equiaxed grains and some gross shrinkage. It appeared that the columnar grains might have been grown further had a greater amount of exothermic material been employed.

To determine if columnar grains could be grown to greater lengths, the ingot design sketched in Fig. 18 was constructed. The ingot was of uniform diameter (4 in.) and was 15 in. high. Moldable exothermic material was prefabricated into a series of sleeves of varying thicknesses; thicker sleeves were used at the top of the casting than at the bottom. The casting was bottom gated, and otherwise experimental procedure was similar to that for the 9 in. ingot. The macrostructure of the casting (Fig. 19) exhibited columnar grains 11 in. long out of a total possible 13 in. (total linear shrinkage in the casting was approximately 2 in. so the overall length of the final casting was only 13 in.).

### *Solidification Cooling Curves*

Cooling curves were taken during solidification of two castings (a 9 in. cylinder and the 15 in. cylinder). Thermocouples were placed along the centerline of the castings at varying distances from the water cooled base plate. Three thermocouples were employed in the 9 in. cylinder and four in the 15 in. cylinder. In analyzing the thermal data obtained, the liquidus and solidus temperatures for the alloy were assumed to be 2700 F (1483 C) and 2600 F (1427 C), respectively. While these temperatures are not precise, relationships developed through their use would be affected only slightly by changes in the assumed temperatures.

In each of the castings studied, the advance of the fully solid interface was proportional to the square root of time. In the smaller casting, the position of the fully solid interface at any time corresponded to the equation:

$$D = 0.95\sqrt{t} - 0.4 \qquad (1)$$

where:

D = distance from chill to fully solid interface (in.),
t = time (min).

In the larger casting the relationship was:

$$D = 1.15\sqrt{t} - 0.1 \qquad (2)$$

TABLE 3 — TENSILE PROPERTIES OF SAND CAST 9 IN. HIGH CYLINDER CASTINGS

| Casting No. | Location | Tensile Str., psi | Yield Str. 0.2 per cent offset, psi | Elong., % | Red. in area, % |
|---|---|---|---|---|---|
| A | 1 | 258,000 | 211,000 | 4.3 | 10.9 |
| | 2 | 258,000 | 215,000 | 4.3 | 9.3 |
| | 3 | 254,000 | 215,000 | 2.1 | 4.4 |
| | 4 | 254,000 | 217,000 | 3.0 | 4.2 |
| | 5 | 264,000 | 219,000 | 5.0 | 5.6 |
| | 6 | 252,000 | 216,000 | 2.0 | 2.6 |
| B | 2 | 264,000 | 207,000 | 5.0 | 6.6 |
| | 4 | 263,000 | 206,000 | 2.9 | 3.3 |
| | 7 | 268,000 | 211,000 | 2.9 | 4.9 |
| C-1 | 3 | 272,000 | 218,000 | 5.0 | 7.1 |
| | 5 | 273,000 | 223,000 | 4.3 | 6.0 |

TABLE 4 — TENSILE PROPERTIES OF BOTTOM CHILLED SAND CAST 9 IN. HIGH CYLINDER CASTING

| Casting No. | Location | Tensile Str., psi | Yield Str. 0.2 per cent offset, psi | Elong., % | Red. in area, % |
|---|---|---|---|---|---|
| C-2 | 1 | 270,000 | 218,000 | 10.7 | 27.0 |
| | 3 | 271,000 | 219,500 | 6.4 | 13.0 |
| | 5 | 272,000 | 224,000 | 3.6 | 5.5 |
| | 7 | 281,000 | 232,700 | 4.0 | 5.5 |

TABLE 5 — MICROPOROSITY AND REDUCTION IN AREA OF SAND CAST 9 IN. HIGH CYLINDER CASTINGS

| Casting | Location | Porosity, % | Red. in area, % |
|---|---|---|---|
| unchilled | 2 | 1.3 | 7.1 |
| unchilled | 6 | 1.0 | 6.0 |
| bottom chilled | 1 | 0.4 | 27.0 |
| bottom chilled | 3 | 0.5 | 13.0 |
| bottom chilled | 5 | 0.1 | 5.5 |

The linear relationship with the square root of time, in agreement with theory,[19] indicates that heat flow in solidification of the castings was essentially unidirectional. Had significant heat in these castings been extracted through the sidewalls, a plot of distance solidified versus the square root of time would not have been linear but would have been of continually increasing slope. Experimental work using other metal systems has indicated this to be the case.[5]

Solidification rates measured in this work are not greatly different from those found by other investigators studying solidification against flat walls of large steel ingots. For example, Marburg[20] found:

$$D = 1.20\sqrt{t} - 0.90 \qquad (3)$$

and Chipman and FonDersmith's equation[21] is:

$$D = 0.9\sqrt{t} - 0.12 \qquad (4)$$

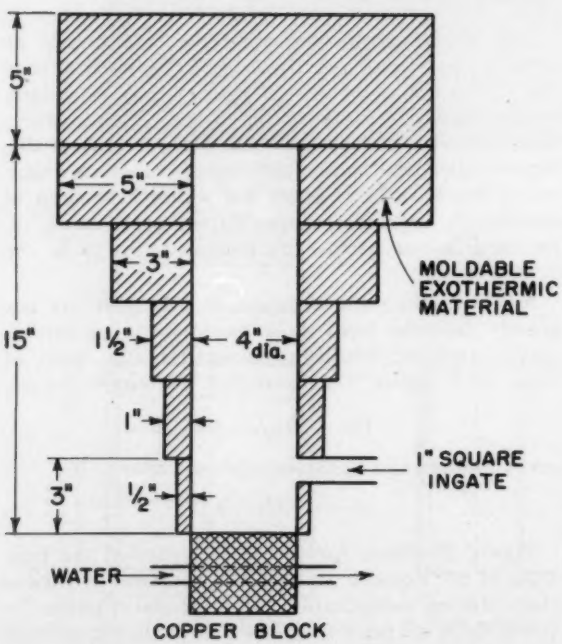
Figure 20 shows, for the 15 in. cylinder, the positions of the liquidus and solidus isotherms at various times during solidification. The vertical distance between these curves gives the width of the liquid-solid



Fig. 17 — *Left — top* — Macrostructure of unidirectional 9 in. ingot.

Fig. 18 — *Left — bottom* — Mold for unidirectionally solidified 15 in. high cylinder.

Fig. 19 — *Below* — Macrostructure of a unidirectionally solidified 15 in. high cylinder.






region at any time during freezing (or when a given amount $X$ is frozen). This distance, sometimes termed the width of the mushy zone, is unusually small; for comparison, in ideal unidirectional solidification of many nonferrous alloys, the mushy zone is at least as long as the length of fully solidified metal.

To compare thermal data for the unidirectionally solidified cylinders with those given earlier for the plate castings, it is convenient to consider (1) cooling rate during solidification and (2) thermal gradients during solidification. These parameters were determined here in the same manner as they were for the plates (they may be obtained directly from Fig. 20).

A characteristic of unidirectional solidification is that while high cooling rates are obtained near the chill, these rates drop quickly (the square root of the reciprocal of cooling rate is approximately proportional to distance from the chill, Fig. 21). Cooling rates near the chill in the cylinder castings compare with those near the chill in the plate castings (Fig. 5). However, at distances of more than several inches from the chill, cooling rates in the cylinders are substantially lower than those in the plates.

On the other hand, thermal gradients during solidification are steeper in the cylinder castings than in the plate castings. Figure 22 shows the longitudinal gradients in the 15 in. cylinder casting. These are inversely proportional to the amount of metal solidified, and are significantly greater than the gradients in any of the plates, Fig. 2. It is these steep temperature gradients that result in columnar grain formation, and a narrow mushy zone during solidification.

### *Columnar Grain Structure*

The width of the columnar grains in the unidirectionally solidified ingots increases with increasing distance from the chill. This has been observed for other metal systems, and is because grains more favorably oriented for growth gradually crowd out their neighbors during solidification. In dendritic growth of cubic metals, the preferred direction of columnar growth is the heat flow direction <100>. If, in two adjacent columnar grains, the angle between their respective <100> directions is divergent, the more favorably oriented grain will grow at the expense of the one less favorably oriented.[22] Table 6 lists average grain width at different distances from the chill for several ingots (as measured on a vertical section).

Columnar grains of the type grown herein possess an internal dendritic structure and so are often referred to as "columnar dendrites." The internal structure is apparent to some degree in the macrostructure of Fig. 17, and more clearly in the microstructure of Fig. 23. Ideally, a dendrite consists of one or several similarly oriented columnar stalks (or arms) with secondary arms emanating therefrom; tertiary arms may then emanate from these, etc.

Generally, slow solidification rates result in a coarse dendritic structure (relatively large distances between branches, and little or no formation of tertiary branches); higher solidification rates result in finer structures. It is now clear that the reason cooling rate and other variables affect the fineness of dendrite structures in an alloy such as the one considered here, is because the variables affect the amount of solute that must be transported from the liquid-solid interface in a given time; i.e., the dendrites are solute dendrites.

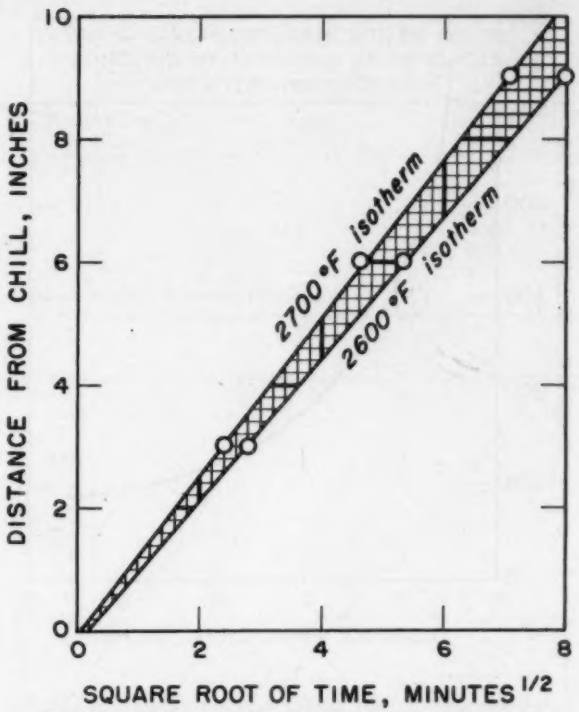



Fig. 20 — Distance to liquidus and solidus vs. sq rt of time for 15 in. cylinder.

One measure of the fineness of dendritic structure is the distance between branches of the dendrite. It

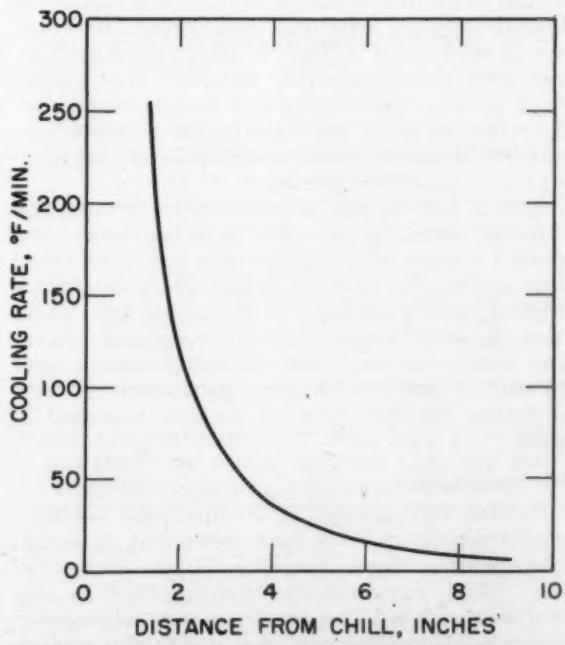



Fig. 21 — Average cooling rate during solidification as a function of locations in the cylinder for 15 in. cylinder (cooling rates measured between 2700 and 2600 F).

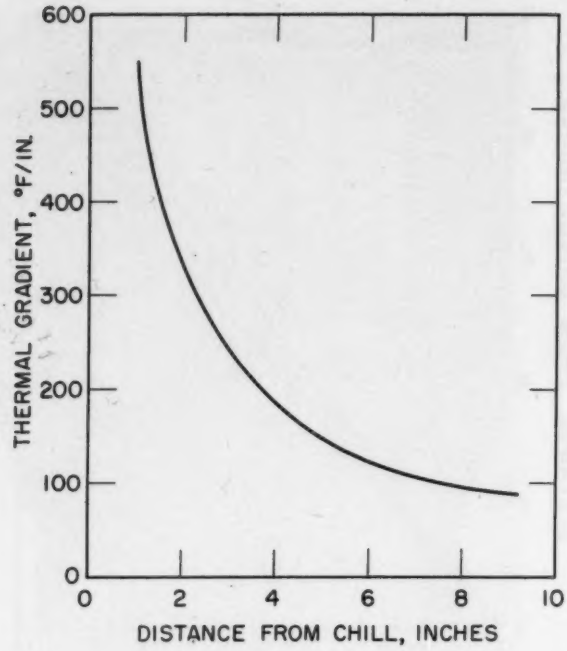
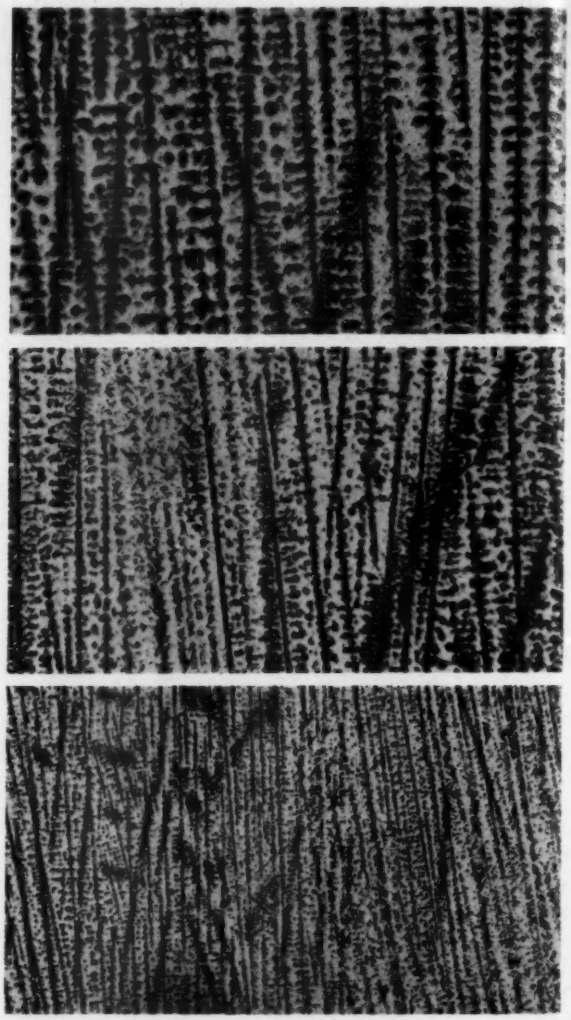



Fig. 22 — *Left* — Average thermal gradients in the liquid-solid zone as a function of length of metal solidified for 15 in. high cylinder (gradients measured between 2700 and 2600 F).

Fig. 23 — *Below* — *top* — 3 in., *center* — 1¾-in. and *bottom* — ½-in. from chill. Dendrite arm spacing in a 9 in. unidirectional casting.



has been shown that under a wide variety of conditions this dendrite arm spacing is proportional to the square root of solidification time (or inversely proportional to the square root of average solidification rate).[23,24] Theoretical analysis has indicated this can be explained from fundamental mass transport considerations.[24]

In this work, spacings between both primary arms and secondary arms were measured on several castings. The spacing between primary arms is the average distance between the arms which appear as vertical lines in Fig. 23; secondary arm spacing is distance between horizontal lines. Results from one casting are plotted in Fig. 24; results for other castings were nearly identical. Dendrite arm spacing (both primary and secondary) is directly proportional to distance from the chill, and is therefore proportional to the reciprocal of the square root of freezing rate as would be expected.

It is of interest that in addition to primary and secondary arms, the dendrites examined herein possessed a coarser, superimposed structure. This structure can be seen in Figs. 17 and 23; it consists of dendrite arms at an angle to the vertical axis. These arms, like those described above are regularly spaced; their spacing increases with increasing distance from the chill. Several possibilities might account for their formation, but these have not yet been examined in detail.

### *Macrosegregation*

Previous investigations of unidirectional solidification of nonferrous alloys have shown that no significant positive macrosegregation occurs (except in alloys which expand during freezing).[25,28] In fact, analysis has shown that significant macrosegregation cannot result in most such cases if diffusion alone is the driving force.[29] The type of segregation which has been observed in unidirectional solidification of nonferrous ingots is inverse (increase in low melting constitutents in the portion of the ingot to freeze first), resulting primarily from flow of solute rich liquid between growing dendrites to feed solidification shrinkage.

In the experimental work conducted essentially no positive macrosegregation was observed, except in the elements silicon and sulfur (Tables 7 and 8). Most of the remaining elements showed a slight tendency to segregate inversely; however, in most instances the amount of this inverse segregation barely exceeded analytical error. One exception was in the case of the analysis in the upper portion of the 15 in. ingot. Here carbon content was 0.34 per cent as compared with a nominal of 0.40 per cent, due to flow downwards of solute rich liquid. This portion of the ingot was quite porous.

In the case of silicon and sulfur, analyses invari-

ably showed higher contents at the top than the bottom. In one ingot 0.065 per cent sulfur was found near the top, while a similar analysis at the bottom showed only 0.008 per cent sulfur. It would appear that flotation of inclusions is the major factor responsible for the segregation. Tables 7 and 8 show a progressive but gradual increase in alloy element until near the top of the ingot; then, at the top, a discontinuous increase occurs. Also, the microstructures near the top of the ingot were generally considerably dirtier than those from the remainder of the ingot.

Contamination of the metal by the exothermic material was probably also a contributing factor to the segregation observed, at least in the case of silicon. Table 8 compares melt analyses of 2 heats with analyses from the castings poured in these heats. In general, slightly more silicon is present in the castings than was in the original metal; reaction of the metal with the molding material during pouring and/or solidification would readily account for this.

### *Inclusion Flotation*

The ease of flotation of inclusions during freezing of a unidirectionally solidified ingot can be seen by considering Stoke's law:

$$u = \frac{gD^2 (\rho_i - \rho_e)}{18\mu} \qquad (5)$$

where

$u$ = terminal settling velocity (cm/sec).
$g$ = acceleration due to gravity (cm/sec$^2$).
$D$ = diameter (cm).
$\rho_i$ = density of inclusion (gm/cm$^3$).
$\rho_e$ = density of liquid metal (gm/cm$^3$).
$\mu$ = viscosity (poise).

Assuming the viscosity of molten steel to be 0.025 poise, and density of inclusions to be about 4 gm/cm$^3$, it can be shown that inclusions as small as 20 microns in diameter will float upwards at a rate of 0.6 in./min. This rate exceeds the rate of progress of the solidification front at only 0.55 in. from the chill. At distances greater than this from the chill, rate of progress of the solidification front is much slower, and even smaller inclusions can float in advance of the solidification front.

It is clear that small inclusions can float if they form above the liquidus temperature of the molten steel; here, no solid metal is present to prevent their movement. Inclusions which form below the liquidus temperature grow in the mushy zone (Fig. 20), and float or not depending on the extent to which dendrite arms have reached across interdendritic pools. It is probable that in unidirectional freezing some inclusions do float out of the mushy zone due to its narrow width.

### *Microporosity and Mechanical Properties*

Each of the unidirectionally solidified castings were examined microscopically and microradiographically to determine amount and distribution of microporosity. In the case of the cylinders cast in sand molds, amount of porosity varied from zero per cent up to 2 per cent, depending on location in the casting and on chilling techniques.

**TABLE 6 — AVERAGE GRAIN WIDTH (IN IN.) AS A FUNCTION OF DISTANCE FROM THE CHILL FOR CYLINDER CASTINGS**

| Distance from chill, in. | 9 in. high cylinder* | 15 in. high cylinder |
|---|---|---|
| 1 | 0.15 | 0.16 |
| 2 | 0.18 | 0.23 |
| 3 | 0.23 | 0.30 |
| 4 | 0.28 | 0.37 |
| 6 | — | 0.45 |
| 8 | — | 0.57 |

*Data averaged from two cylinders.

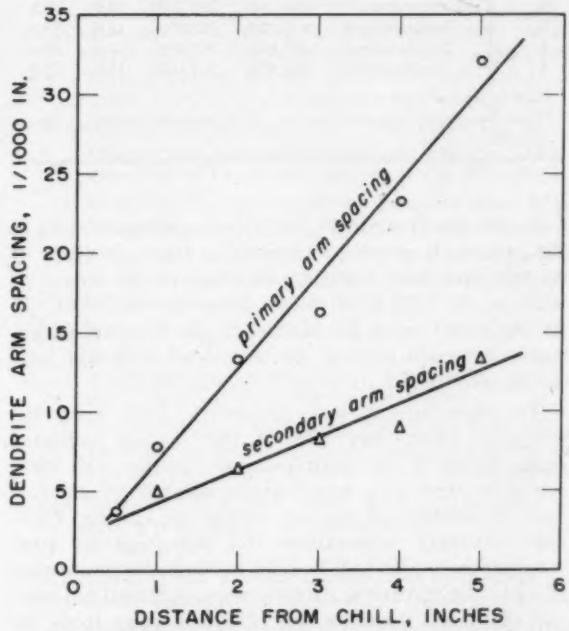



Fig. 24 —Dendrite arm spacing as a function of location in a unidirectionally solidified cylinder from cylinder shown in Fig. 17.

**TABLE 7 — SEGREGATION DATA FROM A UNIDIRECTIONALLY SOLIDIFIED CASTING (15 IN. HIGH CYLINDER)**

| Location (Distance from chill, in.) | C | Mn | Si | Ni | Cr | Mo | P | S |
|---|---|---|---|---|---|---|---|---|
| ¼ | 0.40 | 0.79 | 0.25 | 1.85 | 0.89 | 0.25 | 0.002 | 0.013 |
| 4 | 0.40 | 0.78 | 0.38 | 1.91 | 0.90 | 0.24 | 0.002 | 0.011 |
| 8 | 0.43 | 0.80 | 0.25 | 1.91 | 0.90 | 0.24 | 0.002 | 0.010 |
| 12-13 | 0.34 | 0.70 | 0.76 | 1.83 | 0.85 | 0.23 | 0.009 | 0.028 |

**TABLE 8 — COMPARISON OF MELT ANALYSES WITH SAMPLES CUT FROM BOTTOM AND TOP OF UNIDIRECTIONALLY SOLIDIFIED CASTINGS**

| Casting | Location | Composition Element, % | | | | | | | |
|---|---|---|---|---|---|---|---|---|---|
| | | C | Mn | Si | Ni | Cr | Mo | P | S |
| 5 | melt | 0.46 | 0.80 | 0.31 | 1.83 | 0.83 | 0.30 | 0.012 | 0.012 |
| | near chill | 0.45 | 0.77 | 0.32 | 1.80 | 0.85 | 0.26 | 0.007 | 0.010 |
| | near top | 0.42 | 0.71 | 0.45 | 1.83 | 0.84 | 0.26 | 0.008 | 0.027 |
| 6 | melt | 0.45 | 0.92 | 0.36 | 1.77 | 0.91 | 0.23 | 0.008 | 0.031 |
| | near chill | 0.44 | 0.86 | 0.35 | 1.79 | 0.89 | 0.26 | 0.007 | 0.024 |
| | near top | 0.43 | 0.81 | 0.51 | 1.79 | 0.88 | 0.26 | 0.006 | 0.039 |

TABLE 9 — TENSILE PROPERTIES OF UNIDIRECTIONALLY SOLIDIFIED 9 IN. HIGH CYLINDER CASTINGS

| Casting No. | Bar No.* | Orientation** | Tensile Str., psi | Yield Str., psi, 0.2 % offset | Elong., % | Red. in area, % |
|---|---|---|---|---|---|---|
| 2 | 1 | transverse | 292,000 | 231,000 | 8.6 | 23.6 |
| 2 | 3 | transverse | 290,000 | 230,000 | 6.4 | 20.9 |
| 2 | 5 | transverse | 292,000 | 235,500 | 5.7 | 8.7 |
| 2 | 9 | longitudinal | 292,000 | 235,000 | 6.4 | 11.9 |
| 2 | 10 | longitudinal | 294,000 | 244,000 | 7.9 | 20.6 |
| 2 | 12 | longitudinal | 292,000 | 231,000 | 7.1 | 19.6 |
| 4 | 1 | transverse | 262,000 | 208,500 | 7.9 | 20.1 |
| 4 | 3 | transverse | 260,000 | 208,000 | 6.4 | 8.7 |
| 4 | 5 | transverse | 262,500 | 210,000 | 5.7 | 7.6 |
| 4 | 9 | longitudinal | 259,000 | 208,500 | 10.0 | 33.1 |
| 4 | 10 | longitudinal | 261,000 | 210,000 | 10.0 | 30.8 |
| 4 | 12 | longitudinal | 261,400 | 209,500 | 10.0 | 32.2 |

*Bar locations shown in Fig. 15.
**Transverse or longitudinal to direction of columnar grain growth.

In the unidirectionally solidified castings, results of the microradiographic examination were (1) little or no microporosity could be detected in the ingots at least up to 4 to 6 in. away from the chill but (2) in the top 1 to 2 in. of each ingot amount of porosity increased rapidly (in this location it was present in amounts up to 10 per cent, Fig. 19).

To date, mechanical properties have been determined from only two of the castings reported here, Table 9. In both castings, tensile and yield strengths were not significantly affected by orientation or position of the bar within the casting. However, ductility (elongation and reduction in area) was markedly affected. In both castings, optimum ductility in the transverse direction was obtained adjacent the chill; less ductility was obtained away from the chill. In one of the two castings transverse properties were not greatly different from properties taken longitudinally to the chill, but in the other casting ductility was markedly superior in the longitudinal direction.

Ductility (at a given strength level) was substantially higher in the unidirectionally solidified ingots then in the sand cast ingots. For example, reductions in area of the sand cast cylinders varied from 3 to 11 per cent, whereas in the unidirectionally solidified castings the range was 8 to 24 per cent. In the case of the sand cast cylinder that was bottom chilled, properties were higher, but except immediately adjacent the chill, were not as high as properties from the unidirectionally solidified plates. The high properties obtained in the unidirectionally solidified ingots are attributed to (1) reduction of microporosity, (2) improved chemical homogeneity on a micro- and macro-scale and (3) increased freedom from inclusions.

## SUMMARY AND CONCLUSIONS

A summary has been presented of work conducted at Massachusetts Institute of Technology over the last three years relating to solidification of low alloy steel castings and ingots. The main emphasis of the work has been on examining (1) different methods of obtaining directional solidification in cast steel and (2) effects of solidification variables on structure, segregation and properties of cast steel. Flat plates and cylindrical castings weighing approximately 50 lb were cast using techniques to achieve varying degrees of directional solidification. Thermal gradients and cooling rates were measured during solidification and related to structure and properties.

Microporosity can be discerned in flat plates by careful microradiography even when solidification takes place under relatively steep thermal gradients (in excess of 150-200 F/in.). Ductility at high strength levels is markedly affected by presence of the microporosity.

Feeding distance in steel castings is a quantity that is strongly dependent on degree of soundness required. For example, in an unchilled ½-in. thick plate casting, feeding distance is 6 in. if a casting of ordinary commercial standards is to be produced; it is 4 in. if the casting must be radiographically sound, 2 in. if it must be essentially sound when examined by semi-microradiography and one in. if it must be essentially sound when examined by microradiography.

Microporosity is reduced in plates and cylinders by increasing thermal gradients during solidification. Techniques employed in this work were (1) use of chills and (2) use of chills plus an insulating molding material. Generally, techniques which result in steep thermal gradients during solidification also result in a fast cooling rate during solidification; as a result, they refine inclusions and decrease dendrite arm spacing (in addition to reducing microporosity). Techniques which promote solidification under steep thermal gradients increase ductility in cast steel as a result of the improved soundness and homogeneity.

### *Thermal Gradients*

Steepest thermal gradients measured in this study were obtained in cylinder castings using an arrangement designed to permit essentially unidirectional heat flow during solidification. A mold was employed with a water cooled base plate; sides and top of the mold were insulated with moldable exothermic material. By this technique, ingots were cast with columnar grains extending from the base plate nearly to the top of a 15 in. high ingot.

Microporosity was essentially absent in unidirectionally solidified cylinders, except at the topmost portions. For comparison, it was present in amounts up to 3 per cent in plates (porosity measured by point counting microradiographs), and in amounts the order of one per cent in cylinders solidified in sand molds. Dendrite arm spacing in the unidirectional ingots is inversely proportional to the square root of cooling rate; this spacing increases with increasing distance from the chill. Grain width also increases with increasing distance from the chill.

Unidirectionally solidified ingots are remarkably free of macrosegregation; the only significant segregation being an increase in sulfur and silicon at the top primarily due to flotation of inclusions during freezing.

Mechanical properties of unidirectionally solidified cylinders (particularly reduction in area and elongation) were markedly superior to those of cylinders cast in sand. For example, in one unidirectional ingot (in the heat treated condition) mechanical properties parallel to the columnar grains averaged approximately 290,000 psi ultimate tensile strength, 230,000 psi yield strength and 17 per cent reduction in area; properties were essentially the same transverse to the columnar grains. In another unidirectionally solidified ingot, mechanical properties parallel to the columnar grains averaged approximately 260,000 psi tensile strength, 209,000 psi yield strength and 31 per cent reduction in area; in the transverse direction, properties averaged 260,000 psi tensile strength, 209,000 psi yield strength and 9 per cent elongation. In a comparable ingot produced in a sand mold average properties obtained were, respectively, 275,000, 225,000 and 6 per cent.

The improvement in properties obtained by unidirectional solidification is attributed to (1) reduction and microporosity, (2) improved chemical homogeneity on a micro- and macro-scale and (3) increased freedom from inclusions. As a result of improved soundness and homogeneity, unidirectionally solidified ingots should prove to be more readily forgeable than conventional ingots. Also, wrought material produced from such ingots should be of unusually high quality (i.e., possess high transverse ductility, little or no chemical or other heterogeneities, etc.).

## ACKNOWLEDGMENT

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# HORIZON REQUIREMENTS FOR CASTINGS

by S. R. Carpenter


## ABSTRACT

**Current application of castings in Aerospace vehicles at the author's company are reviewed as a springboard to delineate additional casting requirements for design solutions ahead.**

**Power available for use in recent years is increasing at an exponential rate. During the period 1945-1965 power plant thrusts will increase from 3000 lb to 1,500,000 lb. The increase in power has created technological deficiencies in material requirements. Materials such as beryllium, titanium, stainless steel super alloys and the refractories are under intensive development to meet the needs. Parallel effort with castings is required to solve fabrication problems and conservation of high cost materials.**

**A study is presented showing cost benefits of castings in relation to forgings. Savings in material by use of castings can amount to as much as $63,000 for 100 lb of finished parts in a material like columbium.**

**The future outlook for high quality castings is reviewed with favorable findings due to technical developments in fields applicable to foundry practices.**


## INTRODUCTION

The aerospace industry is one of rapid change, dealing with complex systems. Capabilities are constantly being extended to the limit, and this dynamic action is needed in all industries supporting the product. Technology has advanced rapidly, and many newspaper reports of scientific achievement, routine as they seem, were imaginative only a decade earlier. As an example of how far and how fast progress has gone, the Aerospace Industries Association 1960 Forecast[1] cites technical advancement in these words:

> "The decade 1960-1970 will see man's first personal entry into space. That this achievement postdates by only sixty years the birth of American aviation industry is indicative of the rate at which technical capability is advancing. The rate of development of new scientific concepts now is such that plans for the accommodation of future needs must be determined at the earliest possible date."

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Many tools are required for these future needs, and they are near. Some of the tools become available much sooner than others. The full complement for effective use is paced by a single one. Required structural materials in desired forms including castings is a pacing item for aerospace vehicles of the future.

## AEROSPACE CASTING USES REVIEW

Currently, one of the author's company's jet liners makes use of castings in accordance with the breakdown on Table 1. The spectrum of application ranges from nonstructural to primary structural applications.

TABLE 1 — AIR FRAME CASTINGS FOR JET LINER

| Casting Alloy | Sand Cast | | Investment Cast | | Total | |
|---|---|---|---|---|---|---|
| | Designs | Castings | Designs | Castings | Designs | Castings |
| AZ91 Magnesium | 27 | 38 | | | 27 | 38 |
| ZK51 Magnesium | 104 | 174 | | | 104 | 174 |
| 356 Aluminum | 24 | 72 | 17 | 29 | 41 | 101 |
| 4130 Steel | | | 1 | 1 | 1 | 1 |
| 304 Stainless Steel | | | 3 | 6 | 3 | 6 |
| 347 Stainless Steel | | | 9 | 66 | 9 | 66 |
| 410 Stainless Steel | | | 11 | 35 | 11 | 35 |
| 17-4 Stainless Steel | | | 4 | 34 | 4 | 34 |
| AMS 5385 | | | 1 | 1 | 1 | 2 |
| Total | 155 | 284 | 46 | 173 | 201 | 457 |

There is flexibility in design of small complex shapes by the investment casting process. Figure 1 shows typical examples of investment castings with which there is a high degree of success in steel and aluminum alloys. Application studies of aircraft quality steel investment castings with a fair amount of complexity (hollow sections with 0.060 in. internal webs), and success is confined to about the 5 lb



Fig. 1 — Typical steel investment casting in a jet liner.

category. It is not intended to infer this size is maximum. Steel investment castings 30 in. long with a weight of 30 lb have been noted.

Experience with shell mold steel castings is limited to the type of Fig. 2. The round section is cored, and casting weight is 30 lb. Development was quite lengthy, and there was sporadic difficulties throughout the production run.

Nonferrous types of sand castings are represented in size and complexity by those of Figs. 3 and 4. There are notable examples of much larger nonferrous castings currently in use by other aircraft producers. Nevertheless, the castings for the company's applications have defects in various degrees requiring compensations in design by increasing section sizes.

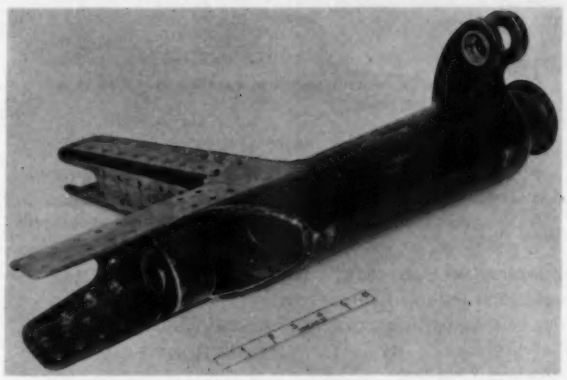


Fig. 2 — Shell mold type 410 steel casting in interceptor airplane.

## CASTING DEFICIENCES

An analysis of 1710 castings representing 19 designs showed a rejection rate of 183 parts, an overall average of 10.7 per cent. Per cent of rejections in various increments is found in Table 2.

Table 3 is an analysis by type of all the rejections reported in Table 2. Casting rejections, such as noted in Tables 2 and 3, are powerful influencing factors in selection of casting applications. Compensating fac-

TABLE 2 — INCREMENTAL REJECTIONS

| No. of Castings | Rejections, % |
|---|---|
| 86 | 0 |
| 240 | 0.1-5 |
| 803 | 5.1-10 |
| 54 | 10.1-15 |
| 331 | 15.1-20 |
| 162 | 20.1-25 |
| 34 | 44 |

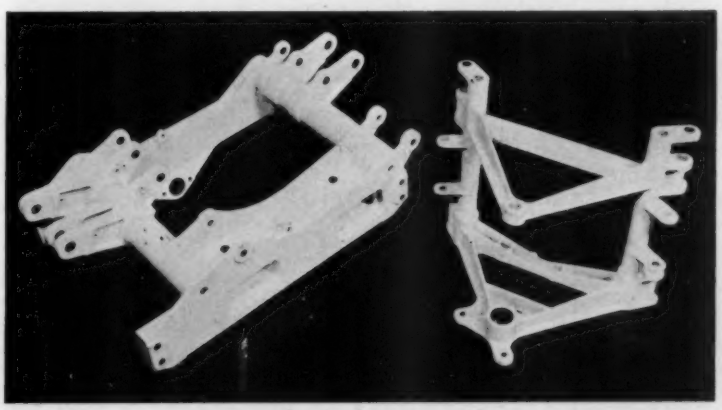


Fig. 3 — Typical flight control castings of AZ91 magnesium in a jet liner.

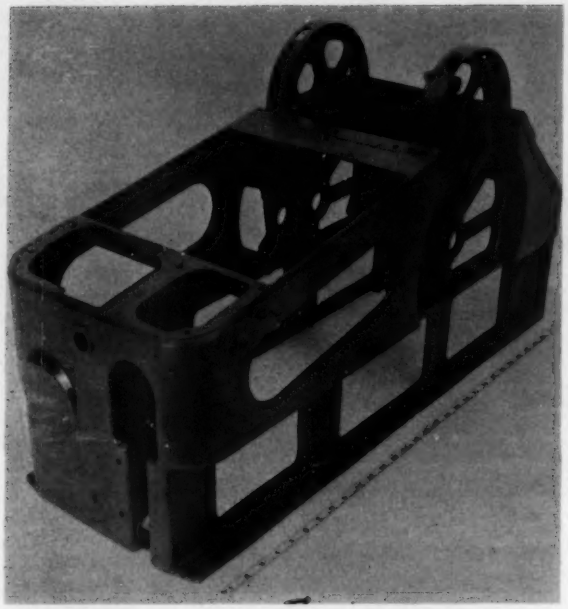


Fig. 4 — Pilot's control pedestal casting in AZ91 magnesium in a jet liner.

TABLE 3 — CASTING REJECT DISTRIBUTION

| Defect Type | Radiographic Rejects, % |
|---|---|
| Shrink cavities | 36.7 |
| Segregation | 26.5 |
| Gas holes | 9.4 |
| Microshrink | 7.7 |
| Dross and inclusions | 7.7 |
| Cracks | 2.5 |
| Hot tears | 2.5 |
| Cold shut | 1.7 |
| Misruns | 0.8 |
| All others | 4.5 |

tors of lower design allowables to account for casting defects in critical design applications rule out many castings. This is especially true of large castings in dense materials like steel, where lower design allowables to account for defects add weight in rapid fashion. It is an unfortunate circumstance because improvements in casting techniques would open the gates to wider casting acceptance than now enjoyed. The principle deterrents are — (1) weight penalties to compensate for casting defects and (2) defects affecting fatigue life.

## ALTERNATE SOLUTIONS

The present status of castings forces companies into other design solutions. Figure 5 is an example of a 4340 machined steel forging with built in cost. It went the forging route because other methods like weldments and present casting techniques did not meet the company's design philosophy. The forging was machined to size in almost its entirety at a strength level of approximately 125,000 psi. Minor machining followed after final heat treatment of 200,000 psi. The rough forging weighed 52 lb. Machining operations generated 40 lb of chips to make a finished part weighing 12 lb. Approximately 12 hr was required for the machining operations, 8 hr of which consisted of pocket end milling.

The forging (Fig. 5) is an interesting study in pointing up two significant cost factors. (1) Large losses in material. The case at hand shows a 77 per cent material loss in the form of chips having only scrap value. A study of 28 other steel forgings confirmed similar losses. 383 lb of rough forgings yielded 87 lb of finished parts. Titanium forgings also have about the same yield. (2) The high cost of shaping to size a part without undue complexity in a material with a reasonable machinability index.

Both these factors have increasing significance as the industry moves ahead in design applications of materials difficult to produce in the original configuration and in the final machined condition.

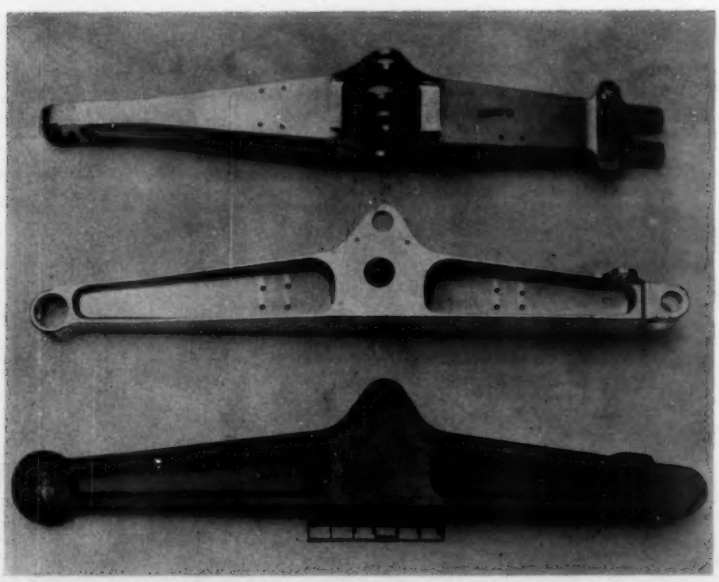


Fig. 5 — Forged 4340 steel wing flap track carriage.

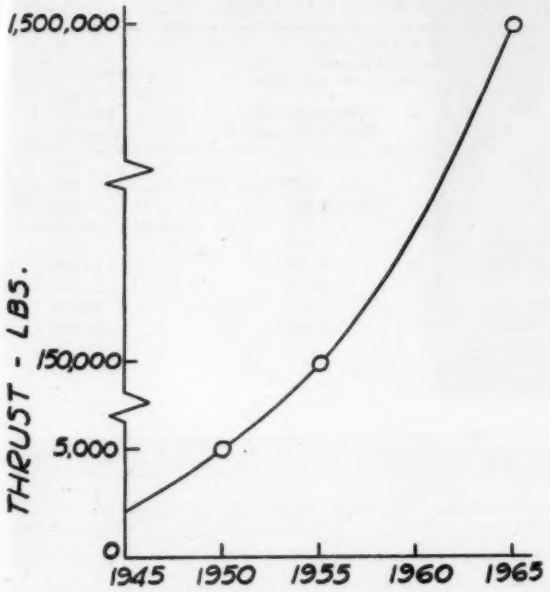



Fig. 6 — Power plant trends.

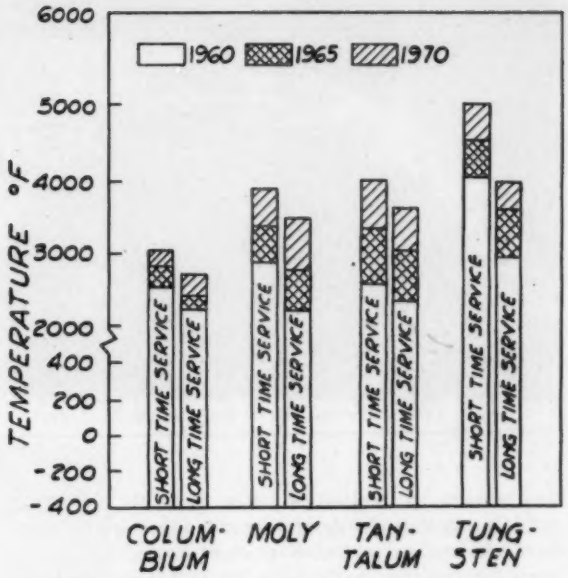



Fig. 8 — Forecast of thermal capability of refractory metals.[1]

## CASTING NEEDS

The hindrances to wider acceptance of aerospace castings require early solution as future requirements are envisioned. Technical feasibility is well established for commercial travel and deep space exploration at many times the speed of sound. Events of this nature are, in part, due to the rapid developments in power. Figure 6 is a graphic representation of the power available for use over a period of years. Power plant thrust will increase from 3000 lb in 1945 to 1,500,000 lb in 1965. Developments of this nature have suddenly created technological deficiencies in material requirements. The Aerospace Industries Association 1960 Forecast[1] emphasized these deficiencies by the following statement:

> "A recent formal report to the National Academy of Sciences concluded that 'major end-item programs in all areas of national security (defense, atomic energy and space) are now up against a serious materials' road block. The properties of most presently available materials are inadequate for the high performance end items that must be made to withstand the severe temperature, pressure, radiation, corrosion and stress conditions of space-age environments.' — fabricability and cost are critical factors."

Figures 7 and 8 show the potential capabilities of various metals. It will be noted that growth is greatest in beryllium, titanium, stainless steel, super alloys and the refractories. There is intensive effort to clear the roadblocks to their ultimate capabilities. Comparable effort is required to make them available in the form of castings at an early date.

## DOLLAR SAVINGS FROM CASTINGS

Fabricability and cost require constant surveillance in designs. Increasing attention is being focused on

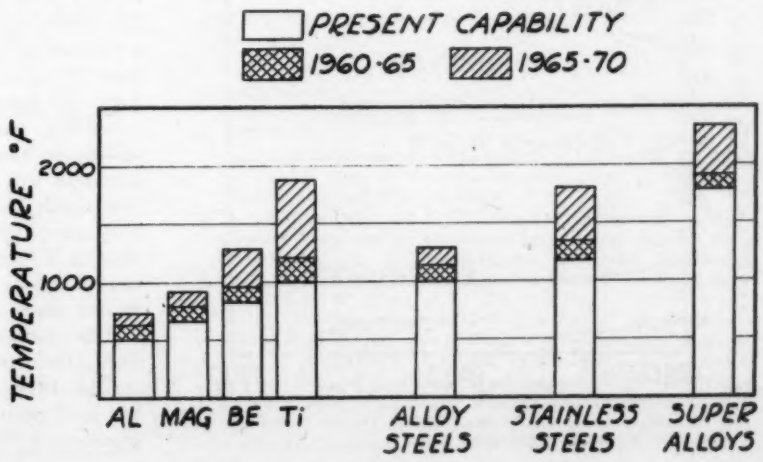



Fig. 7 — Forecast of thermal capability of metals.[1]

TABLE 4 — ESTIMATED MATERIAL COST SAVINGS FROM CASTINGS

| Material | Forging or Casting Material, Cost/lb, $ | Forging or Casting, Cost/lb, $ | Material Loss in Producing 100 Lb of Finished Parts[1] | | Dollar Savings from 100 Lb of Castings, $ |
|---|---|---|---|---|---|
| | | | Conventional Forgings with 77% Reduction, $ | Castings with 30% Reduction, $ | |
| Air Melt Steel | 0.20 | 0.75 | 250 | 30 | 220 |
| Vacuum Melt Steel | 1.00 | 1.50 | 500 | 65 | 435 |
| R-235 Steel | 5.00 | 7.50 | 2,330 | 305 | 2,025 |
| Titanium | 4.50 | 12.00 | 3,850 | 495 | 3,355 |
| Zirconium | 7.00 | 20.00 | 6,435 | 830 | 5,605 |
| Molybdenum | 25.00 | 50.00 | 14,190 | 1,830 | 12,360 |
| Tantalum | 50.00 | 125.00 | 36,675 | 4,730 | 31,945 |
| Beryllium | 60.00 | 150.00 | 44,075 | 5,675 | 38,400 |
| Columbium | 100.00 | 250.00 | 73,350 | 9,460 | 63,890 |

1. Assumed Value of Scrap Recovery:
10% Material Cost of Titanium, Zirconium and R-235 Steel.
30% Material Cost of Refractories.
Scrap Value Neglected for Air and Vacuum Melt Steel.

this subject in realization that there is a limit to available dollars. In designs ahead, the engineer must have the means, more than ever, to reduce his designs to practice with little or no metal removal. The free and easy machining properties as available with aluminum or magnesium are no longer anticipated. Rather, materials which are costly to initially shape and convert to final shape by metal removal processes are being studied.

In consideration of the final part the question is asked, "What else will do the job"? Answer to this question must consider bits and pieces assembled by mechanical fasteners, weldments, brazing, intermetallic bond; forgings and castings. No one method of part manufacture will do all jobs in the best possible manner. All types are needed. Castings become desirable with increase in shape complexity, and when other design solutions are more costly.

Even with moderate complexity where a forging could be used, a high cost penalty from material losses in conversion to the finished part is found. This is particularly true in the newer materials. For example, large forgings in columbium are limited to simple shapes, such as the pancake type with little definition. Similar conditions prevail with forged tantalum. Tungsten has not been moved into much shape by conventional forging practices.

Table 4 is a summary of cost savings in material to be realized by good castings replacing forgings at the present state of development. As an example, savings of $63,000 are possible from 100 lb of finished columbium castings. If a 50 per cent reduction in columbium material costs is considered, the savings would amount to approximately $32,000. The estimates consider 77 and 30 per cent reduction in forging and casting weights, respectively, by metal removal processes. Such savings must receive more than casual attention. They are possible because of inherent capabilities of the casting process to produce parts to or nearly to final size. Further, castings offer an additional bonus by reason of less cost in shaping to final size. Clearly, there are good reasons for accelerating casting developments. The nation's defense budget needs this work to reduce application cost of expensive materials.

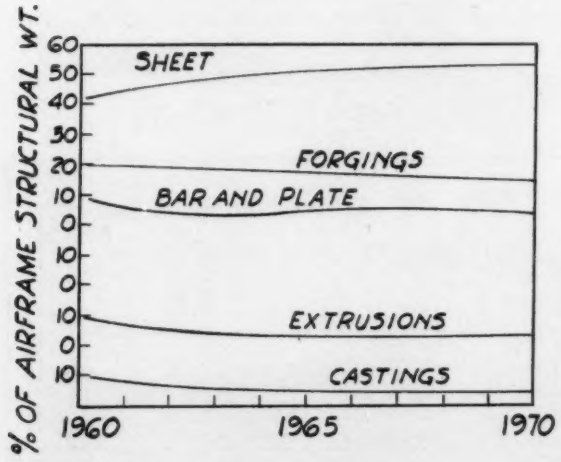



Fig. 9 — Forecast of material forms.[1]

## OUTLOOK

There is mixed opinion about future casting requirements. The Aerospace Industries Association 1960 Forecast[1] of various material forms is given in Fig. 9. A decrease in casting uses is forecast. On the other hand, Battelle Memorial Institute[2] reports a forecast of increased steel casting uses from a statement by a Materials Advisory Board Panel on Castings and Powder Metallurgy.

Casting requirements are subject to upward revisions by breakthroughs in casting technology. Breakthroughs will come from those with creative imagination unshackled by convention. Preconceived ideas and stereotyped approaches will not bring desired results. The technical approach to fundamental casting processes is required to bring about solutions rather than improvements in the art of casting.

The outlook is encouraging for high quality castings produced by improved technology. Sign posts in this direction are vacuum degassing, vacuum melting and pouring processes to improve homogeneity, elongation, fatigue and impact properties,[3] new tech-

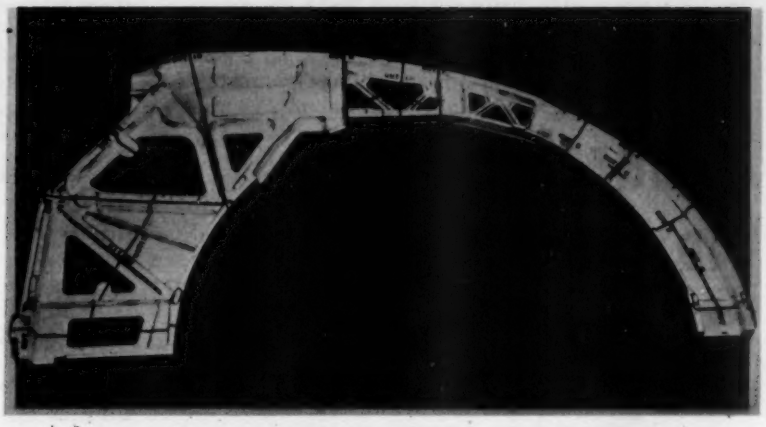


Fig. 10 — Forged 7075 aluminum alloy fuselage bulkhead 66 in. in height for an interceptor airplane.

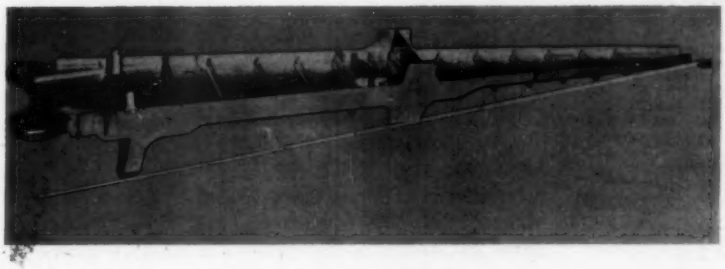


Fig. 11 — Forged 7075 aluminum alloy wing spar 10 ft in length for an interceptor airplane.

niques in melting practices,[4] mold developments, new approaches to formed shapes, such as sintering of pressed metal powders and slip castings, possibilities of grain refinement by ultrasonic energy and metallurgy best suited for casting processes.

The slip casting process is cited as an example of encouragement in casting development. A ½-in. diameter bar approximately 6 in. in length was slip cast in type 316 stainless steel alloy by the author's company.[5] Table 5 summarizes the results.

Casting sizes and complexity similar to the forgings of Figs. 10 and 11 could be used to advantage. Until this plateau of development is reached, there are innumerable casting requirements in the smaller sizes due to high density of aerospace vehicles. Quality comparable to good wrought products is desired, close dimensional control and minimum section thickness of 0.060 in. to reduce the compounding effects of weight and cost of shaping to final size. The achievement of these objectives cannot be said to be impossible by application of growing technology.

TABLE 5 — TYPE 316 STAINLESS STEEL SLIP CASTING PROPERTIES

| | |
|---|---|
| **Physical Properties** | |
| Density of sintered tensile bar, slip casting | 7.9 gr/cc |
| Theoretical density, wrought | 7.86-7.94 gr/cc |
| **Mechanical Properties** | |
| Slip Casting | |
| Yield strength, psi | 27,100 |
| Ultimate strength, psi | 74,800 |
| Elong., % in 2 in. | 57.5 |
| Wrought | |
| Yield strength, psi | 32,000 |
| Ultimate strength, psi | 77,000 |
| Elong., % in 2 in. | 50.0 |

## CONCLUSIONS

The aerospace industry is one of dynamic action, and this type of energy is needed in all industries supporting the product. The casting industry has an important role in this support.

Maximum application of castings has been restricted by certain chronic defects. Alternate design solutions are used for reasons of weight saving and structural reliability. The solutions are not always the most economical approach, which condition becomes increasingly important in application of materials to withstand severe temperature environments.

Castings are a necessary form of structure in beryllium, titanium, stainless steels, super alloys and the refractories. They offer large potential savings in material cost and shaping to final size by metal removal processes.

The outlook for high quality castings is encouraging, because of recent technical developments in fields related to foundry techniques.

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# MAGNESIUM-SILVER-DIDYMIUM-ZIRCONIUM CASTING ALLOY QE22A

by D. J. Whitehead


## ABSTRACT

**Magnesium casting alloy QE22A* (Mg-2.2 per cent Di-2.5 per cent Ag-0.6 per cent Zr) combines high yield strength at room temperature with good tensile properties at temperatures up to 600 F (316 C). The alloy also possesses good fatigue characteristics, is free from stress corrosion troubles, and has good long term creep strength at temperatures up to about 400 F (204 C).**

**QE22A has similar castability to that of other rare earth containing magnesium alloys and is weldable by the argon-arc or heli-arc processes. The alloy shows excellent agreement between the tensile properties of separately cast test bars and test pieces cut from the casting. QE22A should find application for high quality premium grade castings for the aircraft and missile industries.**


## INTRODUCTION

There are today a considerable number of magnesium alloy specifications available to users of sand and permanent mold castings. The alloys may be conveniently grouped into four categories:

1. *The classical alloys* based on the Mg-Al-Zn-Mn system, e.g., AZ91C, AZ92A, AZ63A and AZ81A.
2. *The high temperature creep resistant alloys* containing rare earth metals or thorium and grain refined with zirconium, e.g., EZ33A, HZ32A and HK31A.
3. *The high yield strength Mg-Zn-Zr alloys*—ZK51A.
4. *The hybrid alloys* such as ZH62A and ZE41A, containing substantial amounts of zinc together with thorium or rare earth metals and zirconium. They combine some of the advantages of groups 2 and 3.

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*The composition of QE22A is covered by U.S. Patent No. 2,549,955.

### *Aircraft Industry Alloys*

The alloys described in the present paper have fairly expensive alloying ingredients and along with those in groups 2, 3 and 4 are mainly directed to the aircraft industry. Although they bear some resemblance to group 4 alloys, they break new ground in that they are based on a new alloy system namely, magnesium-didymium-silver-zirconium. They take advantage of three effects.

Only the second of these effects is novel, its discovery being made in England by R. J. M. Payne and N. Bailey[1] of J. Stone & Co. (Charlton) Ltd.

There is a considerable patent literature on magnesium-base alloys containing the individual rare earth metals in various proportions and Leontis,[2,3] and Leontis and Feisel[4] have discussed their properties. Alloy EK31XA (magnesium-3 per cent didymium, 0.6 per cent zirconium)[5] was the first attempt to make commercial use of an alloy containing fractionated rare earths as distinct from ordinary cerium mischmetal. Payne and Bailey[6] describe the effects of silver additions to such alloys, and the British alloys MSRA and MSRB, and their U.S. equivalent QE22A, are the commercial result of this work.

The alloys show high yield strength, good tensile strength and fatigue properties which are maintained up to 600 F (316 C). The composition ranges be-

ing worked to and the respective specification minimum tensile properties are given in Table 1.

The casting qualities of the alloys are generally good and similar to those of other rare earth containing magnesium alloys. The alloys are used in the solution heat treated and aged condition (T6) only.

## TESTING PROCEDURES

The standard British separately cast test bar for sand cast magnesium alloys was used for the majority of the tests recorded in this paper. This is a one in. round, vertically cast test bar (Form B, British Standard 2L.101), and a 0.564 in. diameter, 2 in. gage length specimen is machined from it for tensile test. With some magnesium casting alloys rather higher tensile properties are obtained with the cast to size A.S.T.M. type 0.5 in. diameter test bar than with the British bar. British practice calls for the determination of 0.1 per cent yield stress (proof stress) rather than 0.2 per cent yield stress. Where 0.1 per cent yield stress values are quoted, an approximate conversion to the 0.2 per cent offset value may be obtained by increasing the former values by 10 per cent.

Elevated temperature tensile and creep tests were carried out in accordance with British Standards 1094 and 1687. Elevated temperature yield strengths were determined at a strain rate of 0.0005 in./in./min, and tensile strength at a strain rate of 0.10 in./in./min. The soaking time at temperature prior to tensile testing was one hr and 16 hr for long term creep testing.

Melting and alloying procedures for Mg-Ag-Di-Zr alloys are similar to other Mg-RE-Zr alloys. Melts were prepared from approved pre-alloyed ingots to Specifications DTD5025 and DTD5035 or alternatively from 99.95 per cent purity magnesium. Silver was added as pure bar silver and didymium as a magnesium-didymium hardener alloy. The didymium contained 80-85 per cent neodymium, 15-20 per cent praeseodymium. Zirconium was alloyed via a magnesium zirconium hardener.

## MECHANICAL AND PHYSICAL PROPERTIES

### *Tensile Properties*

It will be noted from Table 1 that the British alloys MSRA and MSRB have chemical compositions which differ only by about one per cent didymium. The alloys are comparatively insensitive to increase in silver content when this exceeds 2 per cent, but changes in didymium have fairly marked effects on properties. Figure 1 illustrates the effect of didymium content on the properties of British separately cast test bars containing 2.5 per cent Ag and 0.6 per cent Zr.

Although maximum yield strength is obtained with a didymium content approaching 3 per cent, it is evident that at this level the alloy is rather brittle with considerably reduced elongation values. Maximum tensile strength is obtained at only 1.5 per cent didymium. This is the reason for the existence of two alloys in the British specifications. MSRB was selected to provide an alloy of high yield

TABLE 1 — CHEMICAL COMPOSITIONS AND MINIMUM TENSILE PROPERTIES OF MAGNESIUM-SILVER-DIDYMIUM-ZIRCONIUM ALLOYS

| Alloy | Spec. | Composition, % Di | Ag | Zr | 0.1 % yield, kpsi | 0.2 % yield, kpsi | Tensile, kpsi | El., % in 2 in. |
|---|---|---|---|---|---|---|---|---|
| MSRA, | British D.T.D., 5025 | 1.2-2.0 | 2.0-3.0 | 0.4 min. | 22.4 | | 34.8 | 4 |
| MSRB, | British D.T.D., 5035 | 2.0-3.0 | 2.0-3.0 | 0.4 min. | 24.6 | | 34.8 | 2 |
| QE22A, T6, | U.S. Tentative | 1.8-2.5 | 2.0-3.0 | 0.4 min. | | 25.0 | 35.0 | 2 |

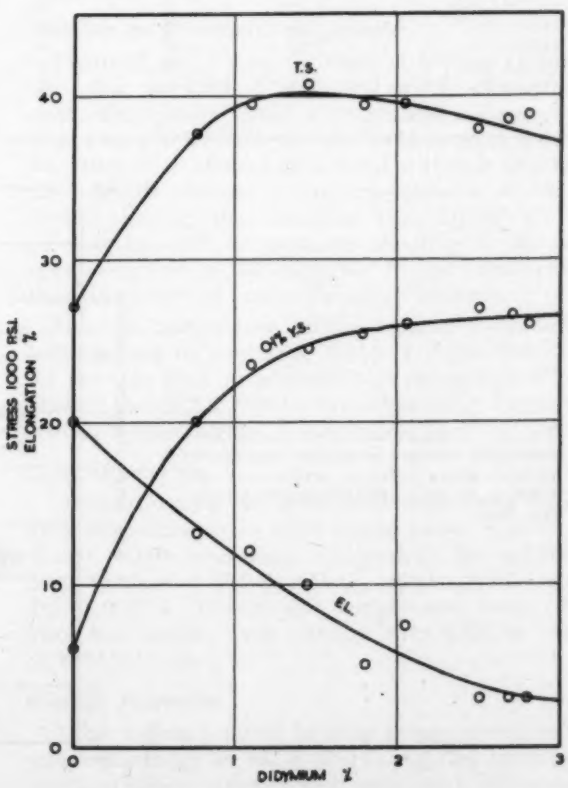



Fig. 1 — Didymium content effect on tensile properties of QE22A-T6 type alloys (British test bars).

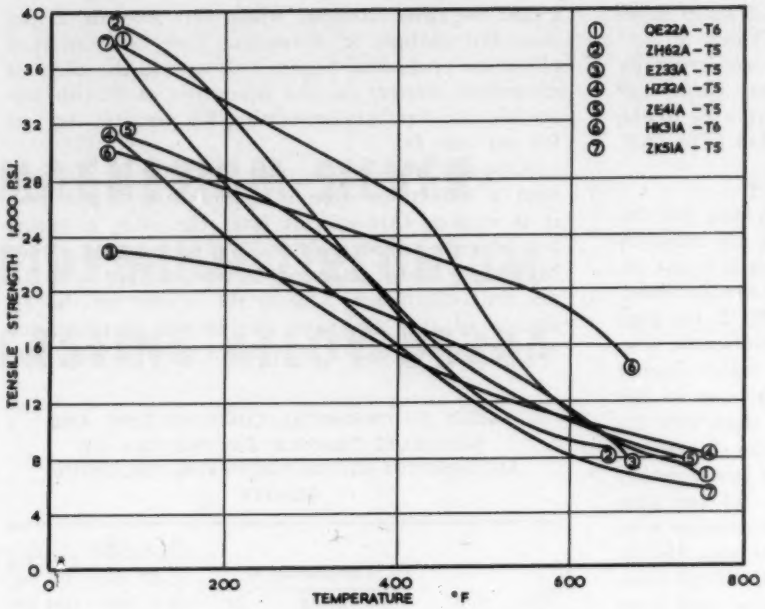



Fig. 2 — Temperature effect on tensile strength of various magnesium casting alloys (British separately cast test bars).

TABLE 2 — MINIMUM TENSILE PROPERTIES FOR VARIOUS MAGNESIUM CASTING ALLOYS (SEPARATELY CAST TEST BARS — A.S.T.M. B80 SPECIFICATION)

| Alloy | 0.2% yield, 1000 psi | Tensile, 1000 psi | Elong., % 2 in. |
|---|---|---|---|
| HK31A-T6 | 13 | 27 | 4 |
| EZ33A-T5 | 14 | 20 | 2 |
| AZ92A-T6 | 18 | 34 | 1 |
| EK31XA-T6* | 19 | 31 | 2½ |
| ZK51A-T5 | 20 | 34 | 5 |
| ZH62A-T5 | 22 | 35 | 4 |
| QE22A-T6* | 25 | 35 | 2 |
| ZK61A-T6 | 25 | 40 | 5 |

*Tentative values.

strength without particular regard to ductility while MSRA alloy is the more ductile composition retaining good yield strength. A single intermediate alloy composition (QE22A) is presently considered adequate in the U.S.A.

Table 2 lists the specification tensile values for various magnesium casting alloys, and indicates that QE22A is in the high strength category.

The rare earth metal content of QE22A alloy results in good retention of strength at elevated temperatures. Figures 2 and 3 show the effect of temperature on the tensile and yield strength of British type test bars in several commercially used rare earth

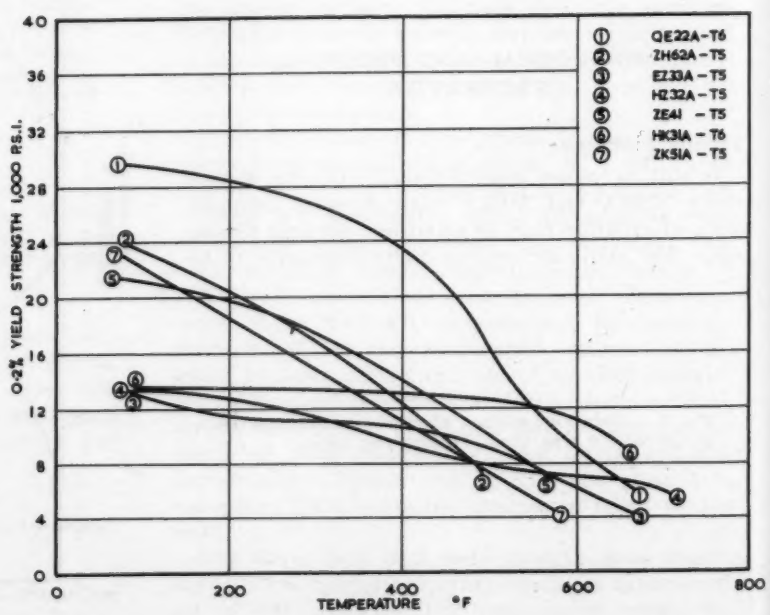



Fig. 3 — Temperature effect on 0.2 per cent yield strength of various magnesium casting alloys using a strain rate of 0.005 in./in./min (British separately cast test bars).

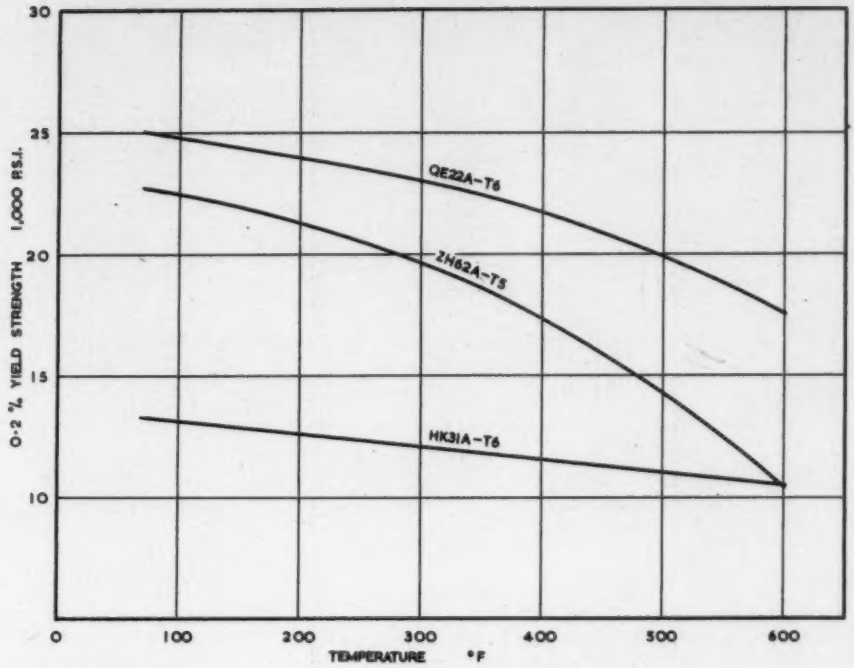



Fig. 4 — Temperature effect on 0.2 per cent yield strength of QE22A-T6, ZH62A-T5 and HK31A-T6 alloys tested at a strain rate of 0.060 in./in./min (after Gronvold[8]).

and thorium containing alloys. QE22A alloy shows the best yield strength up to 540 F (282 C) and tensile strength up to 470 F (243 C). At temperatures higher than these the more stable medium strength thorium alloy HK31A gives higher tensile properties.

The above data apply only however to tests carried out in accordance with the test procedure presented previously in this paper. Table 3[7] lists elevated temperature data on A.S.T.M. type bars using a strain rate of 0.005 in./in./min to one per cent strain and 0.1 in./in./min to fracture and with a heating time of less than one min. The shorter exposures at temperature and higher strain rate result in considerably higher strength values.

TABLE 3 — TEMPERATURE EFFECT ON TENSILE PROPERTIES OF QE22A-T6 (SEPARATELY CAST TEST BARS)

| Temp., F | No. of Tests | | 0.2% Y.S., 1000 psi | T.S., 1000 psi | Elong., % |
|---|---|---|---|---|---|
| 75 | 51 | Avg. | 29.6 | 39.2 | 3.8 |
| | | Min. | 26.9 | 37.7 | 1.2 |
| 200 | 1 | | 29.4 | 35.6 | 25 |
| 300 | 1 | | 26.6 | 31.2 | 24 |
| 400 | 1 | | 24.6 | 28.2 | 22 |
| 500 | 1 | | 20.4 | 24.4 | 37 |
| 600 | 1 | | 12.0 | 15.1 | 51 |
| 700 | 1 | | 6.1 | 7.3 | 82 |

There are a number of applications for this type of alloy for missile castings where exposure times at maximum temperature and load are short and rates of loading high. The consequences of using alloys which are not metallurgically stable at temperature may hence not be serious. In contrast to HK31A and EK31XA alloys, QE22A is particularly sensitive to rate of loading. W. Gronwold[8] and others have investigated this, and Fig. 4 compares the yield strengths of QE22A with ZH62A and HK31A at a strain rate of 0.060 in./in./min and a 30 sec soak time. Under such conditions, QE22A is the strongest magnesium casting alloy at elevated temperatures.

### *Stability at Elevated Temperatures*

Figures 5 and 6 show the effect of holding QE22A-T6 alloy at 482 F (250 C) and 572 F (300 C) on room temperature tensile strength and 0.1 per cent yield strength. The 0.1 per cent yield strength is considerably more affected than tensile strength, and the alloy rapidly overages at both temperatures. At 392 F (200 C) limited data indicate that QE22A-T6 is more stable, 350 hr exposure resulting in 80 per cent retention of the original 0.1 per cent yield strength and 90 per cent of the tensile strength.

Elevated temperature tensile properties are similarly affected by overaging. Table 4 shows how the 0.1 per cent yield stress and tensile strength at 482 F (250 C) and 572 F (300 C) are affected by exposure at these temperatures.

### *Creep Properties*

QE22A-T6 alloy has good long term creep resistance at moderately elevated temperatures. Figures 7, 8 and 9 give stress-time relationships for specified creep strains at 302 F (150 C), 392 F (200 C) and 482 F (250 C). Within this temperature range the alloy has similar creep strength over 1000 hr tests to EZ33A-T5 alloy.

### *Fatigue Properties*

Table 5 gives rotating bending fatigue strengths at temperatures up to 482 F (250 C). QE22A-T6 shows excellent fatigue strength with some notch sensitivity.

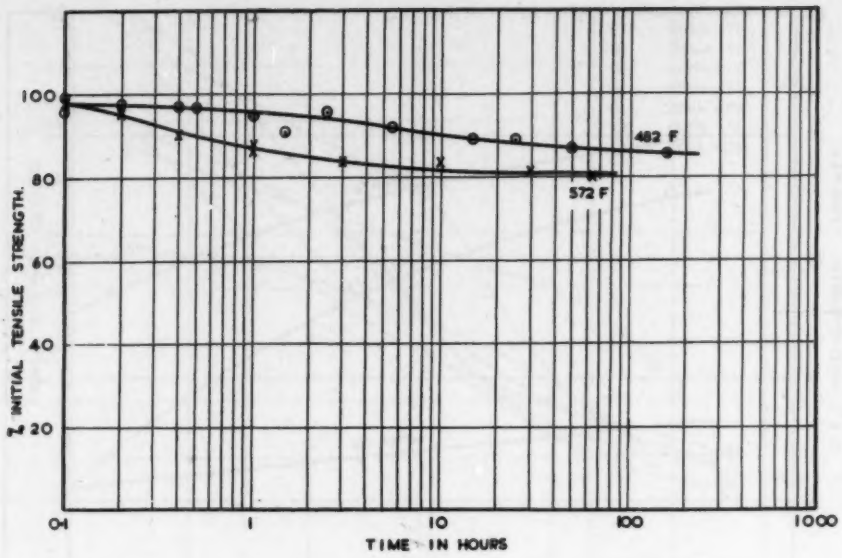



Fig. 5 — Overaging at 482 and 572 F effect on tensile strength of QE22A-T6 alloy (British separately cast test bars).

TABLE 4 — EXPOSURE AT 482 F AND 572 F EFFECT ON THE TENSILE PROPERTIES OF QE22A-T6 SEPARATELY CAST TEST BARS

| Property | Temp., F of Exposure and Testing | % of Initial Property Exposure time (hr)[a] 10 | 50 | 100 |
|---|---|---|---|---|
| Tensile strength | 482 | 87 | 75 | 72 |
| 0.1% yield stress | 482 | 85 | 69 | 66 |
| Tensile strength | 572 | 82 | 71 | 68 |
| 0.1% yield stress | 572 | 76 | 67 | 62 |

a. Includes one hr at test temperature prior to testing. British type test bars.
Strain rates in./in./min. — Yield strength 0.0005.
Tensile strength 0.10.

At moderately elevated temperature, e.g., 392 F (200 C), QE22A-T6 has the highest fatigue strength of any magnesium casting alloy so far tested.

### *Shear Strength*

Table 6 gives shear strength values for specimens machined from QE22A-T6 separately cast tests. The values are similar to recently reported results[5] for EK31XA-T6.

## HEAT TREATMENT OF QE22A-T6 ALLOY

The figures shown in Table 8 illustrate the magnitude of the effects of solution heat treatment and aging on the properties of QE22A.

The alloy is only attractive in the T6 condition.

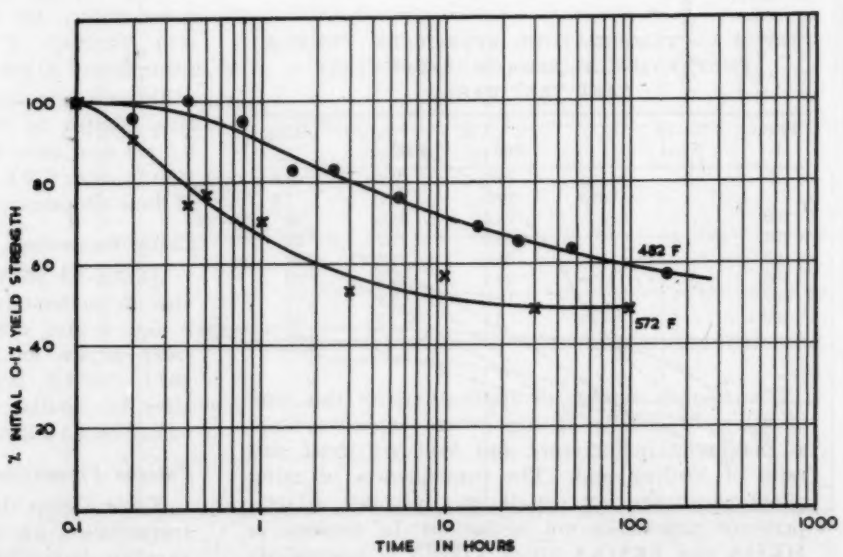



Fig. 6 — Overaging at 482 and 572 F effect on 0.1 per cent yield strength of QE22A-T6 alloy (British separately cast test bars).

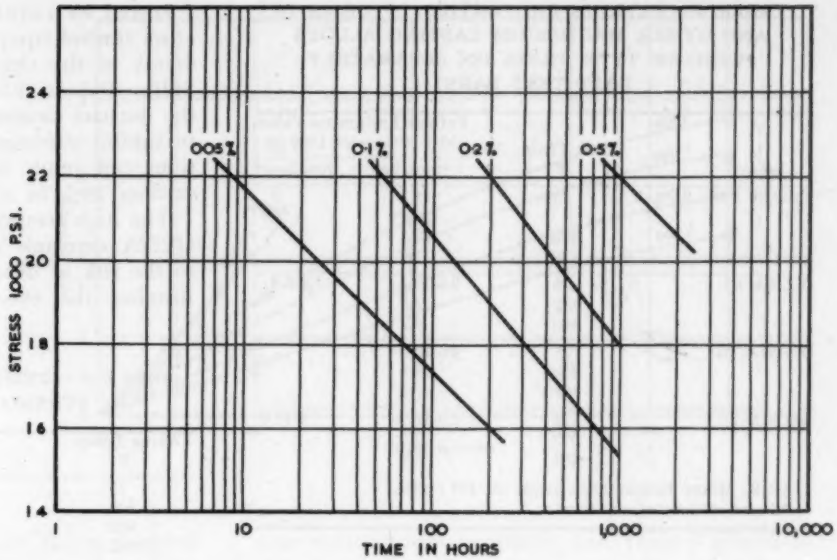



Fig. 7 — Stress/time relationship for specified creep strains at 302 F for QE22A-T6 alloy.

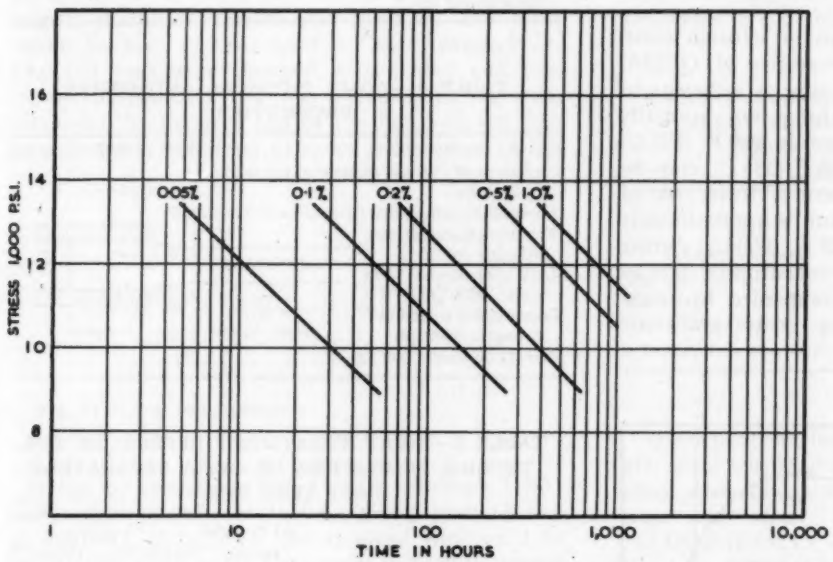



Fig. 8 — Stress/time relationship for specified creep strains at 392 F for QE22A-T6 alloys.

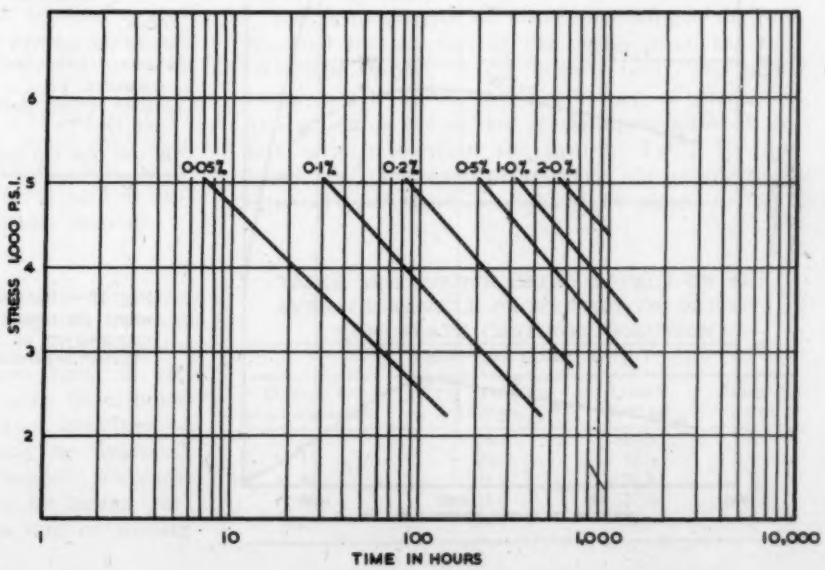



Fig. 9 — Stress/time relationship for specified creep strains at 482 F for QE22A-T6 alloy.

**TABLE 5 — FATIGUE PROPERTIES OF QE22A-T6 AND OTHER MAGNESIUM CASTING ALLOYS (WOHLER TYPE TESTS ON SEPARATELY CAST TEST BARS)**

| Alloy | Test Temp., F | Fatigue Endurance Value, $50 \times 10^6$ cycles, 1000 psi Unnotched | Notched* |
|---|---|---|---|
| QE22A-T6 | 75 | 14.5 | 9 |
| | | 13-17† | 8-10† |
| | 392 | 12.5 | |
| | 482 | 7.5 | |
| EZ33A-T5 | 75 | 9.6-10.5 | 7.4-8.3 |
| | 392 | 7.3 | |
| | 482 | 6.2 | |
| HK31A-T6 | 75 | 9.0-10.0 | |
| | 392 | 8.3 | |
| | 482 | 7.4 | |
| HZ32A-T5 | 75 | 9.5-10.5 | 7.8-10.0 |
| | 392 | 7.5 | |
| | 482 | 5.6 | |

†R.R. Moore fatigue tests; stress for $10^8$ cycles.[7]
*Stress concentration factor = 2.

### *Solution Treatment*

Figure 10 shows how variations in solution treatment temperature effect the properties of QE22A-T6. Four hr solution treatment time is sufficient to achieve good properties and 8 hr is not normally exceeded. The usual temperature is 980 F (527 C) although temperatures up to 995 F (535 C) may be employed for castings where there is little risk of distortion. The solidus temperature is approximately 1022 F (550 C). Exceeding 1022 F (550 C) during solution treatment results in a catastrophic fall in tensile properties, frequently accompanied by cracking of the castings on quenching, coarse grain size and distortion.

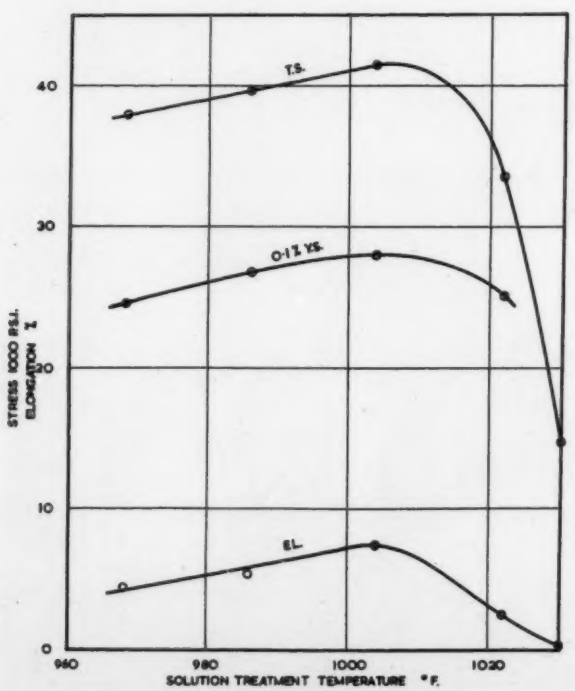



Fig. 10 — Solution heat treatment temperature (4 hr) effect on tensile properties of QE22A-T6 alloy with composition of 2 per cent Di, 2.5 per cent Ag, Zr (British separately cast test bars).

Forced air circulation furnaces with good temperature control equipment and absence of thermal gradients in the chamber are hence highly desirable. Sulfur dioxide (1.5-2 per cent) gas is introduced into the furnace atmosphere during solution treatment to inhibit oxidation. Low sulfur dioxide concentrations can result in oxidation pits forming on the castings and, in severe cases, furnace fires.

The high temperature required for QE22A demands that consideration must be given to the risk of distortion due to creep. Figure 11 illustrates the effects of temperature and time on

**TABLE 6 — TEMPERATURE EFFECT ON SHEAR STRENGTH OF QE22A-T6 ALLOY**

| Testing Temp., F | No. of tests | Double Pin Shear, 1/8-in. dia. 1000 psi |
|---|---|---|
| 75 | 2 | 23.2 |
| 400 | 2 | 18.4 |
| 500 | 2 | 15.8 |
| 600 | 2 | 11.8 |

Data source ref. 7.

**TABLE 7 — SOME PHYSICAL AND OTHER PROPERTIES**

| | |
|---|---|
| Specific Gravity | 1.82 (0.0655 lb/cu in.) |
| Coefficient of thermal expansion per degree C, 20-200 C | $26.7 \times 10^{-6}$ |
| Thermal Conductivity CGS units 20 C | 0.27 |
| Electrical Conductivity mho/cu cm 20 C | $14.6 \times 10^4$ |
| 0.1% compressive yield stress, 1000 psi | 22.5-27 |
| Compressive ultimate strength, 1000 psi | 45-55 |
| Brinell hardness, 500 kg | 78 |

**TABLE 8 — HEAT TREATMENT EFFECT ON THE TENSILE PROPERTIES OF QE22A SEPARATELY CAST TEST BARS***

| Condition | 0.1 % yield stress, 1000 psi | Tensile Str., 1000 psi | Elong., % 2 in. |
|---|---|---|---|
| As-cast | 14.8 | 25.4 | 7.5 |
| As-cast and aged T5 | 17.9 | 24.2 | 2 |
| Solution treated and quenched T4 | 11.9 | 30.0 | 15 |
| Solution treated, quenched and aged T6 | 26.0 | 39.0 | 4 |

*British type test bars.

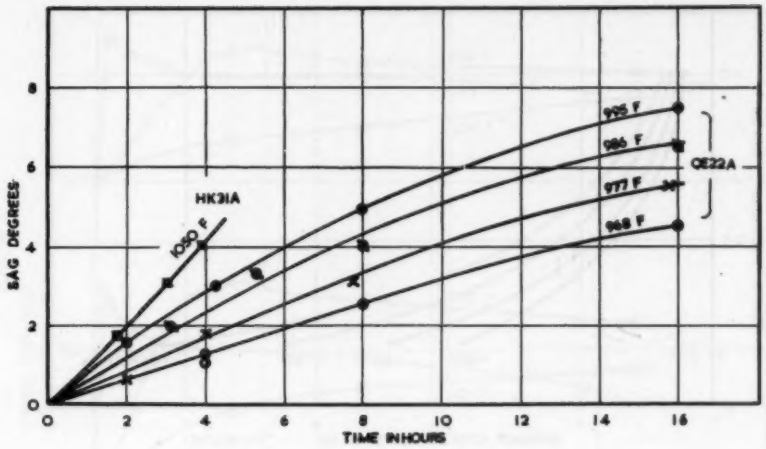



Fig. 11 — Sag of QE22A and HK31A alloys at solution treatment temperatures.

the distortion of QE22A and HK31A during solution treatment. A standard 0.564 in. diameter, 2 in. gage length, machined tensile test piece of the form, shown in Fig. 12, was used for this investigation. The test bars were clamped at one end and heat treated horizontally. Sag was measured as the angle subtended by the free end of the bar with the horizontal after air cooling to room temperature.

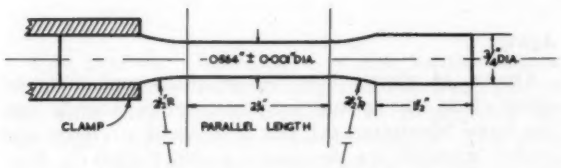



Fig. 12 — Sag test specimen.

It can be seen from Fig. 11 that although HK31 creeps more rapidly at 1050 F (566 C) than QE22A at 968-999 F (520-537 C), the accepted time of 2 hr for solution treating HK31A results in slightly smaller deflections than occur with QE22A solution treated 8 hr at 968-986 F (520-530 C). However reduction of the solution treatment time of QE22A to 4 hr at 977 F (525 C) gives a similar level of distortion to that applying to HK31A. The sag of both HK31A and QE22A is appreciably worse than that of AZ81A heat treated at 780 F (416 C), and necessitates considerably more attention being paid to support fixtures when heat treating slender castings.

### *Quenching*

Rates of diffusion are sufficiently slow in most magnesium base alloys that air cooling after solution treatment is normally sufficient to guarantee adequate response of the alloy in subsequent aging at elevated temperature. Busk and Anderson[9] have shown however that the tensile and fatigue properties of magnesium-aluminum-zinc alloys may be improved by water quenching, and K. E. Nelson[5] advocates accelerated cooling of EK31XA alloy for similar purposes. QE22A alloy is sensitive to rate of cooling after solution heat treatment, and Table 9 illustrates that water quenching is necessary if the best tensile properties are to be realized.

TABLE 9 — TENSILE PROPERTIES OF QE22A-T6 ALLOY SEPARATELY CAST TEST BARS[a] QUENCHED FROM 980 F AND AGED 8 HR AT 392 F

| Cooling Medium | Average Tensile Properties 0.1% yield stress, 1000 psi | Tensile str., 1000 psi | Elong., % |
|---|---|---|---|
| Still air | 20.8 | 33.6 | 3.5 |
| Air blast | 24.0 | 36.2 | 3.0 |
| Water (150 F) | 25.6 | 39.2 | 3.8 |

a. British type test bars.

Quenching in light oil has also been found to give good tensile properties, and cold water quenched alloy does not give appreciably higher properties than hot water quenched material. Water at 140-180 F (60-82 C) is hence the normal quench medium for QE22A-T6.

The main variable on water quenching is the time required for transferring the charge from the heat treatment furnace to the quench tank. The major factor which affects the significance of prequench air cooling is the section thickness or ratio of surface area to mass of the casting. Table 10 illustrates how prequench delays affect the tensile properties of ½-in. sand cast QE22A-T6 plate.

TABLE 10 — QUENCH DELAY EFFECT ON THE AVERAGE TENSILE PROPERTIES OF QE22A-T6 ½-IN. PLATE CASTINGS. SOLUTION TEMPERATURE 975 F

| Delay in sec prior to quench | 0.1% yield stress, 1000 psi | Tensile, 1000 psi | Elong., %, 2 in. |
|---|---|---|---|
| 0 | 26.2 | 38.5 | 5 |
| 15 | 25.6 | 37.0 | 3 |
| 30 | 25.8 | 37.2 | 3 |
| 60 | 25.6 | 37.0 | 3.5 |
| 120 | 24.4 | 34.1 | 5 |

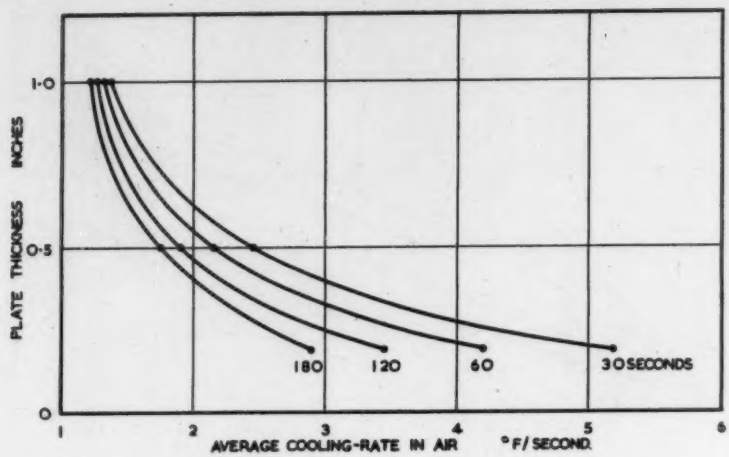



**Fig. 13 — Section thickness effect on air cooling rate of sand cast and solution heat treated QE22A plate.**

Similar results were obtained with one in. round test bars and one in. plate, and provided such material is not allowed to air cool below about 850 F (454 C) within a time limit of one min prior to quenching the delay is of little consequence.

With casting sections of less than 0.5 in. rates of air cooling are appreciably faster and the subsolvus temperatures attained with quench delays may be correspondingly lower. A general limitation of 30 sec on the prequench delay is hence considered good practice for QE22A alloy. Figure 13 shows how thickness affects the average cooling rates of QE22A sand cast plates over specific quench delays. The plates were solution treated at 975 F (524 C) and suspended in still air on removal from the heat treatment furnace.

The extreme case of a $\frac{3}{16}$-in. plate air cools to 820 F (438 C) in 30 sec. However, the rate of cooling is sufficiently rapid to preserve the solid solution, and even after 180 sec air cooling, i.e., down to 450 F (454 C), a 0.1 per cent yield strength of 24,600 psi was obtained compared with 27,400 psi for the immediately quenched material and 25,300 psi for plate quenched with a delay of 30 sec. The fast rates of cooling associated with water quenching (of the order of 80 F/sec) are hence not essential with QE22A, and other forms of rapid cooling, e.g., water spray, may be used, and provided the average cooling rate is around 5 F/sec good properties will be obtained.

The difficulties with such methods of cooling are those of achieving sufficient chill on thick section castings, and reproducible results with variable shapes and heat treatment loads. Founders hence prefer to water quench, and frequently use existing water quench facilities used for aluminum-base alloys.

The possibility of cracking and distortion must always be considered when water quenching castings. QE22A alloy however is not susceptible to quench cracking, provided the solution heat treatment temperature does not exceed the solidus.* Warpage on quenching has not presented a problem with normally designed castings. Laboratory tests on thin plates of ¼-in. and less have shown that distortion can occur, some of which has been due to changes occasioned by solution treatment and the release of casting stresses. Some difficulties may hence be encountered with thin slender castings of appreciable area. However, as indicated previously, less drastic cooling methods may be employed with such castings.

*Notched test panels of the type described by Busk and Anderson[9] fail to crack in QE22A alloy when quenched into cold water from temperatures below 1020 F (550 C).

### *Aging*

Figure 14 shows how temperature and time of aging affect the tensile properties of separately cast test bars. Maximum 0.1 per cent yield strength and tensile strength are developed at 392 F (200 C). Normal minimum and maximum aging times are 8 and 16 hr, and the alloy is air cooled after aging. It will be observed that appreciable gains in ductility of the fully hardened alloy can only be obtained by underaging and at considerable sacrifice of yield strength.

### *Metallography of Heat Treated QE22A*

No extensive study of the precipitation process in QE22A alloy has been made. Acetic glycol is a useful grain boundary etchant, but the alloy stains readily due to its silver content and structures within the grain should be interpreted with caution. The as-cast structure (Fig. 15) resembles EK41 and EK-31XA alloys. Solution heat treatment and quenching results in partial solution and coalescence of the magnesium-didymium eutectic phase. Traces of ring-like precipitate are frequently also visible within the grain, and are associated with zirconium or zirconium hydride (Fig. 16).

The effects of overheating are readily revealed by grain coarsening and the curved edges of the eutectic phase (Fig. 17). Inefficient quenching is revealed at high magnifications by the presence of fine precipitate in the grain boundaries (Fig. 18). QE22A alloy must normally be seriously overaged before definite signs of didymium compound precipitation are visible, and it is not usually possible to check aging procedures by microscopic examination.

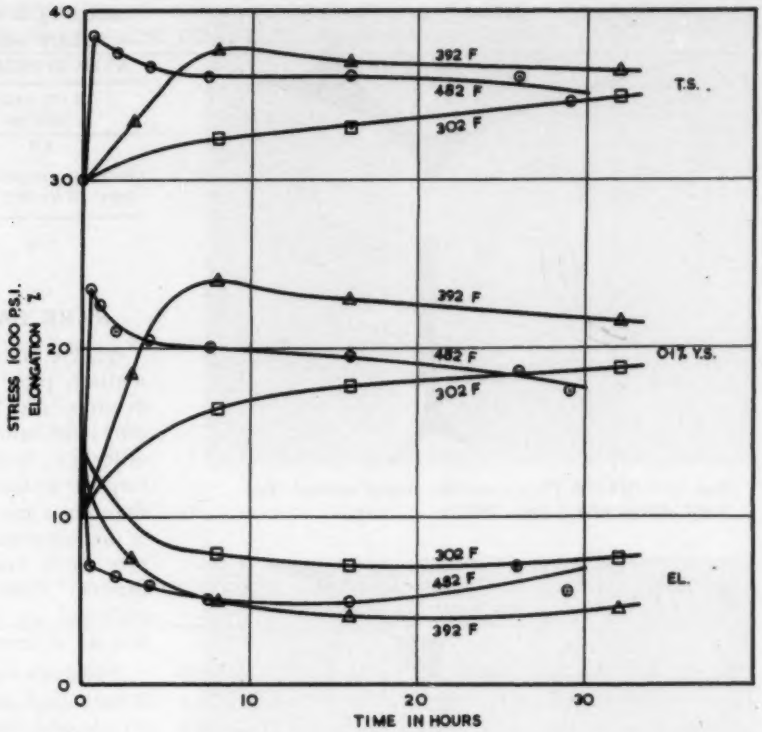



Fig. 14 — Time and temperature of aging effect on properties of QE22A-T6 alloy (British separately cast test bars).

The presence of hydrogen in the furnace atmosphere must be avoided during solution treatment or the alloy will fail to respond to subsequent age hardening and have low yield strength. The presence of 1½-2 per cent sulfur dioxide in the furnace atmosphere is sufficient to avoid trouble due to water vapor. When oxidation pits do form on castings however due to incorrect furnace atmosphere, local abnormal microstructures will be found in the metal immediately adjacent to them (Fig. 19). The grain boundaries are abnormally free to didymium eutectic and heavy precipitation occurs within the grain. Solution heat treatment in hydrogen gives similar structures and low tensile properties. (Table 11).

The effect is thought to be due to precipitation of rare earth and zirconium hydrides during solution treatment. Similar effects have been observed with HK31A-T6 alloy.

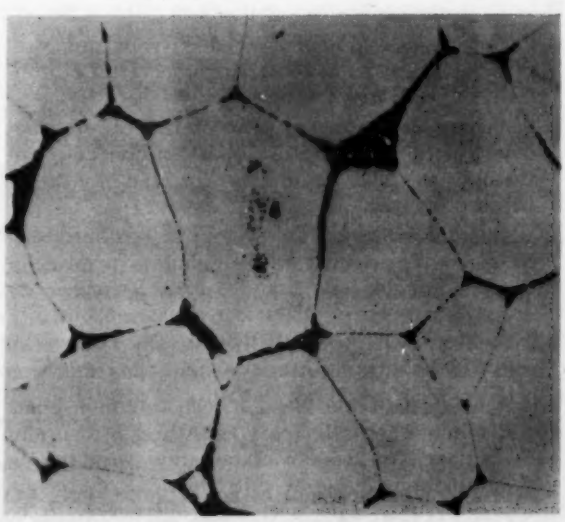


Fig. 15 — QE22A as-cast. Acetic glycol etch. 600 ×.

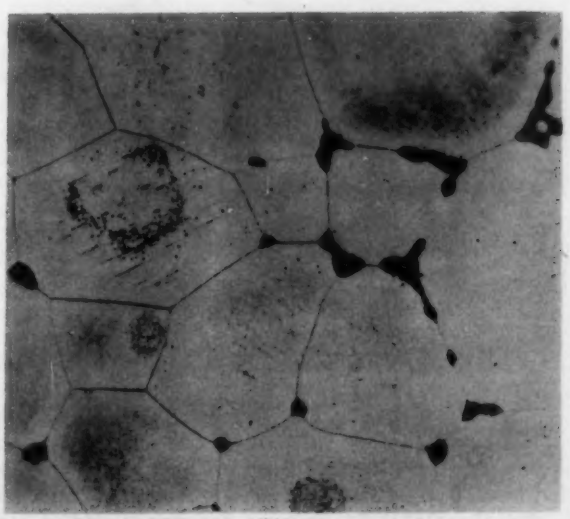


Fig. 16 — QE22A-T6. Acetic glycol etch. 600 ×.

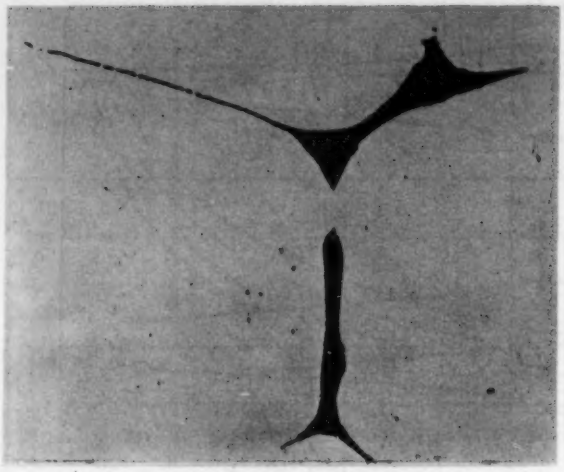


Fig. 17 — QE22A-T6, overheated during solution treatment. Acetic glycol etch. 600 ×.

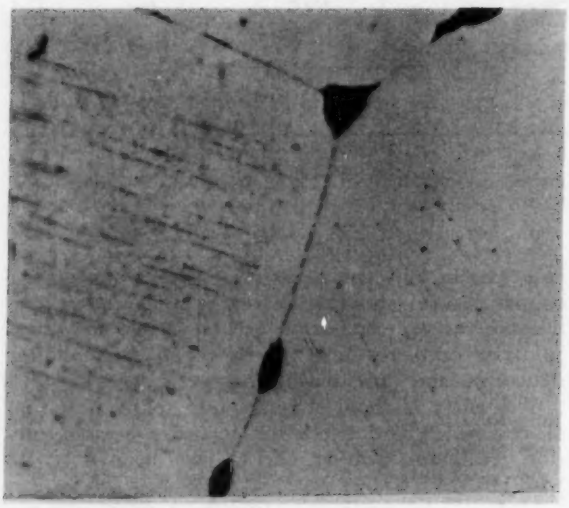


Fig. 18 — QE22A-T6, air cooled from solution treatment temperature. Acetic glycol etch. 1500 ×.

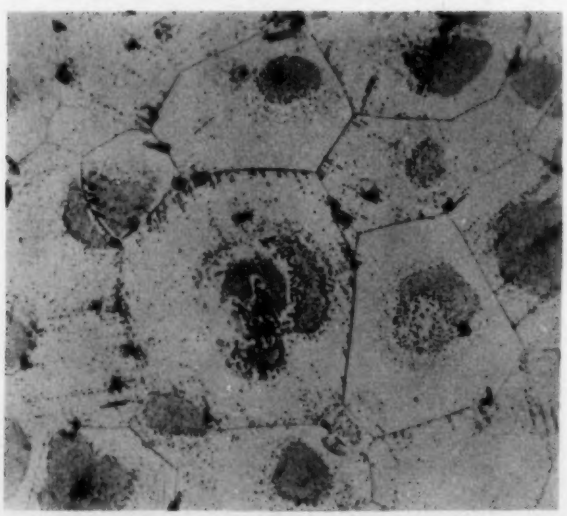


Fig. 19 — QE22A-T6, structure underlying oxidation pit. Acetic glycol etch. 600 ×.

TABLE 11 — TENSILE PROPERTIES OF QE22A-T6* ALLOY SOLUTION TREATED IN HYDROGEN ATMOSPHERE. (SEPARATELY CAST TEST BARS)

| 0.1% yield, 1000 psi | Tensile, 1000 psi | Elong., %, 2 in. |
|---|---|---|
| 9.9 | 28.2 | 15 |

* Solution treated 7 hr 986 F in hydrogen, hot water quenched, aged 16 hr 392 F in air.

## RARE EARTH METALS COMPOSITION

QE22A alloy has a rare earth metal content of essentially pure didymium comprising 80 per cent neodymium and 20 per cent praeseodymium. The cerium, lanthanum and samarium contents are of a low order, i.e., less than one per cent each, and use of a rare earth metal reasonably low in cerium and lanthanum is necessary if the best tensile properties are to be obtained. The number of sources of suitable rare earth metal available at an economic price is limited. Table 12 illustrates the tensile properties obtained on QE22A-T6 type alloys prepared from several different rare earth mixtures.

The tests indicate clearly the importance of a relatively high neodymium content for obtaining good tensile and yield strengths. However there is nothing to be gained by using a rare earth source with a neodymium content in excess of 80 per cent, and indeed there is a tendency for elongation values to fall at high neodymium content. The didymium mischmetal 2 was interesting in that it gave good tensile properties in the 1.2-2.0 per cent RE range with good ductility. It has been observed in particular that this material gives rather higher elongations than 80 : 20 Nd/Pr in the range 1.6-2.0 per cent RE. However yield strength values at higher rare earth contents (2.0-3.0 per cent) are not as high, and on balance the 80 : 20 Nd/Pr (didymium A) is superior.

Certain other conclusions may also be drawn concerning the influence of the individual rare earth metals on the properties of QE22A-T6 type alloys, eg.:

1. Cerium is among the least effective of the commoner rare earths in conferring good tensile properties to QE22A type alloys.
2. Replacement of 50 per cent of the cerium content with a 50-50 mixture of didymium and lanthanum to give cerium mischmetal results in better properties.
3. Reduction of the cerium content of cerium mischmetal by replacement with didymium to give didymium mischmetals 1 and 3 gives a further increase in properties. These are enhanced, but to a lesser degree, if the lanthanum is also replaced by didymium to give didymium A.

It is hence desirable to use a relatively high grade of didymium when alloying QE22A and precautions must be taken to avoid contamination with cerium and lanthanum (e.g., by introduction of cerium mischmetal) if a high level of properties is to be maintained.

TABLE 12 — TENSILE PROPERTIES OF SEPARATELY CAST TEST BARS‡ OF QE22A-T6 TYPE ALLOYS MADE FROM DIFFERENT RARE EARTH METALS

| Rare Earth Type | Rare Earth Analysis, % | | | | | | Tensile Properties | | | | | |
|---|---|---|---|---|---|---|---|---|---|---|---|---|
| | | | | | | | MSRA type (1.7 % RE) | | | MSRB type (2.5% RE) | | |
| | Nd | Pr | Ce | La | Gd | Other R.E.s. | 0.1 % yield, 1000 psi | Tensile, 1000 psi | Elong., %, 2 in. | 0.1 % yield, 1000 psi | Tensile, 1000 psi | Elong., %, 2 in. |
| Cerium | — | — | 99.5 | — | — | — | 18.6 | 31.3 | 6 | 18.0 | 27.8 | 4.7 |
| Cerium Mischmetal | 16.9 | 5.4 | 50.7 | 27* | — | — | 19.0 | 32.3 | 6 | 20.4 | 30.3 | 3 |
| Didymium Mischmetal 1 | | 62† | 12 | 23* | — | — | 22.4 | 35.6 | 7 | 23.7 | 35.6 | 6 |
| Didymium Mischmetal 2 | 57 | 16.6 | 7.8 | 5.8 | 4.8 | 7.7 | 23.7 | 38.3 | 7.2 | 24.4 | 36.0 | 3.5 |
| Didymium Mischmetal 3 | 64 | 14 | 1 | 21* | — | — | — | — | — | 24.4 | 34.9 | 3 |
| Didymium A | 79.1 | 19.9 | 0.75 | — | — | — | 25.2 | 39.4 | 6 | 26.0 | 38.0 | 3 |
| Didymium B | 79.5 | 12.2 | 0.6 | — | — | 7.6 | 24.8 | 39.4 | 6 | 26.0 | 36.5 | 2 |
| Didymium C | 85.7 | 9.9 | 0.34 | 3.2 | — | — | 25.3 | 38.0 | 4 | 25.8 | 37.8 | 3 |
| Neodymium | 90.4 | 2.1 | 2.1 | — | — | 5.4 | 24.4 | 38.2 | 4 | 25.7 | 38.0 | 2 |

†Nd + Pr.
*Includes gadolinium and other rare earth metals.
‡British type test bars. All alloys were solution treated 8 hr at 40 F below their solidus temperatures, water quenched and aged 16 hr at 350 - 392 F.

## CASTABILITY

The term castability is a comprehensive one and includes many factors. The main alloy properties which determine the quality and ease of production of magnesium sand castings may be listed as:

1. Hot tearing tendency.
2. Susceptibility to surface sinks.
3. Incidence of microporosity.
4. Sensitivity to mass effect.
5. Tendency to oxide and other inclusions.
6. Fluidity or mold filling capacity.
7. Segregation.

It is normal to gain an assessment of the castibility of a new alloy by pouring special shapes designed to emphasize defects attributable to one or more of the above factors, and also by making and examining castings in the production foundry. The latter are usually production castings on which there is a good background of experience, and the characteristics of which are well known. Generally speaking, it is not possible or necessary to determine by laboratory test all the factors listed above. There is however now a reasonable background of casting experience on QE22A alloy which leads to the following evaluation.

### *Tendency Towards Hot Tearing and Cracking*

Like other rare earth containing alloys such as EZ33A and EK41A, QE22A alloy is not hot short, and no difficulties are likely to be encountered with cracking on complex castings using established magnesium sand foundry practices. The good weldability of QE22A is a further indication of low cracking tendency.

### *Susceptibility to Sinks*

QE22A is more prone to this type of defect than zinc bearing alloys such as EZ33A, ZK51A and AZ81A - AZ92A type alloys. Other magnesium alloys with limited freezing ranges, e.g., K1A, HK31A and EK31XA, also show surface sinks and depressions at changes in section and where massive sections solidify under conditions of poor feed. The overall assessment is that QE22A is less susceptible to sinks than K1A and HK31A.

### *Microporosity*

QE22A shows some tendency to microporosity when solidifying under conditions of poor feed, and is inferior in this respect to EZ33A alloy. K. E. Nelson[7] has classified QE22A and EK31XA alloys as being comparable to AZ91C and AZ92A alloys. The incidence of microshrinkage in QE22A is related to the didymium content, and hence to the volume of eutectic liquid available to feed the incipient pores. Cerium mischmetal is more effective than didymium in suppressing microporosity for equivalent additions to a magnesium base alloy. MSRA alloy (Di 1.2-2.0 per cent) is more susceptible to microporosity than MSRB alloy (Di 2.0-3.0 per cent), and we classify the alloys in the same general category as ZE41A. Little difficulty should be experienced in making complex castings to high radiographic standards, and English founders will undertake pressure tight work in the alloy.

QE22A is outstanding in that high properties are obtainable with thick section castings. Table 13 surveys the results of cut up tests in a variety of castings made by both U.S. and English founders. In the majority of cases the castings were gated and risered as for the normal production magnesium alloy. The test results illustrate two features, namely the relative insensitivity of tensile properties to increase in section and the presence of moderate amounts of microshrinkage. Payne and Bailey[6] have also compared the properties of MSRA and MSRB with other magnesium and aluminum-base alloys

TABLE 13 — TENSILE DATA ON TEST BARS CUT FROM CASTINGS

| Type of casting | Alloy | Weight, lb | Range of section size, in. | No. of Test Bars | Min. Tensile Properties 0.1% Y.S. | 0.2% Y.S. | T.S. | El., % | Max. Tensile Properties 0.1% Y.S. | 0.2% Y.S. | T.S. | El., % | Avg. Tensile Properties 0.1% Y.S. | 0.2% Y.S. | T.S. | El., % | Source |
|---|---|---|---|---|---|---|---|---|---|---|---|---|---|---|---|---|---|
| 1. Helicopter air frame | QE22A | 14.7 | ¼ - 2 | 9 | | 26.9 | 35.6 | 1.5 | | 29.3 | 38.9 | 4.0 | | 28.4 | 37.3 | 2.9 | Dow |
| 2. Aircraft access door | QE22A | 20.5 | ¼ - 1⅛ | 11 | | 25.8 | 35.5 | 2.0 | | 30.1 | 40.3 | 4.0 | | 28.3 | 37.2 | 2.7 | Dow |
| 3. Jet engine part | QE22A | 51.5 | ¼ - 2½ | 12 | | 27.7 | 36.8 | 1.3 | | 29.9 | 39.6 | 5.0 | | 28.9 | 37.6 | 2.8 | Dow |
| 4. Jet engine part | QE22A | 49.0 | ⅝ - 1⅛ | 10 | | 27.3 | 34.8 | 2.0 | | 29.3 | 38.2 | 4.0 | | 28.2 | 36.6 | 2.3 | Dow |
| 5. Aircraft wheel | QE22A | 132 | ⅝ - 3⅛ | 17 | | 26.8 | 33.7 | 1.0 | | 29.9 | 37.7 | 5.0 | | 28.1 | 35.7 | 1.8 | Dow |
| 6. Missile casting | MSRB | 43 | 2¼ | 16 | 22.4 | | 33.2 | 1.0 | 25.3 | | 37.2 | 3.0 | 24.0 | | 35.2 | 1.2 | M.E.L. |
| | ZK61A-T6 | 43 | 2¼ | 16 | 16.6 | | 28.4 | 1.5 | 19.8 | | 33.2 | 3.5 | 17.8 | | 30.5 | 2.8 | M.E.L. |
| 7. Electronic support casting | MSRA | 6.5 | 3/16 - ⅝ | 18 | 21.7 | | 32.5 | 2.5 | 25.1 | | 38.5 | 12 | 23.3 | | 36.3 | 7.2 | M.E.L. |
| 8. Aircraft structural | MSRA | 14.9 | ½ - 3 | 9 | 24.6 | | 37.6 | 2 | 27.1 | | 41.2 | 7.5 | 25.8 | | 39.2 | 5.4 | J. Stone |
| structural | MSRB | 14.9 | ½ - 3 | 9 | 24.8 | | 35.8 | 1 | 27.5 | | 38.0 | 5.5 | 26.2 | | 37.0 | 3.2 | J. Stone |
| 9. Half hoop | MSRA | 12.25 | 1⅛ min. | 9 | 22.6 | | 37.0 | 2.5 | 24.6 | | 39.4 | 10.5 | 23.8 | | 38.5 | 7.9 | J. Stone |
| | MSRB | 12.25 | 1⅛ min. | 9 | 24.9 | | 35.2 | 3.0 | 26.8 | | 37.7 | 7.0 | 25.7 | | 37.6 | 5.0 | J. Stone |
| 10. Aircraft structural | MSRB | — | 1¾ | 7 | 24.0 | | 33.1 | 1.6 | 26.2 | | 37.4 | 4.0 | 25.5 | | 36.0 | 2.8 | J. Stone |

All stresses in 1000 psi.

cast into 4 in. dia. x 8 in. long cylinders (Table 14).

Figures 20-22 provide a quantitative indication of the effect of section thickness on QE22A, ZK51, AZ-81A and EZ33A alloy using standard size tensile specimens machined from square section slabs. The gating and feeding system selected was such that moderately heavy porosity was obtained with alloys other than EZ33A, and the properties may be regarded as conservative for premium grade castings. The data confirm the comparative insensitivity of QE-22A to mass effect compared with other alloys such as AZ81A and ZH62A.

The susceptibility of ZH62A to mass effect has been noted elsewhere,[11] and occurs also with the other 6 per cent zinc casting alloy ZK61A. Thus casting 6 (Table 13) with a ruling section of 2¼-in. x 3 in. gives markedly inferior properties when cast in ZK61A compared to QE22A. High tensile properties are however obtainable from ZK61A with thin section castings. This is reflected in the type of separately cast test bar selected for control of melt quality, and Table 15 compares the tensile properties of A.S.T.M. and British test bars. The test bars of each alloy were cast from the same melt and heat treated together.

The rare earth content of QE22A results in some tendency towards oxide inclusions comparable to that in EZ33A and EK31XA, but less than that in the

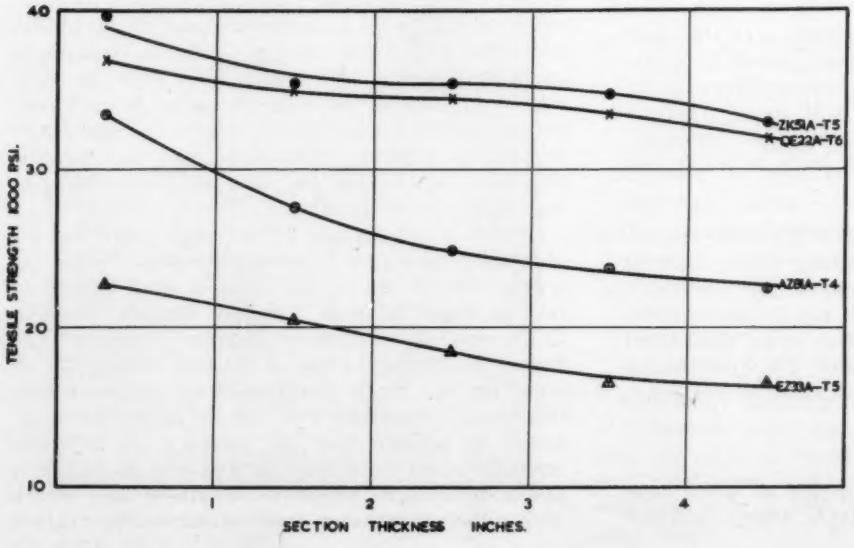



Fig. 20 — Section thickness effect on tensile strength of several magnesium sand casting alloys.

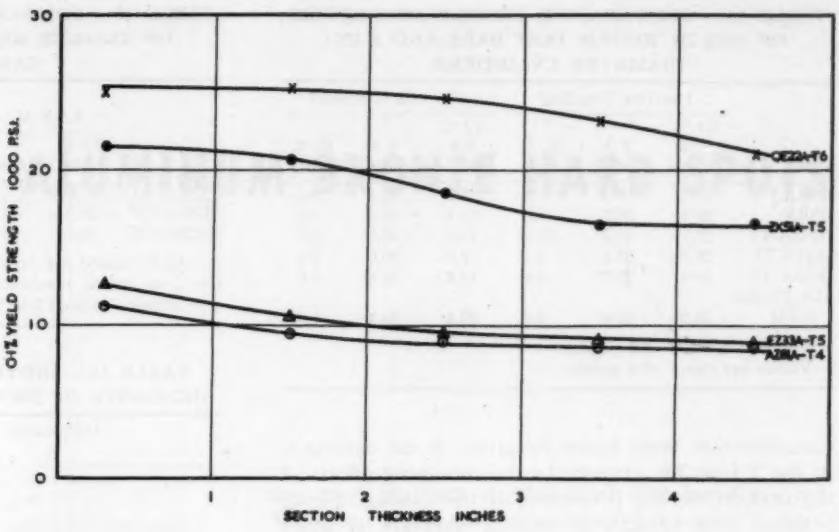



Fig. 21 — Section thickness effect on 0.1 per cent yield strength of several magnesium sand casting alloys.

thorium alloys HK31A and HZ32A. Gating systems must hence be designed with a view to achieving minimum turbulence. A tendency has been observed for oxide inclusions to increase with didymium content (Table 16).

Other inclusions which may be encountered in QE22A are associated with flux, sand and zirconium rich segregates. The incidence of flux and sand inclusions follows a similar pattern to other alloys, and is a function of technique rather than the alloy. K. E. Nelson[7] has observed that the presence of silver apparently results in a slower rate of settling of nonmetallics in the melt than with other alloys. British founders however have not encountered this tendency.

The mold filling capacity of QE22A is rated as similar to that of EZ33A and intermediate between the fluid Mg-Al-Zn-Mn alloys AZ91C, AZ81A and the sluggish thorium alloys HK31A and HZ32A.

With correct alloying and pouring techniques, the principal form of segregation met in QE22A alloy is of the dense rare earth eutectic type and similar to that described by Skelly and Sunnucks[10] and occurring in EZ33A alloy. It arises from the healing of incipient hot tears and microshrinkage by didymium rich eutectic liquid. Although no systematic investigation of the effects of such segregates on strength properties of QE22A has been made, they are known to have relatively small influence when encountered in EZ33A type alloys. The remedy for such defects normally lies in attention to gating and feeding techniques.

QE22A alloy may be welded by the argon-arc and heli-arc methods using QE22A filler rod. Minor casting defects are readily rectified by this means and high weld efficiencies are obtainable. Preheating of castings to temperatures between 400-500 F (204-260 C) is normally necessary before welding.

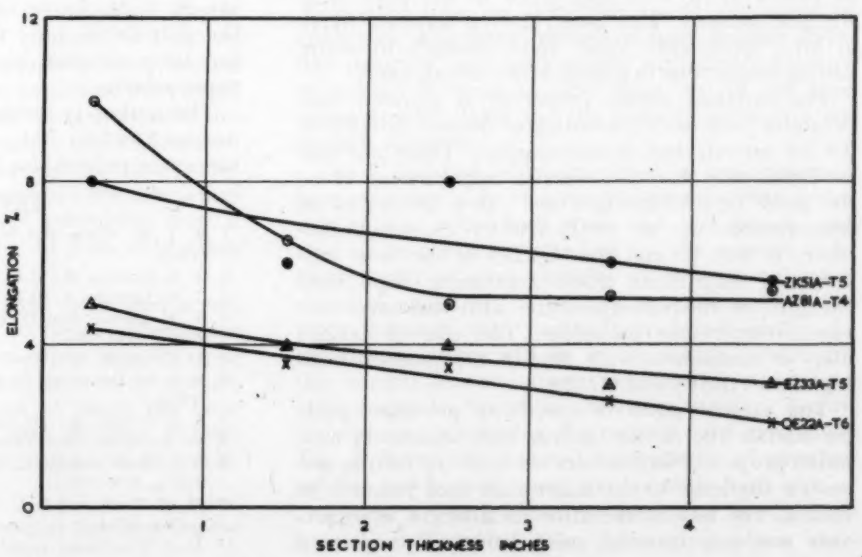



Fig. 22 — Section thickness effect on elongation of several magnesium sand casting alloys.

TABLE 14 — COMPARATIVE TENSILE PROPERTIES OF ONE IN. ROUND TEST BARS AND 4 IN. DIAMETER CYLINDERS

| | 1 in. Dia. Test Bar[1] | | | 4 in. Dia. Cylinder[2] | | |
|---|---|---|---|---|---|---|
| Alloy | 0.1 % Y.S. 1000 psi | T.S. 1000 psi | E., % 2 in. | 0.1 % Y.S. 1000 psi | T.S. 1000 psi | E., % 2 in. |
| MSRA | 27.3 | 40.5 | 7.0 | 24.2 | 36.5 | 3.8 |
| MSRB | 28.4 | 39.2 | 5.2 | 25.3 | 33.6 | 1.0 |
| ZH62A-T5 | 21.7 | 40.0 | 12.0 | 16.6 | 30.2 | 2.8 |
| ZE41A-T5 | 20.8 | 33.1 | 5.5 | 17.0 | 26.0 | 1.8 |
| ZK51A-T5 | 20.4 | 36.5 | 6.0 | 16.3 | 31.8 | 3.4 |
| Al-4½% Cu (C4A) | 28.6 | 48.8 | 9.3 | 27.1 | 36.5 | 1.0 |

1. Values are mean of 3 results.
2. Values are mean of 5 results.

TABLE 15 — COMPARATIVE TENSILE PROPERTIES OF ZK61A-T6 AND QE22A-T6 SEPARATELY CAST TEST BARS

| | A.S.T.M. type bar[a] | | | British bar (BS2L101 form B)[b] | | |
|---|---|---|---|---|---|---|
| Alloy | 0.1 % yield, 1000 psi | Tensile, 1000 psi | E., %, 2 in. | 0.1 % yield, 1000 psi | Tensile, 1000 psi | E., %, 2 in. |
| ZK61A-T6[c] | 29.3 | 45.2 | 5 | 23.7 | 39.0 | 3.5 |
| QE22A-T6 | 25.5 | 38.0 | 3 | 24.8 | 37.8 | 4 |

a — ½-in. round cast to size.
b — 1 in. round, machined.
c — Solution treated 2 hr 932 F, air blast cooled, aged 48 hr 265 F.

TABLE 16 — DIDYMIUM CONTENT EFFECT ON INCIDENCE OF INCLUSIONS IN QE22A TEST BARS

| Didymium, % | % of Test bars with inclusions (Total of 472 test bars) |
|---|---|
| 1.0-1.49 | 0 |
| 1.5-1.99 | 9.5 |
| 2.0-2.49 | 18 |
| 2.5-3.00 | 25.5 |

Consideration must hence be given, if the casting is in the T4 or T6 tempers, to the overaging effects of the preheat on the properties of the casting. Figure 5 shows that QE22A-T6 rapidly overages at 500 F (260 C). This, and the fact that the weld is in the as-cast condition, normally requires that the casting be fully heat treated after welding. Welds should be re-examined after heat treatment to ensure that minor discontinuities have not been opened up by the quenching operation.

The corrosion resistance of QE22A is generally similar to that of other magnesium base casting alloys. An important consideration where high strength alloys are concerned is susceptibility to stress corrosion. Tests by R. J. M. Payne and N. Bailey[6] have indicated that neither the normally heat treated nor overaged alloy is susceptible to stress corrosion. QE22A may be chromated and anodized by the normal methods applied to rare earth containing magnesium alloys.

## DISCUSSION

The differences in the levels of properties obtainable from QE22A-T6 and alloys like AZ81A-T5, HK31A-T6 and EZ33A-T5 are such that the alloy must be placed along with ZK61A-T6 in the high yield strength category. The latter alloy is however likely to give appreciably lower yield strength in many casting designs due to susceptibility to mass effect.

The excellent tensile properties at elevated temperatures have led to considerable interest in QE22A-T6 for aircraft and missile castings. There are also projected uses for long service applications where the good fatigue strength and creep properties at temperatures up to 400 F (204 C) are of importance. It may be said that QE22A is the most universal of magnesium alloys combining high yield strength at room temperature with attractive elevated temperature properties. The cost of QE22A alloy is comparable with that of the thorium bearing alloys HK31A and HZ32A.

The present trend of specifying premium grade for aircraft and missile castings with mandatory minimum properties on test bars cut from the casting presents a challenge to the magnesium sand foundry industry. The task of the alloy metallurgist is to provide not only materials with enhanced mechanical properties, but also good castability so that gating development is not unduly protracted nor is excessive design simplification required before the necessary properties and quality standards are realized.

It has been an unfortunate corollary, and not confined to magnesium-base alloys, that some sacrifice in castability has been demanded when specifying high strength alloys. With QE22A the sacrifice has been minimized, and it is felt that the alloy merits its inclusion in alloy specifications and will be increasingly utilized for high strength, high quality, magnesium castings.

## ACKNOWLEDGMENT

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# ALUMINUM BRONZE HARD SPOTS

by Norman A. Birch


## ABSTRACT

**Practically everyone in any way connected with aluminum bronze castings (whether ingot maker, foundryman or machinist) has at one time or another encountered an expensive defect known simply as "hard spots." These are unalloyed metallic inclusions near to diamond hardness, and range in size from barely discernible (but damaging even to carbide cutting tools) to as large as ⅜-in. diameter.**

**While the nature, cause and cure of this defect is still a matter of speculation, the author reports detailed studies, including their microstructure and chemical composition and native habitat and is able to recommend inspection and shop procedures which virtually guarantee freedom from this particular defect.**


## INTRODUCTION

Practically everyone in any way connected with aluminum bronze castings, whether ingot maker, foundryman or machinist, has at one time or another encountered an expensive defect known simply as "hard spots." Because they are troublesome to everyone, just about everyone has contributed in some measure to the ten-year compilation which is presented here in the form of quiz answers to the original "Twelve Questions About Hard Spots" project. At least six ingot makers have run independent tests, several crucible makers have yielded information they would normally have classified as trade secrets, melters and shop people have been most patient and understanding about the many records imposed on them in the early days, and the author's company's research center at Mahwah, N.J., has employed everything from a hand magnet to microchemistry to x-ray diffraction to help resolve the problem.

As of today, exactly what hard spots are is not known, nor is it known how to destroy them once they are formed. However, fortunately how to avoid them in melting operations and how to screen incoming ingot to keep them out of shops has been learned. This left the problem with the ingot makers. They in turn have been equally aggressive in attaining good practices, so that hard spot ingot has not been seen at the author's company for almost 4 years.

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## QUESTIONS AND ANSWERS

***What are the hard spots?***

Hard spots are small silvery inclusions randomly located and intimately bonded in the metal. They are usually first noticed in machining because even a small hard spot is enough to chip the edge of a carbide cutting tool or break a tool steel hob. They can usually be picked out on a machined surface because there is a definite discontinuity in the finish where the tool was damaged by running into the hard spot. The visible hard spots range in size from small specks to as large as ⅜-in. in diameter. Usually, the ones which cause damage to a tool are approximately 1/32- or 1/64-in. in diameter.

Hard spots can be readily identified by scratching with the point of a knife over the spot; unless the spot has been gouged out, the feeling is as though a shattered glass particle were embedded in the metal.

They are not to be confused with unalloyed metallic iron inclusions which are rare, random and soft. True hard spots are invariably round and occur in epidemics. The larger inclusions show definite flake-like structure, and spall readily with silvery fracture if attempt is made to chip them out of the base metal. Spalled flakes and the particles themselves are strongly magnetic.

On a polished micro specimen, a cross-sectioned hard spot shows up as a roughly round silvery inclusion cutting across both alpha and beta phases, and may or may not be attacked by the normal etch. Nearby may be small particles which appear to have either sloughed off or not yet agglomerated to the main inclusion. Still smaller particles of similar appearing inclusions will be in the matrix, which are the normal iron-aluminum intermetallic compound (also magnetic) found in aluminum bronzes. Figure 1 shows the cross-section of a massive hard spot, while Fig. 2 shows the internal components of another smaller and possibly more characteristic hard spot.

Figures 3 and 4 show the details of yet another hard spot. To avoid being misleading, please note

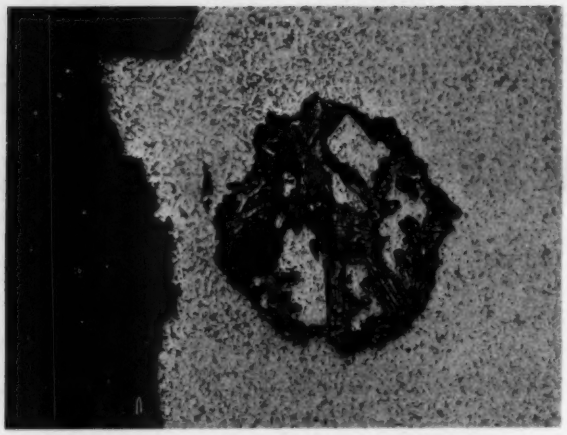


Fig. 1 — Massive hard spot. Notice essentially unaltered aluminum bronze structure within the inclusion although all hard spots do not show this. Concentrated nitric acid etch. 15 ×.

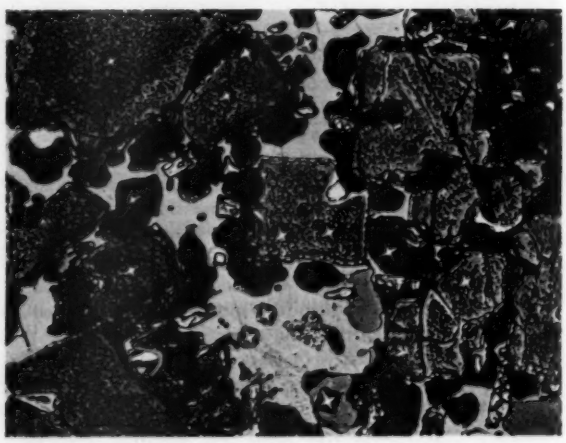


Fig. 2 — Hard spot structure. Predominating speckled phase is 2400-2700 Vickers hardness, magnetic and is assumed to be iron rich. Small diamond shaped indications are micro hardness impressions. Micro Vickers hardness of individual phases (and colors, as-etched) are 790 (yellow), 946 (white), 1400-1700 (pink and magnetic), 2400-2700 (brown and magnetic). Concentrated nitric acid etch. Magnetism determined under separate magnetic etch. 500 ×.

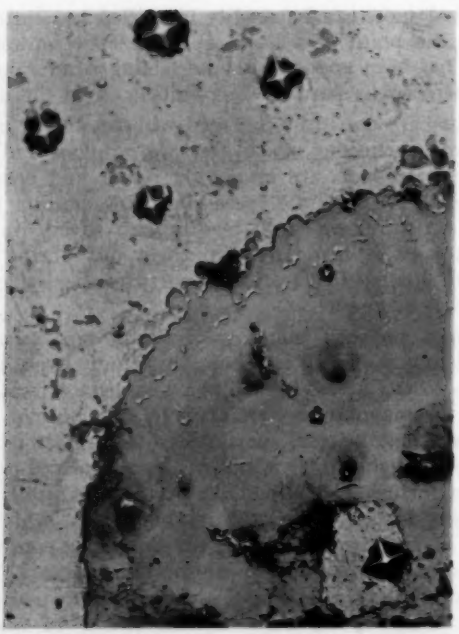


Fig. 3 — Unetched hard spot (gray). Note small indentor impressions (Vickers hardness 2700-3300) in hard spot compared to much softer aluminum bronze both surrounding and within the hard spot. 500 ×.

that Figs. 1 and 2 show "natural" hard spots found in defective castings, whereas the hard spots of Figs. 3 and 4 were characteristic of an "accidental" epidemic which developed during a laboratory investigation of the effects of boron as an alloying element to the aluminum bronzes.

*Have any hard spots been analyzed?*

Yes, the author's company has furnished natural epidemic hard spot samples to a number of interested sources and received the interesting answers shown in Table 1. It would appear that natural hard spots are not all the same, but are principally iron hardened by aluminum or small amounts of other elements.

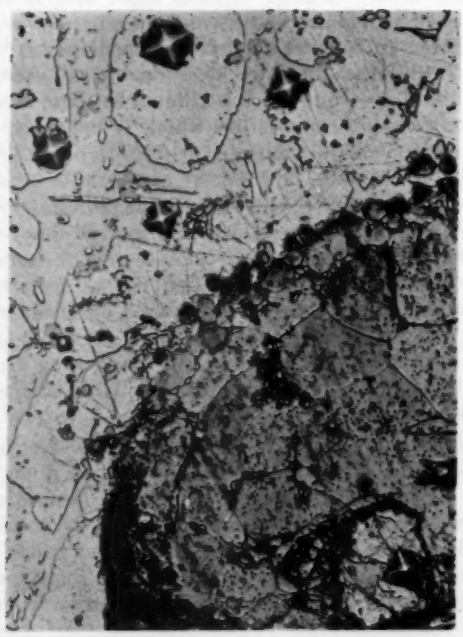


Fig. 4 — Same area as in Fig. 3, etched one sec in concentrated nitric acid to bring out aluminum bronze matrix phases of varying hardness. 500 ×.

TABLE 1 — ANALYSIS OF NATURAL EPIDEMIC HARD SPOTS

| Source | Method | Reported Composition |
|---|---|---|
| A | Microchemistry (sample of Fig. 2) | Quantitative: Iron 98.08 - 98.29%<br>Carbon 0.84 - 1.09%<br>Qualitative: Al, Ti, V positive<br>Mo, Sn negative |
| B | Spectrographic | Approx. 75% Fe, 25% Al, no silicon, very little evidence of other elements. |
| C | Spectrographic | Approx. 98.5% Fe, trace Cr |
| D | Petrographic | 70 - 80% Fe, balance Al, no silicon |
| E | Spectrographic | "An extremely high melting point alloy of iron, molybdenum, vanadium and titanium. Most remarkable is that unless the arc hit the hard spot, there are no indications whatever (less than 0.001%) of the presence of Mo, V or Ti." |

***Is the occurrence of hard spots in aluminum bronze generally known in the industry?***

Yes, but it is treated like epilepsy and is generally not discussed except in the family, or with a specialist. Fortunately, the metallurgical specialists to both the ingot and foundry families are generally agreed that the source of most hard spots is faulty furnace practice. Certain crucible glazes, fluxes and certain grades of raw material are believed to trigger off an epidemic, but most epidemics can be traced to careless compounding of the alloy. The author's company has never had trouble from good ingot properly melted, even when hardeners have been used to modify the composition. However, hard spot castings have come from bad ingot properly melted, and from good ingot melted with shop borings, especially if pure metallic aluminum has been added to adjust the composition.

***Do hard spots segregate at the top or bottom of a melt of metal—might we get rid of them by skimming or decanting?***

Everyone agrees hard spots do not segregate. Their density, calculated from an assumed analysis of about 75 per cent Fe-25 per cent Al would be about 0.27 lb/cu in., which is about the same as that of the aluminum bronze base metal. Hence, they will not float nor be concentrated by centrifuging. They have been observed both on the outside and the inside of machined centrifugal castings, and are located randomly when seen in a fractured ingot.

***Can hard spots be located by nondestructive tests?***

The author's company has been unable to locate hard spots by nondestructive tests. At one time on hand were approximately 1000 rough machined rings weighing 16 lb each. By close visual inspection all suspect castings were culled out, often the only guide was a change in surface finish, since the spot which caused the tool damage which changed the surface finish was too small to be seen or had been machined away. Partial finish machining (and random fracture) showed that less than 10 per cent of the remaining 819 castings were from hard spot melts. Unfortunately, no way was found to separate the good ones from the bad.

X-ray through a $1\frac{3}{16}$-in. metal section containing a visible hard spot approximately $\frac{1}{32}$-in. in diameter picked up the 0.030 in. jump in thickness caused by the tool hitting the spot but showed no trace of the spot, presumably because of similar density and intimate bonding.

The ultrasonic reflectoscopic inspection, which picks up discontinuities in structure and changes in modulus of elasticity, failed to give reliable evidence with either direct or angle searching units.

A magnetic analysis company reported the retentivity of the base metal as being too high to allow reliable location of the more magnetic but relatively small hard spots. Another company could offer no suggestions on locating other than exposed hard spots, most of which could be seen by eye.

***Has the analysis of the base metal anything to do with hard spots?***

Quite probably, but the effect of alloying elements is not certain. Knowing that hard spots are principally iron, it is believed there is statistically more chance of hard spot trouble when compounding the 5 per cent iron aluminum bronzes than when alloying the low iron (one per cent) grades. The author's company's main hard spot experience has been with Alloy 9C (A.S.T.M.-B148), since this is the general-purpose grade most commonly melted, but hard spot ingot from several sources in Alloys 9A (one per cent Fe) through 9D (4.5 per cent Fe) has been rejected. There appears to be no correlation with nickel in the range 0-5 per cent, and no hard spot trouble came from a single shop melt containing 2.0 per cent zinc or from others containing silicon up to 0.23 per cent.

Only the high manganese aluminum bronzes (nominal composition 75% Cu, 12% Mn, 8% Al, 3% Fe, 2% Ni) appear to be immune from hard spots. In common with aluminum bronzes these alloys are subject to a low-grade infection known as "sparkle" or "silvery needles" which is manifest in reduced ductility and a tendency for brittle fracture, on which the sparkle is readily visible to the trained eye. As will be discussed later, there is evidence that the sparkle is a hard brittle microconstituent similar to and possibly of the same generic origin as hard spots.

Add to this the knowledge that small amounts of manganese are beneficial to the strength-ductility ratio of aluminum bronzes and that small manganese additions have been helpful in destroying hard spots and it appears a reasonable hypothesis that sparkle can be treated as a mild case of the disease known as hard spots. No doubt it frequently passes unnoticed, except that test bar elongation values are a little lower than normal and tool regrind on machining is a little more frequent.

Manganese, therefore, appears to act as a possible vaccine against hard spots, and the combination of high manganese and low iron may explain why they have not been seen in the high manganese aluminum bronzes.

TABLE 2 — BORON TREATED ALUMINUM BRONZES

| Heat No. | Analysis, % | | | | | | | | |
|---|---|---|---|---|---|---|---|---|---|
| | Cu | Al | Fe | Ni | Zn | Sn | Mn | Si | B |
| 53-632 | | | | | | | | | |
| Tap 1 | 85.45 | 10.21 | 4.30 | <0.01 | <0.01 | <0.01 | <0.01 | 0.09 | 0.036 |
| +FeB | | | | | | | | | |
| Tap 2 | 85.81 | 10.27 | 3.92 | <0.01 | <0.01 | <0.01 | <0.01 | 0.09 | 0.015 |
| 53-633 | | | | | | | | | |
| Tap 1 | 88.32 | 10.26 | 1.30 | <0.01 | <0.01 | <0.01 | <0.01 | 0.08 | 0.033 |
| +FeB | | | | | | | | | |
| Tap 2 | 88.41 | 10.18 | 1.27 | <0.01 | <0.01 | <0.01 | <0.01 | 0.07 | 0.029 |

*How do hard spots form?*

Concensus among the metallurgical specialists who treat with this disease is that most natural epidemic hard spots are born during the exothermic reaction which occurs when aluminum is added to the bath. Normally, the nickel (if required) and iron have already been alloyed with the copper. The iron is frequently in the form of punchings which have been bundled, and the bundles are worked back and forth in hearth furnaces or are assimilated by the metal action of rotary furnaces. In local crucible furnaces the iron may have been added as nails or as core rods stirred in with a strap iron bar. In any event, circumstantial evidence indicates that the contagious moment is when the aluminum enters the bath.

If it finds unalloyed iron present, whether from the bundles, the nails or shreddings from an overly-used overheated stirring rod, the conversion of soft unalloyed metallic iron appears to take place at exothermic heat and the melt succumbs to a case of hard spots or sparkle or both. It appears that the single best protection against these diseases is good stirring and adequate temperature and time to assure absolute assimilation of the iron before the aluminum is added. Shop remelts of straight aluminum bronze need watching too, since any free iron from overheated stirring rods, iron in contaminated borings, broken tool tips in partially machined castings, cleaning room shot or grit trapped in the riser cavities of remelt metal, or even the nails frequently used in aluminum bronze gating for skimming purposes and coming back in shop revert can be dangerous sources of free iron. This is especially true if the shop practice is to add a little metallic aluminum to compensate for melt loss.

There has been among the specialists a small minority who believed that hard spots were formed in the presence of silicon, and that 0.10 per cent was the maximum which could normally be tolerated, with 0.05 per cent a reasonable level and with 0.03 per cent a safe level. The data of Table 1 (at which time the author's company was much concerned with silicon) plus incidence of hard spots in ingot with silicon below the presumably safe level, combined with no trouble at high silicon (0.23 per cent), has pretty much dissipated the worries about silicon. Low silicon content is preferred, but there is no concern about the normal spread of 0.01-0.11 per cent silicon found in commercial aluminum bronze ingot.

Other commonly suspected sources of hard spots in foundry melting include:

a. Worries about the use of silicon carbide as opposed to clay graphite crucibles.
b. The glazes used by the crucible makers, especially the borosilicates or fluoborosilicates.
c. The fluxes sometimes used as metal cover during melting, specifically the borides and fluorides.

While it is true that the worst case of hard spots came from a new silicon carbide crucible, with reportedly a silicate glaze on the inside, melting suspect ingot containing 0.12 per cent silicon, this same crucible later made perfectly beautiful metal when melting good ingot. No reason can be seen for taking advantage of the longer life of silicon carbide crucibles. Nor has trouble come from using new crucibles with presumably different glazes from several manufacturers.

**Fluxes.** The matter of fluxes is not quite so clear. As noted earlier, the laboratory developed an accidental epidemic of hard spots while investigating whether boron could be induced into the alloy merely by using a boric acid cover. The procedure was to melt 300 lb of metal under a boric acid flux, pour Tap 1 test castings; add 0.03 per cent boron as 18 per cent ferroboron to the remainder of the melt; pour Tap 2 test castings. The chemistries are reported in Table 2. Test bar properties are not reproduced because they varied randomly, depending on the number and location of hard spots within any given tensile specimen.

The type of hard spot encountered is the one shown in Figs. 3 and 4. The gray color (unetched) and the microhardness coincided exactly with a sample of 18 per cent ferroboron (unetched) mounted, polished and tested in a similar manner. It could be argued that these particular hard spots are merely remnants of unalloyed ferroboron, except that the melts obviously picked up boron from the flux alone, and hard spots and sparkle were found in all castings, both before and after the ferroboron addition. Other observations included:

a. That the condition was more severe in both melts after the innoculation than before, although the boron content of Tap 2 castings is lower in both cases. Presumably the boron concentrated into hard spots, and the drillings for chemical sample failed to pick up a proportionate share of hard spots.
b. That the condition was more severe with the high iron rather than with the low iron alloy.
c. That another heat made with 3.0 per cent manganese had a sugary fracture rather than the characteristic sparkle or hard spots of the melts of Table 2.

*Are visible hard spots the only ones that cause trouble?*

No. The hard spots big enough to be seen may cause tool breakage and can be cause for rejection of the casting, but the "invisible" ones can apparently cause excessive tool wear. This brings us to a more detailed consideration of the disease is termed silvery needles or sparkle, which it is believed could equally well be described as "invisible hard spots"

TABLE 3 — NICKEL ALUMINUM BRONZE INGOT, LOT 4555

| | Composition, % | | | | | | | | | Bhn, |
|---|---|---|---|---|---|---|---|---|---|---|
| | Cu | Sn | Pb | Fe | Ni | Zn | Al | Mn | Si | 300 kg |
| Ingot 1 | 81.09 | 0.16 | 0.05 | 3.78 | 4.09 | 0.23 | 10.58 | 0.03 | 0.02 | 217 |
| Ingot 2 | 81.08 | 0.12 | — | 4.16 | 4.08 | — | 10.45 | 0.07 | 0.03 | 212 |

Ingot 1 had fine fibrous fracture.
Ingot 2 had fine fibrous fracture containing sparkle.

Fractures of each ingot are representative of many fractured from this lot, and while vendor claimed all metal was from same melt, the differences in tramp elements and in fracture would indicate the lot to have been a mixture of two similar melts, one with sparkle trouble, the other without.

or "hard spot needles." These are shown in Figs. 5 and 6. Of a number of photographs and micros available, Fig. 5 was selected as representative of natural sparkle received in commercial ingot of the chemistry shown in Table 3, and Fig. 6 as typical of accidental sparkle developed in the boron treated melts of Table 2 and visible in the same fracture as the hard spot of Figs. 3 and 4. Micros of natural sparkle correlate closely with Fig. 6, and in all cases the needles appear to be different from and much larger than the normal Fe Al constituent found in well-dispersed small particles in all iron-containing aluminum bronzes.

Recognizing that the sparkle constituent is harder and larger than the normal microconstituents, and that it can influence machining, and that it is a melt characteristic (although less evident than hard spots), sparkle ingot is avoided as much as possible. Usually this is a choice of source.

### *Do the hard spots affect mechanical properties and casting serviceability?*

Yes, but not seriously. The costly nuisance of hard spots is tool wear and breakage on machined parts. Castings to be used unmachined, or with only one surface faced and a few holes drilled into a flange, are equally as strong and satisfactory with or without hard spots. Even when the spots appear on a machined surface, they are so few, so random and so intimately bonded within the base metal that they do not lessen serviceability of the casting. Justification for this view is based on many tensile test results where spots found in the fracture had not appreciably altered the expected test bar values unless they occurred at or close to the machined surface of the bar, in which case yield strengths were normal but the elongation and ultimate strength values were considerably reduced. Only on critical surfaces, such as the bore of a bushing where they might score the shaft, would hard spots be significant cause for rejection.

Sparkle can be more serious, as shown in Table 4. The comparison is especially interesting because the British and American data were developed independently, and both were the result of a distinct fracture characteristic noticed by ever-curious trained metallurgists (this kind of detective work can be endlessly fascinating). It has been called "silvery needles," the British called it "sparkle," which term has been



Fig. 5 — Sparkle fracture (silvery needles) of aluminum bronze ingot no. 2 (lot 4555). Microhardness of needles averaged 450 Vpn compared to matrix average of 235 Vpn. 50 ×.

elected to use as being simpler and equally descriptive. Note that the yield and tensile are not significantly affected, but that sparkle reduces the elongation and (according to the British) is even more detrimental to impact. Care should be used in extrapolating these data for purposes other than this comparison since these are averages, and both aim-



Fig. 6 — Hard spot needles in boron treated aluminum bronze. 500 ×.

TABLE 4 — SPARKLE EFFECT ON MECHANICAL PROPERTIES

| | Heat Treated 2% Nickel Aluminum Bronze (A.S.T.M. B148, Alloy 9C-HT) | | | As-Cast Alloy 40 | |
|---|---|---|---|---|---|
| | Required | Normal | Sparkle | Normal | Sparkle |
| Yield Strength, psi* | 40,000 min. | 48,000 | 49,000 | 46,500 | 47,000 |
| Tensile Strength, psi | 90,000 min. | 104,000 | 102,000 | 100,500 | 98,000 |
| Elongation, % in 2 in. | 6 min. | 16 | 10 | 34 | 29 |
| Izod Impact, ft-lb | — | — | — | 31 | 19 |
| Source: | American Brake Shoe | | | J. Stone Co. (England) | |

*Yield at 0.5% extension for American values; at 0.15% offset for British values.

chemistry and the degree of sparkle will affect the average properties.

The British, for example, report another comparison where elongation dropped from 32 per cent normal to 24 per cent with sparkle present. The author's company has, in turn, had rejections because sparkle cut the elongation below the 6 per cent passing figure. Sparkle can be more dangerous than hard spots, although it is a milder disease and less apparent, because it can steal away the safety factor while still permitting test bars to pass requirements. One consolation, however, is that direct comparison between sand cast keel blocks and test bars cut from a similar section thickness centrifugally cast in a metal mold indicates that rapid cooling tends to suppress sparkle.

### *Can hard spots be destroyed by remelting?*

Not by ordinary remelting, at least. If they are an iron-aluminum compound they would have a melting point above 2600 F, and aluminum bronze is normally melted to a maximum of 2200 F. Some of the specialists claim that remelting merely causes hard spots to grow, and that the only safe practice is to run the contaminated metal through a cupola, burning out all the iron and aluminum, to salvage the copper content only.

Figure 7 shows two lots of fractured hard spot ingot, one with sparkle condition in addition to the hard spots. To run a definitive test, from Lot 3448 28 ingots at random were broken (about 500 lb). All were found to have sparkle fracture, with 14 ingots (or 50 per cent) having visible hard spots. The metal was crucible melted without flux or additions to 2550 F, with sample ingots poured off at intermediate temperatures of about 2150 and 2250 F and the balance pigged at 2550 F. All ingots were fractured. The low temperature ones were excellent with no trace of hard spots, but the others varied from excellent, to some with sparkle and one with a beautiful $\frac{1}{8}$-in. diameter ball on a $\frac{3}{16}$-in. diameter shell of what appeared to be the same material. The condition was a sphere within a sphere, as though the outer one had grown around the inner one. The microchemistry reported in Table 1 is that of the outer sphere; the ball was the subject of Fig. 2.

In yet another remelting test 500 lb of 40 per cent visible hard spot ingot were melted to 2600 F, adding a small amount of aluminum to compensate for expected melt loss. All ingots were fractured and none showed any evidence of hard spots. Standard keel block test bars met the requirements and the metal was used, but the fracture of both the ingot and the test bars showed a trace of sparkle and one test bar had one small hard spot.

One vendor asked for minor chemical deviation in order to rework and resubmit five lots of rejected alloy 9C ingot. He remelted each lot separately, deliberately stirring in 0.5 per cent manganese. Three lots cleared up and were accepted, the other two still had hard spots. Melting to high temperatures with or without a manganese addition may be helpful, or may not be helpful.

### *Can we protect against hard spots by inspection of ingot?*

Yes, definitely. The author's company has not experienced a single case of unexplained hard spots since 1951. The closest to it was while melting sparkle ingot in the shell molding department. Hard spots were found in one or two melts, and for a few days there was concern that sparkle alone could degenerate into hard spots, something never experienced. Soon, however, traces of hard spots appeared in centrifugal castings and also in the mechanized department castings, each of which had their own melting furnaces and were using different ingot.



Fig. 7 — Fractured hard spot ingot (alloy 9C). Lot 3448 (left) shows sparkle fracture with one hard spot at upper right. Lot 3840 (right) shows four hard spots in a five fibrous fracture.



Fig. 8 — Fractured good ingot (alloy 9C). Both lots show uniform fine grained fracture free of any defects or inclusions.

Investigation showed that the only raw material common to all three departments was one batch of shop borings, and the company prefers to believe that circumstantial evidence convicted the shop borings of somehow being contaminated rather than to take this single instance as evidence that sparkle can actually agglomerate to produce hard spots, or that our ingot receiving inspection procedures are not fool-proof.

The ingot receiving inspection is simple, inexpensive and fast, if you have a suitable hydraulic press. Ingot is purchased to A.S.T.M. B30, which requires each lot to be uniform, to be of one melt only and to be so identified. Random samples are selected in the ratio of one ingot for each 1000 lb of metal. These are fractured under the press. Figures 7 through 11 show representative fractures of a number of ingot, with and without defects. Next, one-half of each ingot is spot ground on the bottom face to clean metal and is checked for hardness.

If the fractures are uniform in appearance and contain no unalloyed metallic inclusions, the hardnesses are uniform within 14 Bhn points spread, the hardnesses are within the range established as proper for the alloy, the vendors certification shows chemistry and properties within specification and the vendor is known from past experience to be reliable, then the ingot is accepted and used without further question. For a normal lot of 10,000 lb, the receiving clerk can clear the ingot for use in less than 30 min. If the fracture is nonuniform in appearance, if the hardness varies more than 14 points within the lot or is out of the expected range, it usually indicates improper alloying or inadequate mixing before ingotting and a metallurgist takes over. Figures 12 and 13 show representative fractures.

Note that fracture and hardness are not purchase requirements, nor is ingot rejected for any cause other than nonuniformity or out-of-specification. Fracture and hardness are merely quick and inexpensive inspection tools to insure good raw material for the casting quality requirements of these premium alloys.

This is not to say that all ingot not to the above tight inspection standards is rejected. As long as the



Fig. 9 — Fractured good ingot (alloy 9C). Lot 3997 (left, 170 Bhn) has coarse woody fracture. Lot 4003 (right, 179 Bhn) has fine woody fracture).



Fig. 10 — Fractured good ingot (alloy 9C). Lot 4330 (left) has coarse woody fracture. Lot 4331 (right) has columnar woody fracture.



Fig. 11 — Fractured hard spot ingot (alloy 9C). One ingot (lot 3441) showing one large hard spot.

ingot is uniform, and within a reasonable hardness spread of approximately 20 points, the metallurgist will accept and can blend uniform lots of high and low hardness to provide a proper average alloy for melting. He can even use an occasional nonuniform lot, provided the nonuniformity is not excessive and his required melt weight will use up the entire lot, permitting it to properly homogenize on remelting. However, he cannot use a mixed shipment of five 4,500 lb melts unidentified, one of which contained hard spots.

Nor can he use a 27,000 lb melt of manganese bronze which was clean enough but varied in copper content of individual ingot from 56 to 72 per cent (this was discovered under the press when some ingot broke like glass and others would only bend). For the foundry to try to select material for a 600 lb melt from this lot of ingot would have been sheer folly. All ingot sources have been most cooperative and helpful in eliminating this type of shipment, and many are now using the hardness check as a quality control on the ingot line.

### *Are hard spots encountered in other copper-base alloys?*

Yes, the high tensile yellow brasses (commonly called manganese bronzes) experience some of the same difficulties as the aluminum bronzes, since they also contain appreciable iron with aluminum contents up to 6.5 per cent. Some manganese bronzes will exhibit a bright sugary fracture, but never sparkle. There will be occasional unalloyed metallic inclusions, but these are usually a few flakes of unalloyed iron, an occasional ball of metallic-appearing slag (usually immediately under the top surface) or a small spalled piece of ingot mold, but never a true epidemic hard spots of the type described for aluminum bronzes. In fact, manganese bronze ingot has never been rejected for inclusions, recognizing that they will slag off or be assimilated on melting. However, ingot sources are kept advised, since they also are interested in promoting and maintaining a quality reputation for these most important alloys of the copper-base castings industry.

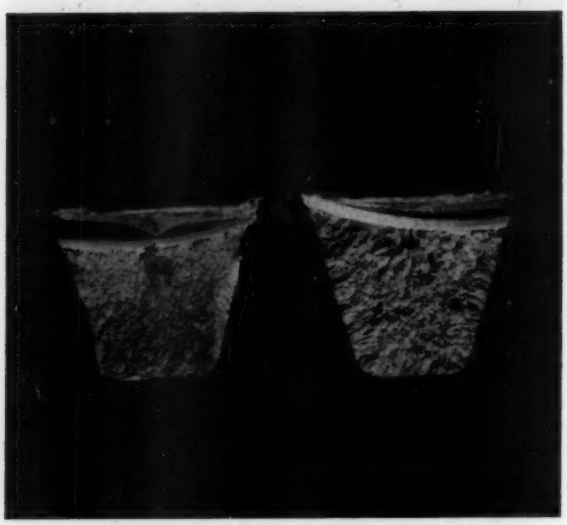


Fig. 12 — Fractured nonuniform ingot (alloy 9C). Lot 3999, fine grained, 143 Bhn (left) and coarse grained, 241 Bhn (right).



Fig. 13 — Fractured nonuniform ingot (alloy 9A). Lot 4721A, silvery crystalline, 302 Bhn (left) and tough bronzy, 115 Bhn (right).

# $CO_2$ PROCESS IMPROVEMENTS

by D. A. Taylor


## ABSTRACT

Consideration of the mechanism of bonding in the $CO_2$ process using silicates of different $SiO_2/Na_2O$ ratio indicates that with low ratio silicates, the molding mixture can tolerate considerable amounts of impurities which are usually regarded as harmful. Subsequent work showed that the presence of clay reduces strength after gassing by only a moderate amount, and where necessary this can be overcome by a small increase in the amount of silicate added.

Clay, or any material containing alumina, reduces the rate of conversion of quartz to crystobalite at high temperatures, and hence the rate of expansion, and reduces metal penetration and the risk of spalling. Burnt clay is as effective as raw clay and has the advantage that it does not make ramming difficult, and mixtures of undried silica sand and burnt sand from a green sand system can be used to replace the dry silica sand in general use.

Clay also affects strength on cooling from high temperatures, and reduces the difficulty in removing cores from castings of moderate size. Mixtures of new silica sand and sand recovered from silicate bonded molds by crushing only are also being successfully used. For reducing the spalling and growth of silicate bonded ladle linings, the use of 4 per cent alumina is more satisfactory than the addition of clay.

The gassing operation can be improved by using a mixture of equal amounts of air and carbon dioxide.


## INTRODUCTION

The $CO_2$ process has two advantages which have been the chief cause of its success:

1. The mold or core can be hardened in contact with the pattern, or core box, and an accurate reproduction is obtained.
2. Hardening is rapid at room temperatures and drying stoves are unnecessary. Transport to and from the stoves of molds which require a crane for lifting is frequently a source of trouble, particularly in the jobbing foundry, and a reduction in the number of times molds have to be moved is of great advantage.

The process as used at the present time has these disadvantages:

1. It is difficult to remove cores from some castings.
2. Metal penetration resulting in poor surface finish and sometimes spalling of the mold surface occurs, particularly with large molds.
3. Large quantities of dry silica sand are required and similar quantities of used sand have to be discarded. Recovery of the used sand is possible only by the use of a difficult cleaning process.

The object of this paper is to show that by proper selection of sand and silicate, core removal can be made somewhat easier, surface finish can be improved, the amount of new silica sand required can be reduced and in some cases undried can be substituted for dried silica sand. Gassing can be improved by using mixtures of carbon dioxide and air in place of carbon dioxide alone.

## NATURE OF THE SILICATE BOND

Soda and silica can be combined in any proportions, and the resulting compounds are miscible with water. The range of possible silicates is shown in Fig. 1.

The silicates commercially available have a silica/soda ratio between 3.3 : 1 and 1.6 : 1, with a water content between 40 and 60 per cent, and are covered by the shaded area in the graph.

When the silica/soda ratio exceeds 3 : 1, the silicate is unstable and is rapidly decomposed by carbon dioxide with the precipitation of silica, as seen in Fig. 2. When these silicates are used in the $CO_2$ process, gassing is extremely rapid, but the precipitated silica has poor bonding properties and gassed molds are weak and friable.

When the silica/soda ratio is reduced to 2 : 1 the effect of carbon dioxide is to form sodium carbonate and a silicate of higher ratio. At the same time water is removed from the silicate and viscosity increases rapidly with a resulting increase in strength, as shown in Fig. 3. The gassing period needed is longer and, if

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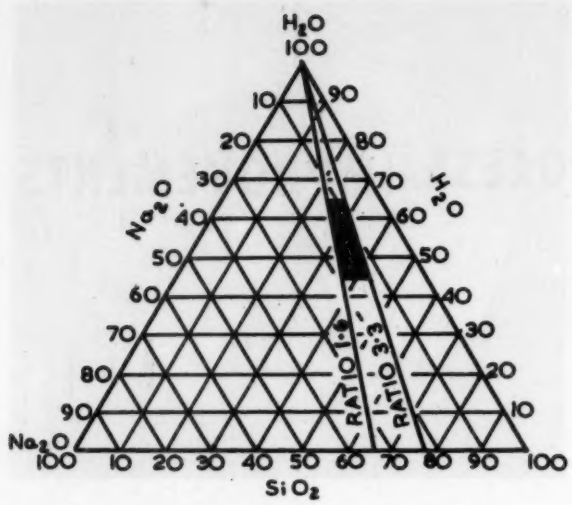



Fig. 1 — Composition triangle showing field of sodium silicates available for foundry use.

this is interrupted and a mold containing silicate dried, there is a further increase in strength, as shown in Fig. 4, since sodium silicate itself has a powerful bonding action. If, on the other hand, gassing is continued the silicate is completely decomposed and the mold becomes weak and friable.

When the silica/soda ratio is still further reduced, gassing time is vastly increased and silicates of this kind are rarely used with pure silica sand.

In practice, three types of bond may be present:

1. Precipitated silica. This is relatively weak and with normal additions of silicate, strength rarely exceeds 100 psi.
2. "Setting" following alteration of silicate ratio and increase in viscosity. This gives strengths of up to 300 psi.
3. Drying of unaltered silicate. This gives strengths exceeding 1000 psi.

In general, the lower the silicate ratio, the longer is the time required for gassing, but the higher the strength of the gassed mold.

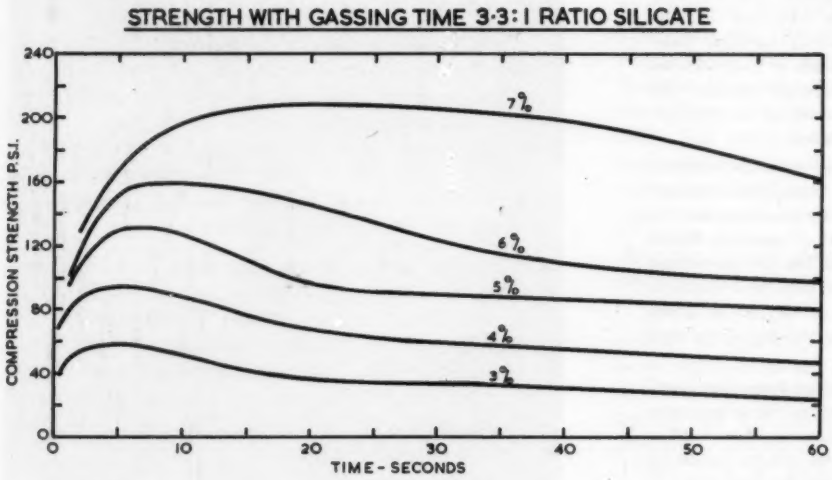



Fig. 2 — Strength with gassing time for 3.3 : 1 ratio silicates.

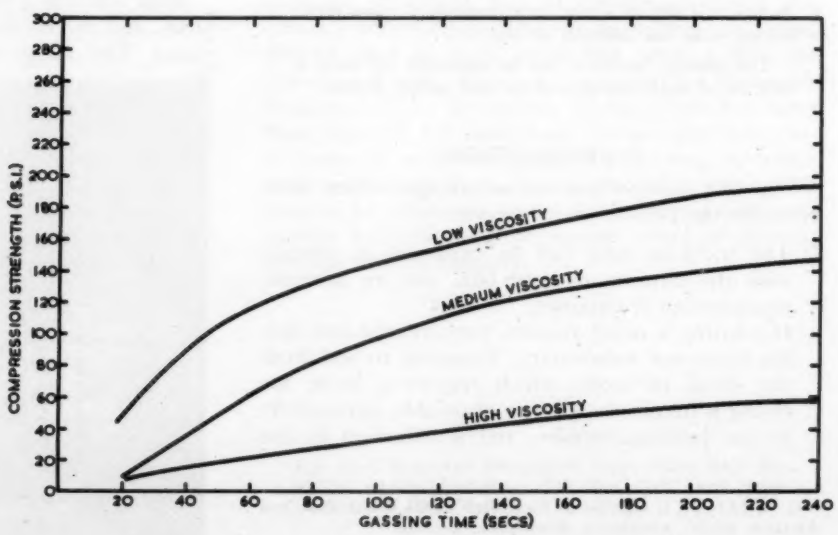



Fig. 3 — Strength with gassing time for 2 : 1 ratio silicate. Silicate addition — 4 per cent.

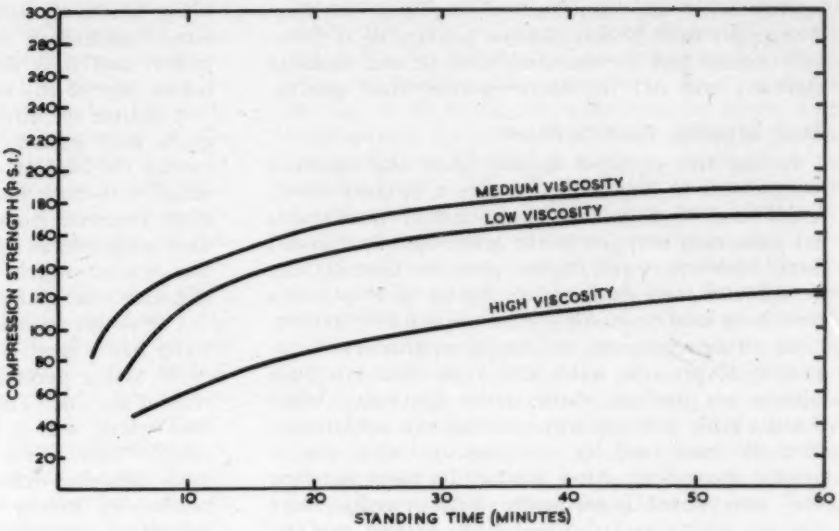



Fig. 4 — Strength with standing time for 2 : 1 ratio silicates. Silicate addition — 4 per cent. Gassing time — 30 sec.

### *Water Content Effect*

The rate of increase in strength on gassing is greater with silicates of lower viscosity, i.e., higher water content. The long gassing time required with the silicates of low ratio can therefore be offset by increasing the water content.

### *Clay Effect*

When the base sand contains a small amount of clay the gassed strength is reduced, but the reduction is small and can readily be counteracted by a small increase in the amount of silicate. In consequence, attempts have been made to use sands containing up to 2 per cent of clay and fines in place of purer sands, which were more costly and difficult to obtain. There is no technical difficulty in this, and the cost of any additional silicate required was more than offset by the inexpensiveness and convenience of a local sand deposit. However, the dust produced in drying sand containing up to 2 per cent of clay and fines caused so much nuisance that the practice was abandoned.

If the clay content is raised appreciably two changes occur:

1. Water is transferred from the silicate to the clay and the mixture appears dry. Gassed strengths are then low.
2. Any exchangeable bases in the clay may react with the silicate and change the silica/soda ratio.

There does not seem to be any difficulty in taking care of these effects of clay; the first can be overcome by adding water or, better, by using undried sand, and the second by using the silicate with a ratio of 1.6 : 1.

In order to examine the possible fields of use of clay-bearing sands, mixtures of Erith silica sand and Bromsgrove clay-bonded sand were made in various proportions and bonded with 4 per cent of 1.6 : 1 ratio silicate. Erith silica sand is a clay-free sand mainly on the 60 and 100 mesh sieves; Bromsgrove sand has a wide spread of grain size below 60 mesh and contains 12 per cent clay. Standard test pieces were prepared and gassed for one min with gas passing at the rate of 10 litres/min.

Test pieces were tested immediately after gassing and also after standing on the bench for 24 hr. The test results are shown in the table.

PROPERTIES OF MIXTURES OF NEW SANDS

| Sand Mixtures | | | | Green Properties | | |
|---|---|---|---|---|---|---|
| Erith, % | Bromsgrove, % | Gassed Strength | Strength after 24 hr | Moisture, % | Permeability No. | Green Strength lb/sq in. |
| 100 | — | <30 | 278 | 1.7 | 268 | 0.9 |
| 80 | 20 | 106 | 455 | 2.1 | 180 | 2.4 |
| 60 | 40 | 133 | 175 | 2.4 | 116 | 5.6 |
| 40 | 60 | 92 | 103 | 2.8 | 70 | 8.6 |
| 20 | 80 | 90 | 100 | 3.8 | 45 | 10.7 |
| — | 100 | 75 | 70 | 4.5 | 34 | 13.0 |

These tests illustrated the field of use and limitations of clay-bearing sands and low ratio silicates.

In the absence of clay the normal gassing procedure had little effect, and although strength developed on standing this was due to drying. If this test piece had been dried in a stove or allowed to stand until dry, its strength would have exceeded 1000 psi.

As clay-bonded sand was added in increasing quantity, gassed strength rose to a maximum with 40 per cent clay-bonded sand, and then fell, but even with clay-bonded sand alone the gassed test pieces were not unduly friable.

Strength after standing for 24 hr reached a high maximum with 20 per cent clay-bonded sand, and then fell. Whenever the proportion of clay-bonded sand exceeded 50 per cent strength changed only slightly on standing after gassing.

Sodium silicate increases the toughness of a clay-bonded sand, and all mixtures containing more than 60 per cent of the naturally bonded sand had a shatter index of 100. These would be too tough to ram

readily with any conventional molding machine. The upper limit of clay content permissible is therefore determined by the toughness of the molding mixture, and not by the properties after gassing.

### *High Moisture Content Effect*

In the test reported in the table, the moisture content was so adjusted as to give a mixture which could be molded with the least difficulty, since molds are sometimes stripped in the green state and gassed later. However, much higher moisture contents can be tolerated, and sands containing up to 10 per cent have been used to produce sound molds after gassing.

In an extreme case, a foundry system sand containing 10 per cent water and 5 per cent low ratio silicate was used to make molds for heavy brake drums. With this moisture content, the sand resembled the loam used for sweeping up rather than a normal green sand. After gassing the mold was firm and was poured immediately. Unexpectedly, there was no undue turbulence during pouring and no unusual evolution of gas, and the castings produced were perfectly sound.

When, however, sands with high moisture content are gassed and allowed to stand, a white deposit forms on the surface which becomes extremely friable. On prolonged standing the mold surface may fall away. This is illustrated in Fig. 5, which shows test pieces from a natural sand containing 8 per cent clay and fines, 10 per cent water and 5 per cent silicate. *A* is a test piece shortly after gassing, and *B* after standing 24 hr.

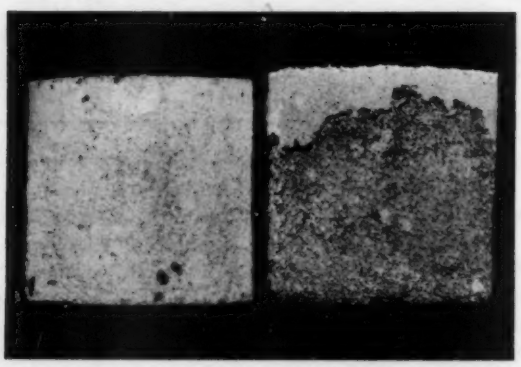


**Fig. 5 — Effect of standing on silicate bonded sands of high moisture content. A (left) and B (right).**

On this account it is recommended that the moisture content of the sand before silicate addition should not in general exceed 3 per cent.

## THE $CO_2$ PROCESS IN THE GREEN SAND FOUNDRY

It has been shown that sands containing clay can be used in the $CO_2$ process, and the way is therefore open to make use of the discard from the green sand system. This consists of clay-bonded sand which has lost some of its bonding power as the result of being heated by molten metal. Some of this sand is almost unchanged and some has lost all its bonding power, and it is necessary to establish a means of taking care of any variations between these extremes.

A sample of system sand returning to a mill in a green sand foundry was taken as the type of sand having the highest amount of active clay, and sand which fell from hot castings as having the lowest. Sand removed from any part of the system would have properties in between these two, and if a method was devised which could be used successfully with either, it would take care of any sand from the system.

The system sand was synthetic, and on milling with water had a green strength of 10 psi, a shatter index of 80 and a permeability of 90. It contained 4 per cent of sea coal. The dry sand was mixed with 3 per cent water, and 5 per cent low ratio silicate was added. This raised the shatter index to 100 and made ramming difficult. After gassing the molds were satisfactory, having a gassed strength of 130 psi rising to 180 psi on standing for 24 hr.

It would not have been possible to ram molds satisfactorily by normal methods, and few foundries have the equipment for dealing with it successfully. Accordingly, the dry system sand was mixed with an equal amount of undried silica sand to give a moisture of 2.5 - 3.0 per cent, 5 per cent silicate was added, giving a shatter index of 60, which could be molded without difficulty, and the gassed strength obtained was 200 psi.

The sand which had been strongly heated was satisfactory for use either with or without the addition of silica sand. It would be possible, by selection of the discard, to use the process without new sand, but it is safer to adopt the method which is successful with any sand from the system.

Undried silica sand as delivered to the foundry usually contains less than 6 per cent water, but occasionally when quarried in wet weather or delivered straight from a washer, the water content may be higher. If, however, the sand is allowed to stand in a bunker, or slightly heaped up, drainage will soon reduce the water to below 6 per cent.

The use of a mixture of used clay-bonded sand with undried silica sand and low ratio silicate is particularly advantageous in the small or medium sized foundry in which the bulk of the production is in green sand molds, the $CO_2$ process being used for some cores and difficult molds. Not only does it enable sand which is unsuitable for use in the green sand system to do another job before it is dumped, but it does away with the need for drying quantities of silica sand which is dusty as well as costly.

### *Gassing Procedure*

The hardening of silicate molds is due to a chemical reaction which, for a silicate of given composition, proceeds at a rate depending only on temperature, provided that $CO_2$ is present. The ideal method of gassing would be to supply gas to all parts of the mold or core at the same time, and at a rate sufficient to replace that used in the reaction. If gas is supplied at a faster rate than this it fulfills no useful purpose and simply escapes to atmosphere.

Applying gas in the ideal way is impracticable. Temperature changes considerably during the day, but it is unlikely that either maintaining a constant sand temperature or matching the gassing rate to the temperature would be worth while. The only exception to this is when the temperature of the sand falls low, as may happen after a shut-down in winter, when occasionally the sand is so cold that the reaction does not take place (a small percentage of really warm sand in the mix takes care of this difficulty).

It is also impracticable to bring the gas to all parts of the mold at once, and in foundries using the $CO_2$ process exclusively it may be worth while to develop special techniques for gassing, but the most widely used method is to introduce the gas at a few points. If the gas is introduced rapidly much of it escapes, whereas if the rate is greatly reduced hardening is uneven, being complete at the points of entry before any great change has taken place elsewhere. By diluting the gas with an equal volume of air, more uniform gassing can be obtained without excessive wastage. Suitable proportioning valves are required.

Displacement of the air in the mold pores by gas is easier and proceeds over a larger part of the mold as permeability rises, and on this account it is desirable that the permeability of the sand should exceed 40, but most molding sands when mixed with an equal quantity of silica sand have a permeability well in excess of this figure. However, sands containing sand from a green sand system usually have a lower permeability than silica sands, and dilution of the gas is desirable. Moreover, the reaction proceeds more slowly with low ratio silicates, and the two factors of gassing rate and permeability combine to make the use of an air/$CO_2$ mixture desirable if excessive wastage is to be avoided.

## SILICATE-BONDED SANDS DURING HEATING AND SUBSEQUENT COOLING

### *Properties at High Temperatures*

When silicate-bonded clay-free sands are heated, strength rises as a result of complete drying of any undecomposed silicate, and reaches a maximum at about 400 C (752 F). Strength then begins to fall and is negligible at temperatures exceeding 700 C (1292 F) when the bond is almost completely liquid, as shown in Fig. 6. If heating is continued at really high temperatures the silica sand itself changes. Quartz, the usual form of silica, is stable at temperatures up to 870 C (1598F), when it changes to tridymite, which in turn changes to crystobalite above 1400 C (2552 F). These changes are, however, extremely slow and rarely occur to any great extent in molds, although small amounts of tridymite have been found in the sand adhering to the surface of large steel castings.

It would be expected that the presence of a fusible silicate would increase the rate of change and reduce the temperature at which it occurs, but the extent of the influence was unexpected. If a silicate-bonded silica sand is heated to a temperature exceeding 1000 C (1832 F) for half an hour or so it changes completely to crystobalite, the tridymite stage being almost completely missed.

Quartz has a specific gravity of 2.65 and crystobalite of 2.32, and the change is therefore accompanied by an expansion. The effect of heating standard test pieces for 20 min at various temperatures can be seen in Fig. 7. What happens is not clear. The fall in specific gravity should lead to an increase in volume of a little over 12 per cent, but the test piece on the right of Fig. 7 has increased in volume by more than 30 per cent. Crystobalite can be found on the surface of molds which have been used for the production of castings of quite moderate size and there are indications that surface finish may be affected.

### *Metal Penetration and Surface Roughness*

Metal penetration and surface roughness are frequently more pronounced on silicate-bonded sands than on clay-bonded sands of the same grain size. It is possible to slow up the change to crystobalite by the addition of a number of substances, particularly those containing alumina. Clays are effective to a considerable extent, but alumina alone has an even more pronounced effect, and almost completely inhibits the change.

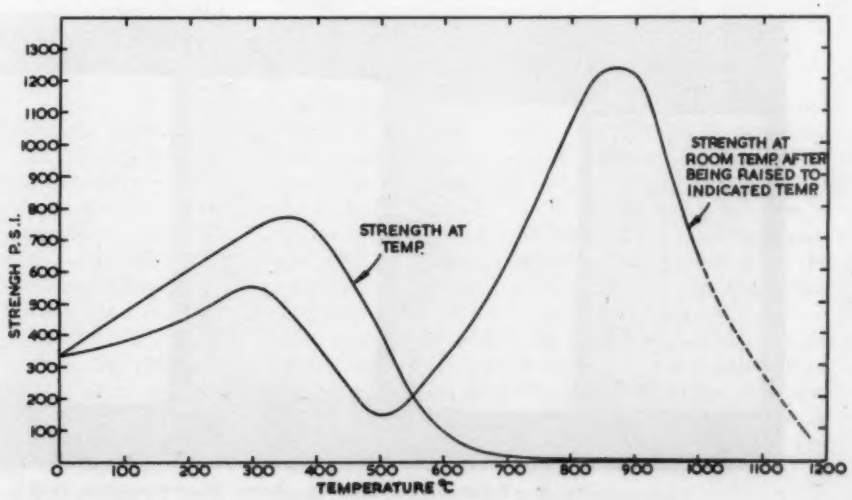



Fig. 6 — Strength of $CO_2$ process material at and after cooling from high temperatures. Four per cent silicate.

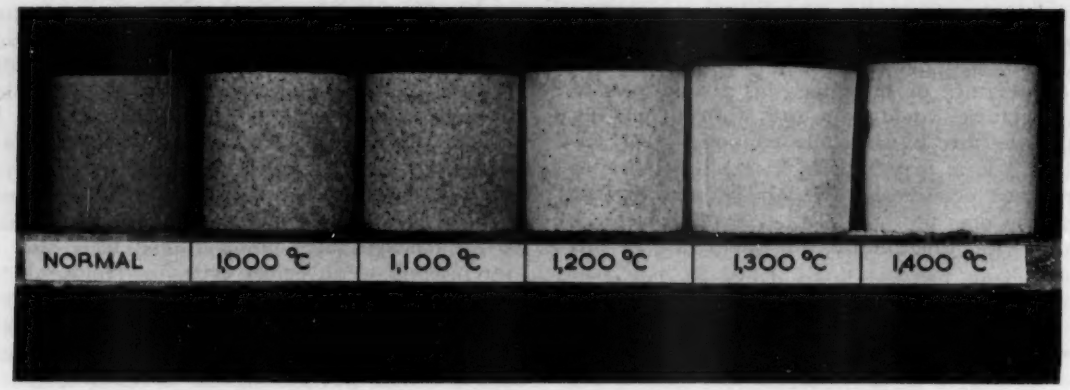



Fig. 7 — Expansion of silicate bonded sand at temperatures above 1000 C (1832 F).

It appears reasonable to suppose that metal in close contact with sand shown at the right of Fig. 7 would have a rougher surface finish than when in contact with that on the left, but it is not so obvious that an expansion of the surface layer could lead to larger pore spaces unless the dimensions of the mold increase. If the mold remains constant, it appears that an increase in the volume of the grains could only be obtained by a reduction in the size of the pores, and an improvement in surface finish. Nevertheless, surface finish is improved by an alumina addition and this technique is being used successfully in some foundries.

Clay itself has a similar effect and on this account an improved surface finish is obtained when using clay-bearing sands with low ratio silicates, even when permeabilities are similar. Since the system sand usually contains a certain amount of fines other than clay, there is a still further improvement in surface finish, and it may be possible to dispense with alcohol-based coatings for deep castings.

The effect of certain additions on the expansion is shown in Fig. 8. Alumina has a much more pronounced effect, which persists even when the sand is heated at 1400 C (2552 F) for 30 min. Test pieces containing 0, 1, 2, 3 and 4 per cent alumina after this treatment are shown in Fig. 9.

## LINING LADLES

An interesting use of the $CO_2$ process is in the lining of ladles. Silicate-bonded sand is easy to apply and is readily hardened, and provided that prolonged heating in an inverted position is avoided, a good firm lining is obtained which remains cleaner than a normal brick or sand-lined ladle during use. However, the high expansion may result in either spalling in the early stages or growth subsequently, and when such a lining was used for quite a small crucible it projected an inch beyond the crucible edge after short operation. The addition of 4 per cent of alumina eliminated both spalling and growth and the lining was satisfactory. There may be an upper limit of ladle size beyond which it is not safe to use silicate-bonded linings, but this limit is not known.

## STRENGTH AFTER COOLING — CORE BREAKDOWN

When a silicate-bonded core is heated its strength both at temperature and on subsequent cooling increases as a result of drying of any unchanged silicate present. When the heating temperature is raised

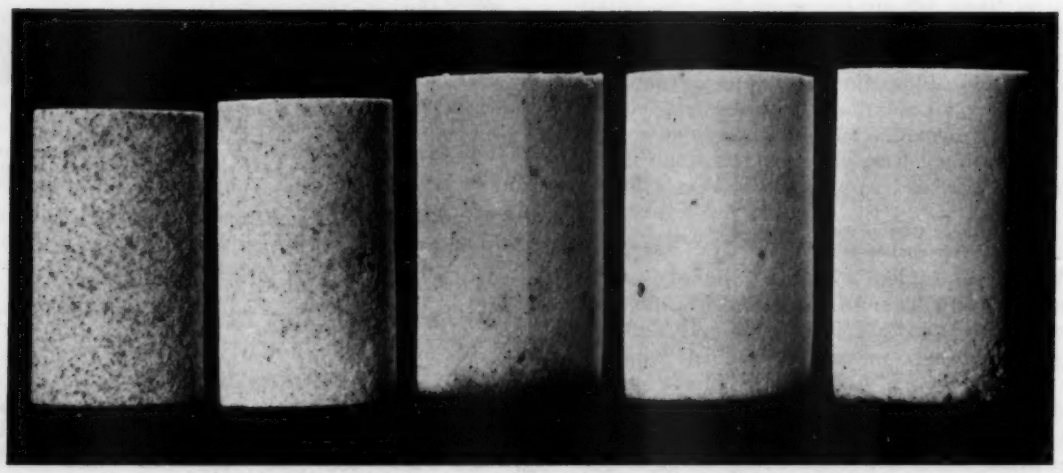


Fig. 8 — Effect of additions on silicate bonded sand at 1400 C (2552 F).

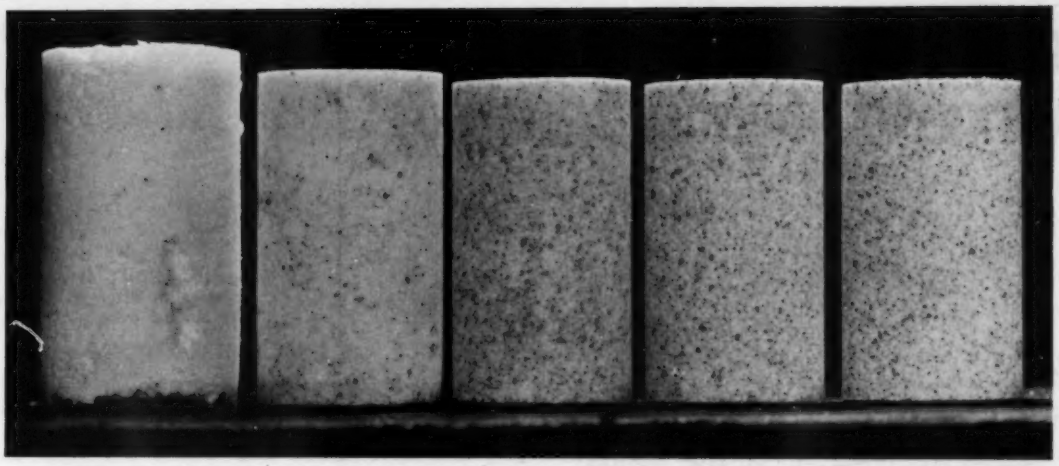


Fig. 9 — Effect of alumina addition on expansion of silicate bonded sand at 1400 C (2552 F).

to beyond 300 C (572 F) strength after cooling falls to a minimum, which is reached when the heating temperature is about 500 C (932 F). As the heating temperature is raised beyond this, strength after cooling increases rapidly, as a result of fusing and later resolidification of the bond. This high strength increases the difficulty of removal of silicate-bonded cores.

Various organic materials are added to assist breakdown. These may be put into two classes:

1. Materials including coal dust, wood flour and peat, reduce the strength of the mold or core after gassing, as well as after heating. If the silicate is increased to give the same gassed strength as before the addition, the strength after heating returns almost to its original value. Coal dust may have a beneficial effect on penetration and surface finish, but from the point of view of obtaining better breakdown it would be simpler to reduce the silicate.
2. Materials which readily develop a bond when they lose a small amount of water. When these are included in the mixture the amount of silicate can be reduced and the gassed strength maintained. At high temperatures the organic binder is destroyed and strength on cooling reduced. These can be effective under some conditions. Sugars are widely used for this purpose.

Clays can have an important effect on breakdown because they completely change the pattern of the curve, showing the relation between temperature of heating and strength on subsequent cooling. The maximum at 300 C (572 F) is suppressed entirely, the minimum at 500 C (932 F) is reduced and takes place at a higher temperature, as does the subsequent increase. Using a normal 2 : 1 silicate, clay reduces gassed strength, but if the amount of silicate is raised in order to bring the gassed strength to the original figure, the changed pattern of strength after cooling persists. This is shown in Fig. 10.

In the absence of clay, cold strength after heating rose to a maximum after heating to 300 C (572 F), fell to a minimum after heating to 500 C (932 F) and then rose rapidly. The addition of 2 per cent zepiolite, a clay chosen because it has negligible dry strength, reduces gassed strength, but this is restored by raising the silicate addition to 6 per cent. The strength after cooling is below that of the clay-free sand over the whole range of heating temperatures except at temperatures above 1000 C (1832 F), when cooled strength begins to fall as a result of quartz crystobalite change.

The effect of using the waste foundry sand with a low ratio silicate is even more interesting. This is shown in Fig. 11, in which the upper curve shows the strength after cooling of a silica sand bonded with a normal silicate, and the lower curve that of waste foundry sand bonded with a low ratio silicate. By using the latter, residual strength is greatly reduced over the whole range and core removal made less difficult.

When, however, the casting is such that the core reaches a temperature of 800 C (1472 F), difficulty in removal is to be expected. With such castings, not only does the core reach a high temperature, but it remains at that temperature for a considerable time, during which fritting of grains surrounded by molten silicates continues, with an increase in residual strength.

## NEED FOR CONTROL

Castings have been made successfully using sand taken at random from a green sand system and with moisture contents as high as 10 per cent. This may give the impression that any sand from either the foundry floor or dump can be used, but this is not so. The sand must be free from foreign material, and the clay content must be such that after the addition of silicate the sand is not too tough for easy ramming.

In most cases, it will be found that a mixture of equal parts of dry used green sand and undried silica sand has a moisture content below 3 per cent and can be rammed without difficulty. It is advisable to make frequent tests for moisture content, and 3 per cent should not be exceeded if the molds have to stand for some time before pouring.

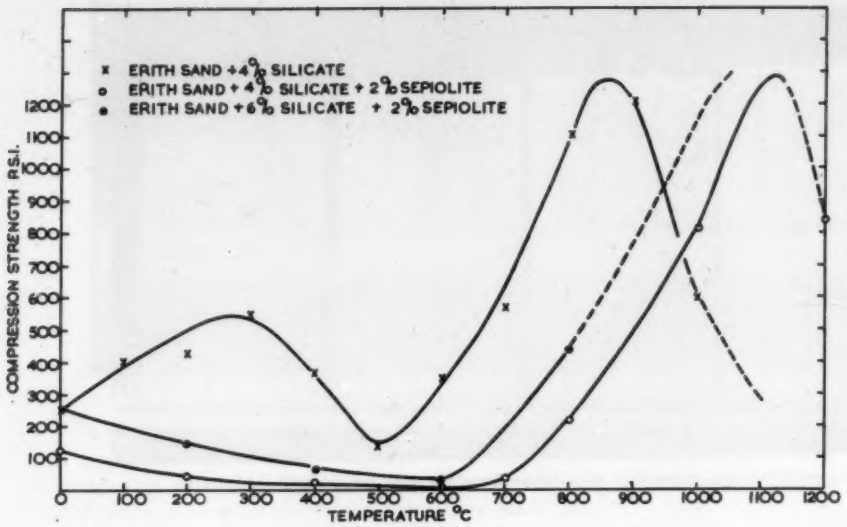



Fig. 10 — Effect of certain clays on strength after cooling of silicate bonded sand.

Satisfactory molds have been made using a silicate with 1.6 : 1 ratio to bond a mixture of equal parts of recovered sand previously bonded with silicate and a new silica sand. The moisture content of the mixed sand was one per cent, and this could be obtained by using the dry recovered sand with new sand of which only half had been dried. It was therefore necessary to dry only one quarter of the sand in the mold. Only normal gassing time was required and the problem of re-using sand was partly solved.

## CONCLUSION

1. Sand containing appreciable quantities of clay and water can be used in the $CO_2$ process provided that the sodium silicate has a silica/soda ratio of about 1.6 : 1.
2. The addition of sodium silicate to a clay-bonded sand increases toughness, and the upper limit of clay content must be such that ramming of the mold or core is not difficult.
3. Sand from a green sand system which has been heated until most of the bond has been destroyed is particularly suitable.
4. Molds with a high moisture content deteriorate rapidly after gassing and the final moisture content should not exceed 3 per cent.
5. The limits of toughness and moisture content are satisfied if the sand consists of equal parts of dry used sand from a green sand system and undried silica sand.
6. Clay reduces strength after cooling from high temperature and makes core removal less difficult.
7. Low ratio silicate can be used to rebond sands containing appreciable quantities of sand recovered after use in the $CO_2$ process, provided that moisture content is raised.
8. The efficiency of the gassing procedure is increased if a mixture of equal parts of air and $CO_2$ are used instead of $CO_2$ alone. The advantage to be obtained from the mixed gases increases as the permeability of the mold falls.

## ACKNOWLEDGMENT

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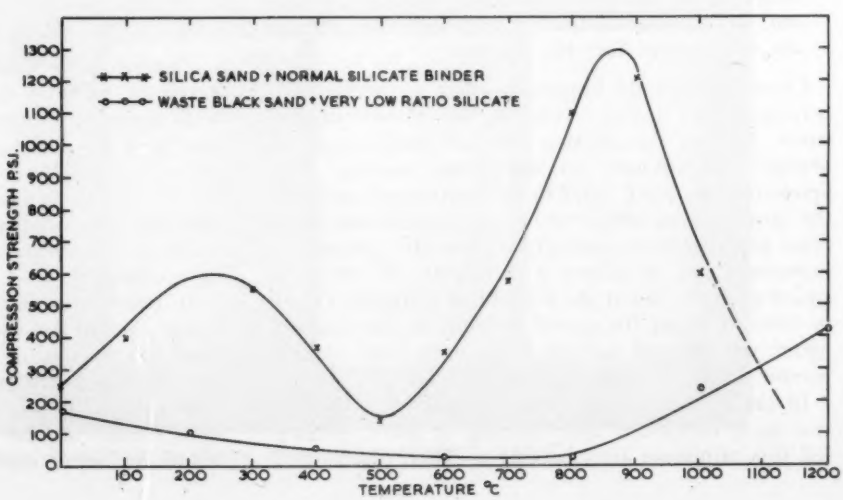



Fig. 11 — Strength after cooling from high temperatures for silica sand bonded with normal silicate, and waste black sand bonded with low ratio silicate.

# DUST AND VAPOR CONTROL FOR SHELL MOLDING OPERATIONS

by Elijah B. Henby


## ABSTRACT

**Shell molding operations emit dusts and vapors as hazardous as conventional foundry molding practices. A description is given of the hazardous compounds, their origin and general toxic effects. Preparation equipment and its exhaust (usually provided by the manufacturer) is described and illustrated. Methods are shown for determining exhaust requirements to reduce fire or explosion hazards. Suggestions concerning exhaust of the molding, curing, glueing and coremaking operations are given. Importance of makeup air and a method for estimating fuel costs are described. Recommendations are made regarding type of dust collectors best suited for shell mold exhaust systems.**


## INTRODUCTION

The thermosetting resins in shell mixtures have been used for molding purposes for over 40 years. Early shell mixture preparation included the use of dust suppressants, and the odors associated with the process afforded some protection to operators because they could not tolerate severe concentrations. It is no wonder then that the process was generally believed to have no potential hazards. Recent studies[1] by industrial hygienists prove that shell molding is as hazardous as conventional molding practice.

## POTENTIAL HAZARDS

An American Foundrymen's Society publication[2] and the studies cited describe the potential hazards, and establish excellent methods and suggestions for controlling the dusts and vapors associated with shell molding. This paper describes:

a. The compounds which are hazardous and their origin.
b. Those operations for which equipment manufacturers usually provide adequate means of exhaust.
c. Reasons for following in-plant ventilation practices established by AFS or as modified by recent experience.
d. Recommended dust collection equipment and disposal of contaminants.

The hazardous (toxic or flammable) compounds involved are given in the table.

The hazards of fire and explosion usually receive immediate attention from insurance underwriters, equipment manufacturers and foundrymen. Figures 1 and 2 illustrate pilot size shell preparation units typical of those widely used for preparation of shell mixtures by many methods, i.e., using powdered, lump or liquid resins and cold, warm or hot preparation. Note that the mixing and material transfer equipment is hooded, enclosed and/or exhausted or ventilated.

Exhaust rates for this preparation equipment are 2.5 to 5 cfm/lb of sand capacity. This exhaust volume prevents escape of dust and vapor into the working area, and keeps the atmosphere inside the mixer less than 25 per cent of the LEL (Lower Explosive Limit) for methanol vapor-air mixtures. These exhaust rates have been thoroughly checked with explosive meter equipment, and with proper equipment operation the methanol-air ratio has never reached the 25 per cent of LEL level. These rates are verified by the dilution formula[3] and by using values for methanol:

$$Q_{LEL} = \frac{387 \times 100 \times C}{\text{Mol Wgt liquid} \times \text{LEL} \times B}$$

where:

$Q_{LEL}$ = Cu ft dilution air required per lb methanol.
C = A safety factor dependent on percentage of LEL.
Mol Wgt = 32 for methanol.
LEL = 6 per cent.
B = A constant accounting for a decreasing LEL at elevated temperatures. B is 1

**Toxic or Flammable Compounds**

| Item | Operation of Origin | Remarks |
|---|---|---|
| Silica | Material handling<br>Preparation<br>Shakeout | Silica content, particle size and concentration is similar to conventional sand practice — risk of silicosis is no different. |
| Resins | Material handling<br>Preparation<br>Molding | Powdered resins present a fire or explosive hazard similar to flour. Some persons subject to dermatitis from contact. |
| Solvents | Material handling<br>Preparation<br>Mold release<br>Glueing | Methanol vapors are toxic, flammable and explosive. Little information is available regarding hydrolyzable silicones, but methyl, mixed methyl and phenylpolysiloxane types are recommended. |
| Products of Combustion | Preparation<br>Molding<br>Curing<br>Pouring<br>Shakeout | Reaction gases and vapors resulting from heating resins and solvents. May include formaldehyde (membrane irritant and hardener), phenol, ammonia (asphyxiant), carbon monoxide and dioxide. |

E. B. HENBY is with National Dust Collector Corp., Chicago.

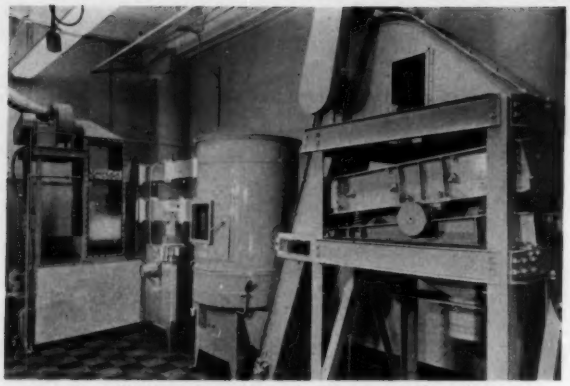


Fig. 1 — Shell preparation unit, heater, muller, elevator and cooler, with all exhausts connected to wet dust collector.

up to 250 F (121 C); B is 0.7 above 250 F (121 C).

Substituting:

$$Q_{LEL} = \frac{387 \times 100 \times 4}{32 \times 6 \times 0.7} = 1{,}150 \text{ cu ft dilution air required per lb methanol evaporated.}$$

NOTE: Derivation of the figures 387 and 100 which appear in the numerator:

The molecular weight in lb of any gas or vapor occupies 387 cu ft at 70 F and 29.9 in. mercury (Avogardro's Principle). Since the LEL is expressed in per cent (parts per 100) rather than ppm (parts per million as for the MAC), the factor of 1,000,000 becomes 100.

A 1000 lb sand batch in which 3 lb methanol are added and evaporated in 3 min:

Air required for safety:

$$3 \text{ lb methanol} \times \frac{1{,}150 \text{ cu ft}}{\text{lb}} \text{ or } 3{,}450 \text{ cu ft}$$

Air actually supplied:

$$3 \text{ min} \times 1000 \text{ lb sand} \times \frac{2.5 \text{ cu ft}}{\text{lb sand}} \text{ or } 7{,}500 \text{ cu ft}$$

Extensive tests have shown that the dilution to prevent fire or explosion in even an open type mixer is at a satisfactory level; however, the toxicity of methanol makes this type of operation unsafe. Toxicity is measured by maximum acceptable concentration (MAC) in parts per million (ppm) of air. MAC's are based on 8 hr exposure, are reviewed yearly and revised when toxicological data warrants.[3] Use of the appropriate formula illustrates why an open mixer can meet LEL dilution, but is unsatisfactory for toxic reasons:

$$Q_{MAC} = \frac{387 \times 1{,}000{,}000 \times K}{\text{Mol Wgt liquid} \times MAC}$$

where:

$Q_{MAC}$ = Cu ft dilution air per lb methanol.

K = An experience factor dependent on toxicity, uniformity of distribution, dilution, location of exhaust, etc., varies from 3 to 10.

Substituting (using the lowest K factor):

$$Q_{MAC} = \frac{387 \times 1{,}000{,}000 \times 3}{32 \times 200} = 182{,}000 \text{ cu ft}$$

dilution air required per lb methanol evaporated.

Assume as before that 3 lb methanol are evaporated, but the cycle is now at room temperature and takes 10 min:

$$\frac{3 \text{ lb methanol}}{10 \text{ min}} \times \frac{182{,}000 \text{ cu ft}}{\text{lb}} = 54{,}500 \text{ cfm}$$

minimum dilution ventilation required for open mixer to meet MAC – quite an exhaust volume from the corner of a shop having an open mixer.

Features available on some types of preparation units are—an expanded metal hoodtop covered with asbestos paper (in event of an explosion this hoodtop would minimize danger to personnel by serving as a "blow-out" panel) and an exhauster arrangement insuring negative pressure within the mixer hood (a safeguard against toxic vapor leakage into immediate working area).

It is obvious that exhaust of preparation equipment, as provided by most equipment manufacturers, is fully adequate as regards to safety, and is certainly more economical in air handled than attempting dilution ventilation in a working area.

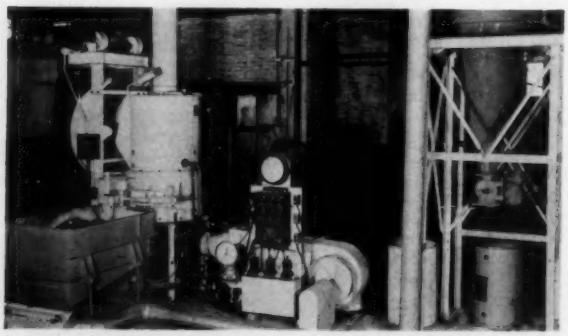


Fig. 2 — Shell preparation unit, heater, muller, screen, air conveyor and cyclone stripper, connected to stacks.

## MIXTURE SCREENING AND TRANSFER

Shell mixtures usually require screening and transfer to molding stations. The exhausts recommended by AFS[2] are satisfactory. Equipment manufacturers may provide hoods for exhausting these points of transfer and use such exhaust both for cooling and mixture. As a transfer medium a cyclone collector is properly used to recover the product because it is usable.

Molding, curing, glueing, core-making and shakeouts are hot; therefore decomposition products are evolved, e.g., phenol, formaldehyde, ammonia, methanol and carbon monoxide or dioxide. All are toxic. These operations require hooding in accordance with AFS[2] hood design practice, preferably enclosing or side draft canopy type hoods, the latter especially for operations requiring operator manipulation.[1,3] Volumes exhausted are calculated on basis of 150 cfm/sq ft of hood opening.

Makeup air is essential for proper operation of in-plant exhaust systems and for worker comfort. This

air should move through the working area and thence through area of possible contamination and on into exhaust hoods to sweep contaminants away from workers. Practically, makeup air is desirable and economical. Its use prevents:[3]

a. Upsetting the design of exhaust systems. Insufficient makeup creates higher statics thus lowering exhaust volumes below design values.
b. Back drafts from natural draft units, e.g., furnaces, boilers, ovens, etc., which can contaminate work areas with sulfur dioxide, carbon monoxide, etc., by short circuiting toward exhaust hoods.
c. Cross drafts which interfere with hood exhausts and can dislodge hazardous dusts from building structures.
d. Worker discomfort and the all too common difficulties with hard to open and slamming doors.

Makeup air is far more economical than attempting to maintain working temperatures by unplanned heating of infiltrated air. The cost of heating makeup air can be estimated as shown in the equation (the cost is based on air entering through a tempering unit):

$$C = \frac{A \times B}{D} \times E$$

where:

C = Estimated cost per hour.
A = Btu/hr to heat 1000 cfm outside air.
B = cfm exhausted in thousands.
D = Available Btus/unit of fuel used.
E = Cost of fuel per unit.

Example:

A = Chicago's average Dec.-Feb. temperature is 25 F (−3 C); therefore 47,000 Btu/hr per 1000 cfm are required.[3]
B = Assume exhaust system is designed for 10,000 cfm.
D = Gas is available at 1000 Btu/cu ft; at 75 per cent combustion heating efficiency, we have 750 Btu/cu ft actually available for heating.
E = Cost of gas is $0.10/therm (100,000 Btu) or $0.001/1000 Btu.

Substituting:

$$C = \frac{47{,}000 \times 10}{750} \times \$0.001$$

= $0.63/hr is cost of fuel to heat 10,000 cfm makeup air in Chicago area.

## GOOD HOUSEKEEPING

Although the importance of exhaust for control of toxic dusts and vapors has been stressed, housekeeping is important for prevention of fire and/or explosion. Some dusts will escape from various operations, become airborne and come to rest on floors, beams, etc.; therefore all areas must be kept free of heavy dust accumulations.

What should be done with the dusts and vapors removed from the working area? These contaminants should be removed from exhausts by dust and fume collection equipment to avoid air pollution. Much of the contaminant load consists of toxic vapors. Dry bag collectors are suitable for collection of the dusts, but are ineffective for removal of vapors. Fire and explosion hazards are present, and operational difficulties may occur with condensation of vapors from a shell system.

Wet scrubbers, not packed towers which would become quickly plugged with dusts, of the type familiar to foundrymen (orifice, impingement, centrifugal, etc., and having 3 to 6 in. static loss) are best for collection of the mixture of dusts and vapors. The vapors are absorbed by water, provided the scrubber gives good gas-liquid contact. Fire and explosion hazards are minimized by the quenching action in a scrubber. They are excellent as aftercleaners, following a cyclone "stripper" mentioned earlier. Scrubbing liquor can be recycled to avoid continuous contamination of a disposal system.

It is usually necessary that large shell installations investigate disposal methods for scrubbing liquors containing phenols, methanol, etc. Phenolics can be destroyed by oxidation with chlorine, chlorine dioxide, ozone or permanganate or absorbed on activated carbon or bituminous coal. However, this waste treatment is beyond the scope of this paper.

As a rule the emission of phenol or formaldehyde from a scrubber is not disagreeable, but the odors of these compounds are so strong that the minutest trace may be objectionable to some people outside a plant even though the concentration is not toxic. It may be necessary to introduce a masking agent (in aerosol form) into the exhaust stack. This is an economical remedy.

## CONCLUSION

In-plant control of dusts and vapors originating from shell molding operations is necessary because of their toxic and flammable nature. Preparation equipment is available with integral hood and exhaust systems to keep equipment interiors well below LEL's. The standards and principles for in-plant exhaust established by AFS are adequate, and should be followed in planning exhaust from molding, curing, glueing and pouring operations. Makeup air is economical. For the Chicago area, during average winter weather, the cost of gas fuel is approximately $0.63/hr for heating 10,000 cfm of outside air to room temperature. Wet collectors are recommended for collection of the dusts and vapors involved and, in case of odors from shell molding noticed outside the plant, a masking agent is available which can be introduced into the exhaust stack.

## ACKNOWLEDGMENT

The author is grateful to National Engineering Co., Chicago, for permission to use Fig. 1; and to the Beardsley & Piper Div., Pettibone-Mulliken Corp., Chicago, for permission to use Fig. 2.

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# TILT POURING PERMANENT MOLD ALUMINUM CASTINGS

by G. W. Stahl


## ABSTRACT

**An investigation to determine relative merits of the tilt pouring technique, as opposed to simple vertical gating of permanent mold castings, was initiated. Some of the factors considered are soundness of castings, appearance, production rates and relative cost of castings and molds. The study indicates definite advantages in quality of castings as well as in costs. Conclusions point up the fact that the process may have more advantages in some shapes of parts than in others.**


## INTRODUCTION

All foundrymen are aware of the tremendous increase in the use of aluminum alloy castings in almost all fields in recent years. Much of this increase is in castings which must be sound to withstand air or hydraulic pressures, or must have high mechanical strength or must eventually be polished to a high finish. The permanent mold process is of course well adapted to producing castings of this type. However one of the problems facing all permanent mold foundrymen is turbulence, or the introduction of air into the metal during pouring operation.

There has been a great amount of work done on designing of gating to help eliminate this problem. Most permanent mold foundries have developed their own pet variations of these developments and many are quite effective. Unfortunately, this gating usually involves extra labor in producing mold, extra metal in gates and much extra time in experimental work on each new job.

The purpose of this investigation is to determine if comparatively simple, foolproof gating could be used to produce sound castings by the tilt pouring technique. To this end then it was necessary to use a casting and conventional gating system, which definitely created turbulence and then apply tilt pouring and evaluate results. Although soundness, or the elimination of turbulence, was the primary objective, it was found that other advantages were also apparent.

## EQUIPMENT AND METHODS

The part selected was a planer base casting 18 x 23 in. in size having a depth of 2½-in. This casting, without gating, is shown in Fig. 1. The mold for this part was produced by machining from a cast iron casting made in green sand. Gating was machined from the solid. Vertical channels were cut ½-in. deep and ¾-in. wide. Metal was introduced into the cavity through ribbon gates 0.080 in. thick for the full length of the part. The riser at top was made ½-in. thick and 1¼-in. wide with a connecting ribbon 0.120 in. thick. Pouring basins were added to two vertical channels at top of mold to be used in tilt pouring.

The casting with gating attached is shown in Fig. 2. Total weight of mold complete with ejection and 18 x 24 in. mount plates was approximately 425 lb.

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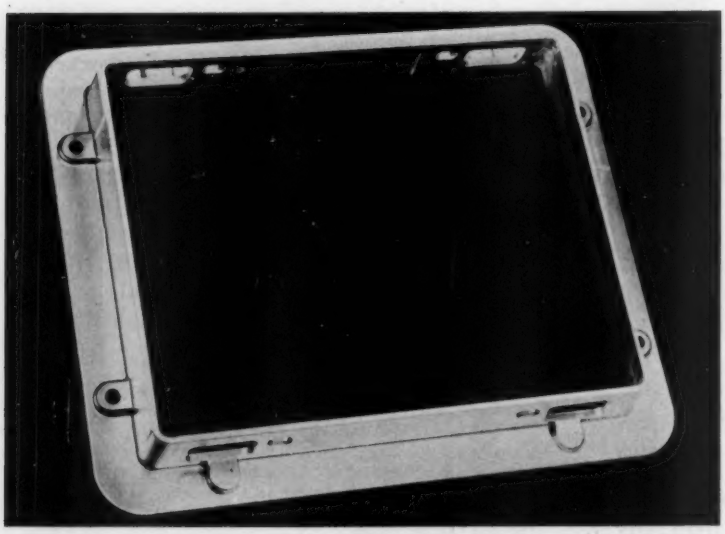


**Fig. 1 — Planer base casting 18 x 23 in. with a depth of 2½-in. and a weight of 7 lb.**

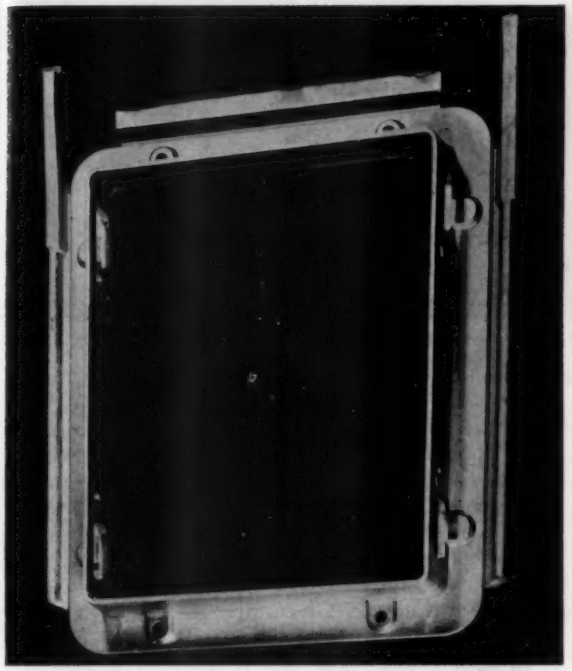


Fig. 2 — Planer base casting complete with simple vertical gating.

The mold was mounted in a standard permanent mold operating machine built to accept standard mount plates. This machine, with mold mounted in place, is shown in Fig. 3, while Fig. 4 shows machine in position for filling pouring wells for tilt pouring. The machine is equipped with a 4 in. diameter hydraulic cylinder for clamping mold and a 2 in. diameter cylinder for tilting. This unit is actuated with a hydraulic unit which furnishes hydraulic power, controls time cycle and controls speed.

The power unit is adapted for automatic operation of each cylinder independent of the other. With the power unit developing 1000 psi, the mold operating machine develops a clamping pressure of 12,000 lb. With the timer and speed control on the power unit, it was possible to control rate of tilt and also time of opening of mold and ejection of castings.

### *Heating the Mold*

The mold was heated to 350 F (176 C) and coated in conventional manner, using a vermiculite base coating on all surfaces with a graphite overspray on cores. Mold was reheated to 800 F (427 C) before mounting in machine. Metal used was commercial A-108 melted in a double chamber reverberatory furnace and held at 1400 F (760 C) in dip out well. Twelve castings were poured by alternately using straight and tilt pouring in order to equalize heat of mold and to allow operators experience in pouring, as well as to determine proper speed for automatic tilt pouring.

After producing these 12 castings mold temperature had leveled off at 725 F (385 C), and it was determined that a pouring time of 25 sec was quite satisfactory for tilt pouring. The test run then was made by producing 25 acceptable castings by each method using same metal of the same temperature in the same mold and with the same cycle.

## RESULTS

### *Cost of Castings*

The first apparent difference noticed was that in tilt pouring only one operator was required, since it was only necessary to fill pouring wells for each cycle. These wells could be filled by pouring rapidly. However, with vertical pouring it was necessary to use two operators, since the vertical channels were 20 in. apart, and it was necessary to pour quite slowly. It would have been possible to design gating system so that one man could pour both sides, but this would have complicated the mold considerably.

In order to get 25 acceptable castings by tilt pouring it was necessary to pour only 26 pieces. This would indicate a scrap rate of 4 per cent. With vertical pouring using two men, it was necessary to pour 32 pieces. The scrap rate then from manual pouring was 22.9 per cent. These scrap pieces were due to misruns because of variation of rate of hand pouring. Since overall cycle was 2 min 30 sec for both

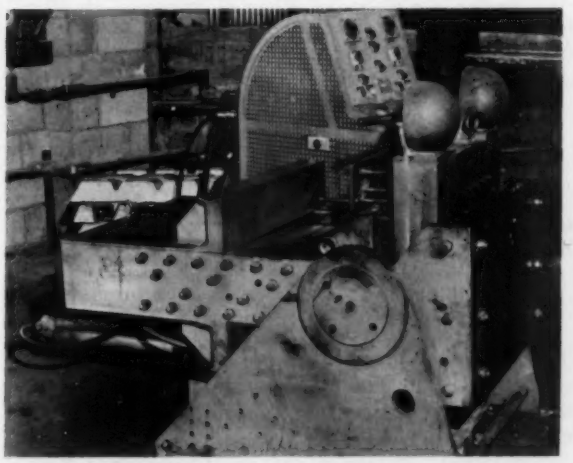


Fig. 3 — Mold for planer base mounted in operating machine with power unit ready for operation.



Fig. 4 — Same machine and mold as in Fig. 3 with machine in position for filling pouring wells.

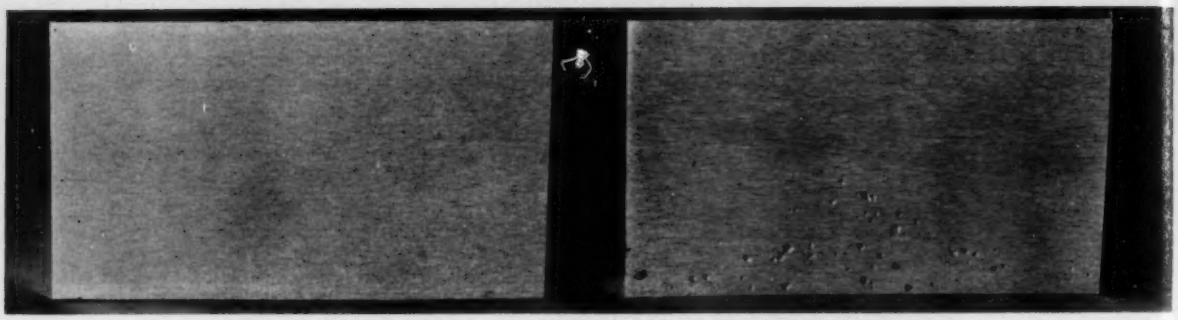


Fig. 5 — *left* — section showing lack of voids in same area in tilt poured casting; *right* — section showing typical voids in one area of vertical poured casting. Both sections were ground to expose structure 0.035 in. below surface.

methods, one man produced 25 castings in 1.08 hr, whereas two men required 1.33 hr, or 2.66 man hr, to produce same number of castings.

In order to check castings for soundness, a number of castings from each group were sectioned and ground on 180 grit belt to expose any voids. On the right in Fig. 5 is a section taken from a typical casting poured vertically. Note the large number of voids. Some of these measured as much as 0.080 in. diameter. On the left in Fig. 5 is a section taken from same area in typical casting made by tilt pouring. This section shows no voids.

Obviously, the vertical poured casting did not show this amount of voids in all areas. The area selected was on a horizontal plane, which normally would show effects of turbulence. However, the section from tilt poured casting was from same area, in fact the tilt poured castings showed absence of voids throughout entire castings.

### *Flash*

Examination of castings showed one other distinct advantage on the side of tilt pouring. This advantage was in the better control of flash on cored holes. On the right in Fig. 6 is a section taken from a typical vertical gated casting, while illustration on left in Fig. 6 is an identical section taken from typical tilt poured casting. Note that flash on vertical poured casting projected into cored hole, in fact this flash measured more than $\frac{5}{16}$-in. in width in some cases. Flash on tilt poured castings measured less than 0.050 in. Please bear in mind these castings were produced in same mold under same conditions except for method of pouring.

It was readily apparent that tilt poured castings were quite uniform in color and finish, whereas the vertical poured castings showed a tendency toward variation in brightness, and even showed tendency toward shrinkage in certain areas. This variation on vertical poured parts was apparently due to changes in hand pouring rate. The tilt poured castings were all poured automatically, which allowed cavity to be filled uniformly each time.

Although there was less difference between the castings in finish than in some other factors, the tilt poured castings showed a better overall surface finish than did the vertical poured castings. In fact the vertical poured castings showed definite roughness in areas of flow and in areas which had voids below the surface.

## CONCLUSIONS

It is apparent that on this part, tilt pouring will give turbulence free castings by using the simplest of gating. Also these better, turbulence free castings can be produced faster and at less cost than by vertical pouring with the same gating.

This study was made on only one particular design of casting. The advantages of tilt pouring may be greater or less on other shapes.

The author realizes that the concept of tilting a permanent mold to pour is not new. However, it is felt that not enough study has been devoted to the technique to realize its full potential. The author believes that unless automatic control is used on the pouring much of the advantage may be lost. Also a study of shapes of pouring basins is indicated, since it is possible to vary the rate of pour from bottom to top of mold by varying shape of basin.

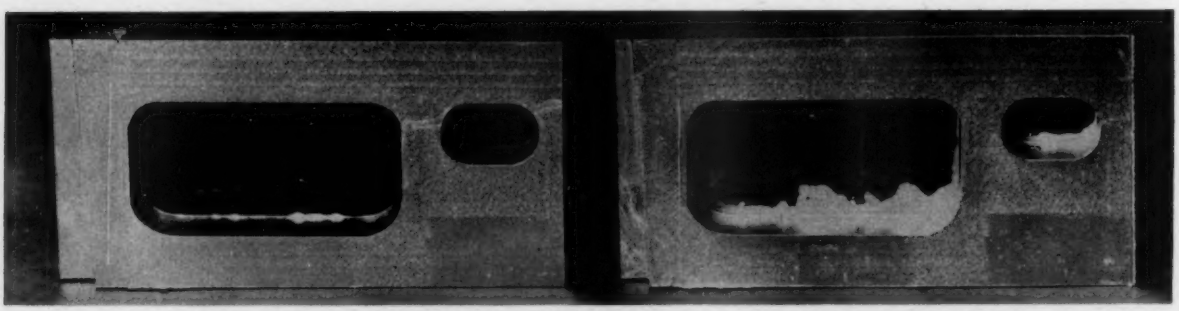


Fig. 6 — *left* — section showing amount of flash in cored holes in tilt poured casting; *right* — section showing amount of flash in cored holes in vertical poured casting.

# ALUMINUM PATTERN CASTINGS

by W. E. Sicha


## ABSTRACT

**Aluminum patterns and core boxes have the advantages of light weight, good machinability, high conductivity of heat and generally satisfactory resistance to abrasion. Problems have been encountered occasionally in obtaining aluminum pattern castings of satisfactory metallurgical quality. The quality requirements vary for different types of pattern equipment. Foundry practices for attainment of the desired quality are discussed in considering each pattern type, shown in the illustrations.**


## MATCHPLATES

Practically all of the aluminum matchplates, or cope and drag plates, with integrally cast patterns, now are pressure cast in plaster molds. This casting process, also used to produce core boxes, has been described in several technical papers[1,2,3,4,5] and depicted in movies.[6,7]

An important feature of the process is the economic advantage gained in making plates incorporating several identical patterns. Relatively inexpensive transfer techniques can be used to produce many patterns from a single master pattern.

Application of pressure in casting affords accurate reproduction of fine details in the pattern. Use of plaster molds provides smooth pattern surfaces and permits maintenance of close dimensional tolerances. It is essential that the master pattern be produced with a slightly higher level of smoothness and dimensional accuracy than is required in the cast pattern plate. This can be accomplished most effectively by using a completely machined metal master pattern.

Since the desired pattern shape is provided by the plaster casting process, little machining is performed on pattern plates. The major metallurgical requirement, therefore, is that the metal be sound on the surface and to a depth of about $\frac{1}{16}$-in. A moderate amount of internal gas porosity or shrinkage will not impair the serviceability of the matchplate.

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Melts of aluminum alloys for casting matchplates do not require thorough fluxing for removal of hydrogen, if the foundry practices produce sound metal at and near the surface of the pattern. Alloys with good casting characteristics, which include the aluminum-silicon and aluminum-silicon-copper alloys, are equally suitable for pressure cast matchplates and core boxes. Heat treatment is not necessary except for pattern equipment that will be operated at an elevated temperature. This subject will be discussed later.

## PATTERNS AND CORE BOXES

Aluminum patterns, other than matchplates, generally are produced by sand casting. A few patterns are fabricated using wrought aluminum products, and material selection can be based on readily available mechanical property data.

Practically all master patterns and models require extensive machining, as do many sand cast patterns and core boxes. Most of the quality problems with cast pattern equipment are encountered when machining operations are necessary. They include porosity, warpage and inadequate surface smoothness.

### *Porosity*

The three principal types of porosity are shrinkage, cope gas holes and gas porosity. The presence of shrinkage is evidence that satisfactory gating and feeding was not provided in producing the casting. This type of defect should not be encountered often, because foundrymen recognize the necessity to compensate for the reduction in volume during solidification of aluminum alloy castings. Many technical articles have been published on gating and feeding, providing ample information on techniques for producing shrinkage-free castings.[8]

The occurrence of localized gas holes at cope surfaces also results from unsuitable gating practices. A gating system that causes turbulence of the molten

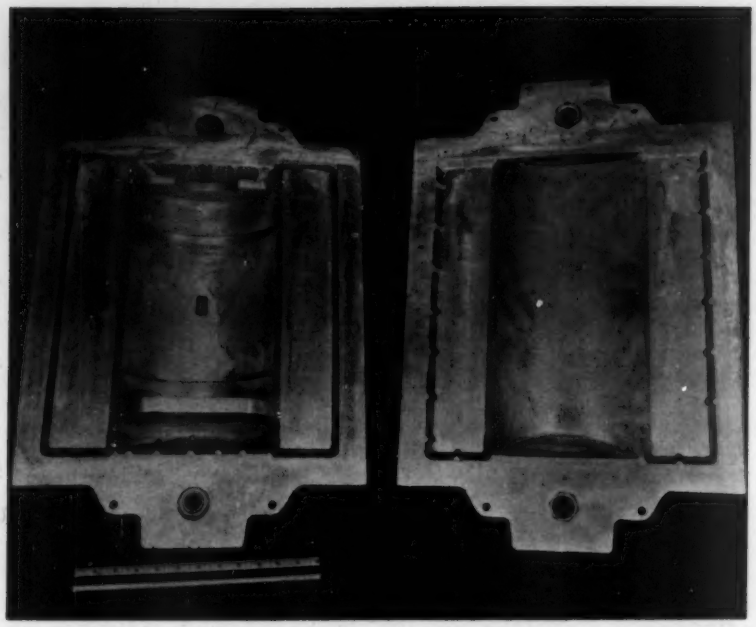


Fig. 1 — Drag and cope sides of a pressure cast aluminum matchplate. Minor surface polishing has been performed.

metal while it is being introduced into the mold or the casting cavity can trap large gas bubbles that rise to cope surfaces. Oxide inclusions usually are associated with these sub-surface cope gas holes. Practices that exclude this type of defect are thorough skimming of the molten metal in the ladle or crucible prior to pouring, keeping the lip of the ladle close to the top of the sprue when pouring, use of small diameter, thin rectangular or tapered sprues and devising a gating system that introduces metal into the casting cavity with a minimum of turbulence.

Generally and uniformly distributed porosity is caused by precipitation of hydrogen bubbles during solidification of the casting. Corrective measures entail prevention of hydrogen absorption by the molten aluminum alloy or removal of the dissolved hydrogen before pouring the metal. Oily, dirty or corroded metal will introduce hydrogen into the melt. Heating the melt to temperatures above 1400 F (760 C), or holding the metal molten for extended time periods, usually will increase the hydrogen content. Melting by directing a flame onto the metal also will introduce hydrogen.

Even though care is exercised to avoid introduction of hydrogen, removal operations may be required. Hydrogen removal can be accomplished by bubbling chlorine or nitrogen gas through the melt. Some salt fluxes also are effective in removing hydrogen. The extent of the treatment required



Fig. 2 — Pressure cast cope and drag aluminum patterns. Backs are contoured to reduce weight.

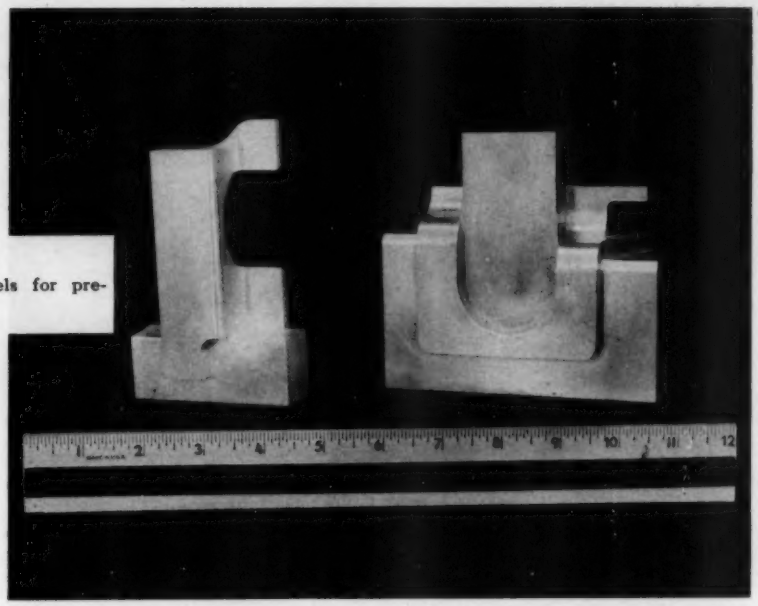


Fig. 3 — Fully machined aluminum models for precision equipment production.

will depend on the hydrogen content of the melt and can be determined by experience, using metal of consistent quality.

Another factor contributing to porosity in castings is the practice of using fine, natural mold sands with low permeability and high moisture contents to obtain smooth cast surfaces. This type of molding sand can cause formation of cope gas holes or gas porosity in castings, and can also be responsible for blow defects. Selection of a synthetic molding sand with an AFS permeability of 15 to 25, and tempering the sand with the lowest practical quantity of water, will provide adequate surface smoothness and avoid introduction of gas into the castings.

### *Warpage*

The most common causes for warpage in machining large pattern castings of aluminum alloys are distortion in clamping, inadequate coolant or dull tools. Machinists experienced in handling ferrous castings may apply excessive force in clamping or chucking aluminum castings. The elastically deformed aluminum casting then changes shape after machining when the clamping forces are removed. Use of insufficient coolant or dull tools can allow overheating, which will cause expansion and distortion of the clamped casting during machining. The casting dimensions and shape then will change while cooling to room temperature.

Some pattern castings have been solution heat treated, quenched and aged to increase hardness and improve machinability. A heat treatment of this type may introduce residual stresses, which can be redistributed during machining and cause warpage. The need for such heat treatment and the attendant risk of warpage in machining can be avoided by adoption of the subsequently recommended pattern alloy.

### *Surface Smoothness*

The smoothness of the machined surfaces will depend largely on the tools and techniques employed. Good machining practices are beyond the scope of this discussion. However, the machinability of pattern castings will be influenced by the alloy composition and heat treatment. Subsequent comments on these two factors will show their effect on machinability and other characteristics.

## ALLOY SELECTION

Some problems that have been encountered with pattern castings of unsatisfactory quality undoubtedly were caused by using alloys with poor casting characteristics. For example, resmelted aluminum sheet scrap is not of suitable composition for producing castings.

A variety of aluminum-silicon and aluminum-copper-silicon alloys have been used successfully in making pattern castings. An investigation of several of these compositions showed that most of them exhibited good feeding characteristics. The experimental castings were free of internal gross shrinkage although relatively few risers were used. However, the best reproduction of the mold cavity shape was obtained with an aluminum alloy containing 7 per cent copper and 5 per cent silicon. This alloy displayed the least surface shrinkage of the depression type in test castings that had inadequate risers.

Significant advantages should be gained by using an alloy especially suitable for the purpose in producing pattern castings. High quality casting can be made more consistently with pattern alloy ingot of controlled and uniform composition. A good pattern alloy composition is aluminum-7 per cent copper-

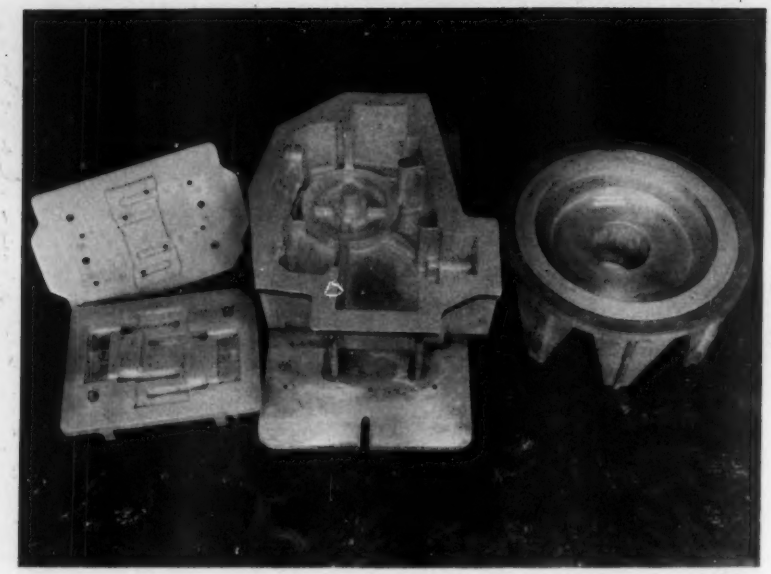


Fig. 4 — Cast and machined aluminum core boxes. *Left* — Box for core blowing, *Middle* — Box for shell core production and *Right* — Box for hand rammed core — steel faced — 15¾-in. O.D.

5 per cent silicon-0.15 per cent titanium with low impurity limits.

Castings of this alloy exhibit generally good machining characteristics, without heat treatment. It is necessary occasionally to make repairs on pattern castings, and established soldering or welding practices are suitable for castings of this alloy.

## SHELL MOLDING PATTERN EQUIPMENT

Production experience has demonstrated that aluminum alloy patterns and core boxes frequently are suitable for shell molding, although cast iron pattern equipment generally has been preferred.[9] The preceding remarks on techniques to produce patterns and core boxes of good quality are equally applicable for shell molding pattern equipment of aluminum. Since shell molding patterns and core boxes operate at elevated temperature, the subsequently described aging treatment always should be employed to insure dimensional stability in service.

## HEAT TREATMENT

Many cast patterns and core boxes are not and do not need to be heat treated. Heat treatment has been used to increase hardness which improved machinability, but this should be unnecessary with the recommended alloy.

Patterns and core boxes that will be operated at elevated temperatures should be heat treated if it is important to retain close dimensional tolerances. Most aluminum alloys are subject to dimensional growth of about 0.001 in./in. when heated to temperatures of 300 F (149 C) or higher for extended periods of time. This dimensional change during use can be avoided by heating the castings at 400 F (204 C) for 4 to 6 hr before machining.

Core driers also would benefit by application of this aging treatment prior to final scraping and fitting. An alternative and suitable annealing treatment for core driers consists of heating at 600 F (316 C) for 2 to 4 hr and cooling in the furnace.

## SUMMARY

Aluminum pattern equipment affords significant advantages for the users. Several factors have been discussed which influence the quality of pattern castings. It is hoped that the recommendations on foundry practices, alloy selection and heat treatment will help producers to supply pattern shops with aluminum castings of high quality.

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# SPECIALTY MELTING AS A SERVICE FUNCTION IN A RESEARCH LABORATORY

by W. F. Moore


## ABSTRACT

**Equipment for vacuum induction melting, consumable and inert electrode arc melting, atomic hydrogen melting and zone melting both horizontally and by the floating zone method are described. Applications of this equipment in the performance of service melting of a wide variety of specialty materials for use in a research laboratory is discussed. In this discussion methods are brought out for determining which of the many types of raw materials, crucibles and melting practices is most suitable to the particular job.**


## INTRODUCTION

There have been for years many methods for melting and alloying metals. The demand for specific materials for more highly specialized uses has forced the improvement of old melting methods and the development of new and better methods for making these materials. It is only natural that this should result in the formation of more and more facilities devoted entirely to specialty melting.

It is extremely important in a research laboratory that each scientist be able to obtain experimental quantities of metals and alloys that he may be working on, without actually building his own equipment to produce them. It is characteristic that the requirements of this type of customer can be satisfied only by special melting procedures.

One of the prime functions of a specialty melting unit in a research laboratory is a service function, and consists of supplying these scientists with samples of any and all types of metals and alloys in forms most useful to them. It is expected that these samples will be of the highest quality possible.

It is not the purpose of this paper to describe the methods for melting any specific metal or alloy but rather to describe some of the equipment in a specialty melting laboratory, and to describe how this equipment can be used to best advantage when practically every job differs from every other job.

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In order to fulfill its mission successfully, a specialty melting unit must be well equipped. While it is necessary to have equipment for all of the conventional types of melting, this paper will be limited to a discussion of those types of melting equipment requiring vacuum, inert, or special atmospheres, which have been used most extensively in the author's company's research laboratory.

The advantages inherent in the vacuum melting processes make it imperative that any specialty melting facility be well supplied with this type of equipment. Some of these advantages are:

1. Melting in the absence of oxygen and nitrogen permits a much fuller utilization of alloying elements.
2. The presence of fewer oxides and nitrides means fewer inclusions and, hence, improved properties.
3. Larger quantities of more reactive elements can be used, thus making possible new and improved alloys.
4. Casting in vacuum permits much higher casting yields with fewer cold shuts.
5. Deoxidation with hydrogen or carbon can be carried out with the consequent formation of gaseous deoxidation products, which can be continuously removed from the vicinity of the molten bath without forming slag.
6. Volatile impurities are removed by evaporation from the melt.
7. Hydrogen is completely removed.

## MELTING PROCESSES AND EQUIPMENT

The methods of melting used extensively in carrying out service melting within the author's company's research laboratory are a) vacuum induction melting, b) cold crucible arc melting, (consumable electrode vacuum arc melting and inert arc electrode melting), c) atomic hydrogen melting and d) zone melting.

Each process along with examples of specific equipment will be described briefly and their use discussed.

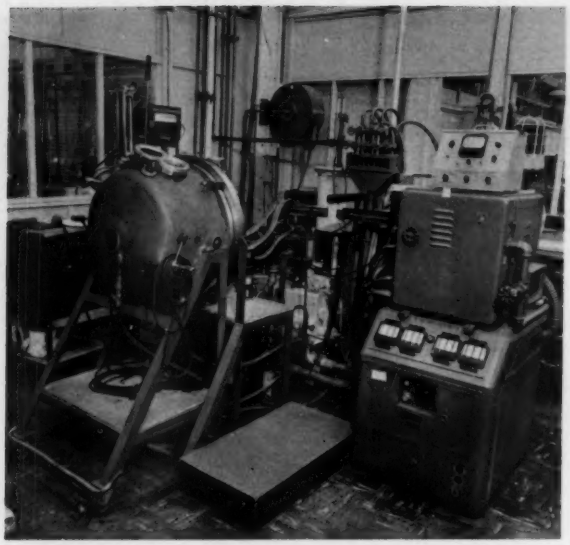


Fig. 1 — Vacuum induction melting furnace (0-2 lb capacity).

### *Vacuum Induction Melting*

The process of vacuum induction melting can be described simply as a charge of metal melted in an induction furnace within an evacuated chamber. The charge is exposed to a refining cycle, determined by the type of material being melted, late additions are made to the heat and the metal is poured within the chamber either in vacuum or in a predetermined atmosphere.

Vacuum induction furnaces are equipped with the necessary gadgetry for performing all of the operations required during the melting cycle. Two separate addition hopper systems are required, one for adding to the original cold charge and one for making late additions to the melt. These can be replenished through air locks. Manual stirring of the melt or breaking of bridges can be easily accomplished. Temperatures can be measured either optically or by immersion thermocouple. Samples can be dipped from the melt and removed from the chamber through air locks.

Inert or deoxidizing gases can be admitted to the chamber. Hot toppings of ingots can be accomplished by means of an electric arc or by electrically heated ceramic sleeves. Shape castings can be made by pouring into ceramic type molds, which have been completely outgassed prior to pour.

There are two types of materials usually made in the vacuum induction furnace. First are those that can be made by conventional air melting methods. In this case the improvements can be attributed to the vacuum melting process, per se. These materials may contain reactive elements in quantities which make air melting a borderline process, or they may be more ordinary materials whose properties are sufficiently improved by vacuum melting to warrant the extra cost involved.

Second are those that must be vacuum melted because of composition. In this case the improvements are largely due to new alloy compositions containing fairly large quantities of reactive materials such as titanium, aluminum or zirconium which can be made successfully only by a vacuum melting process. In either case the quantity of reactive elements that can be melted is limited by their reaction with the crucible material, whether it is of the ceramic oxide type or graphite. The materials that can be vacuum induction melted are also limited to those whose melting points are lower than the temperature at which these same crucibles become unstable.

### *Vacuum Induction Equipment*

Four vacuum induction melting furnaces have been of great use in the specialty melting facility. All of these furnaces permit the same kinds of manipulations. The capacities of these furnaces and melting pressures are shown in Table 1. Photographs of some of these furnaces are shown in Figs. 1 through 4. The 10-30 lb furnace chamber contains a centrifugal casting machine with a 40 in. diameter turntable rotating in a horizontal plane.

TABLE 1 — VACUUM INDUCTION FURNACES

| Capacity, lb | Melting Pressure, Microns Hg |
|---|---|
| 100-450 | 10-35 |
| 10-30 | 10-35 |
| 5-50 | <1 |
| 0-2 | <1 |

### *Cold Crucible Arc Melting*

This type of melting is carried out either by consumable electrode vacuum arc melting or by inert electrode arc melting. In the consumable method, the material to be melted is formed into an electrode. In addition, a small quantity of the material to be melted is placed in the bottom of



Fig. 2 — Vacuum induction melting furnace (0-2 lb capacity).

a cylindrical shaped water cooled copper mold. Crucible and electrode are in high vacuum. An arc is established between the end of the electrode and the starting material in the crucible.

As the material melts off the end of the electrode and drips into the crucible, the electrode is fed downward continuously until it is completely consumed. The heat generated by the arc also maintains a shallow pool of molten metal in the top of the ingot as the ingot is progressively built up from the bottom of the crucible. Degasification of the metal and vaporization of volatile impurities occurs as the metal droplets fall through the high vacuum. Because the molten pool is always quite shallow, the problem of segregation normally encountered in ingot casting is kept to a minimum.

On the other hand, because the metal is maintained in the molten state for such a short period of time, it is sometimes not possible to accomplish as much refining as is desired. Alloys as well as pure metals can be melted in this way by forming the electrodes from blended or pre-alloyed powders or by fabricating composite electrodes from the necessary alloying ingredients. Electrodes of the required composition may also be melted and cast by some other melting process such as vacuum induction melting. Carbon deoxidation may be accomplished by incorporating a predetermined quantity of carbon in the preformed electrode.

In the inert electrode method, the material to be melted is placed in the crucible and the arc is formed between this material and a water cooled tungsten electrode. Additional material may be fed into the crucible to build up the length of the ingot. This method has some disadvantages. There is always the possibility of contamination of the melt by tungsten. This is likely to happen if any of the charge splatters on the tungsten electrode and forms a lower melting constituent than the tungsten itself. This process is also carried out in an inert atmosphere such as argon rather than in vacuum, so that no significant purification takes place.

Both of these methods, of course, eliminate the problem of crucible reaction which is encountered in induction melting. As a result it is possible to melt the more reactive materials such as titanium and zirconium without danger of contamination. Because of the high temperatures attainable in the arc, it is also possible to melt the more refractory materials such as tungsten, molybdenum, tantalum, and columbium. Most of these not only have high melting points but are highly reactive as well.

While it is possible to pour a casting from this type of furnace it is not easily done, and in general this type of melting is limited to buttons and ingot shapes.

### *Cold Crucible Arc Melting Equipment*

The cold crucible arc melting facilities in the company's research laboratory consist of equipment for making both ingots and buttons by consumable electrode vacuum arc melting and by inert arc electrode melting.

There are in operation three principal furnaces,

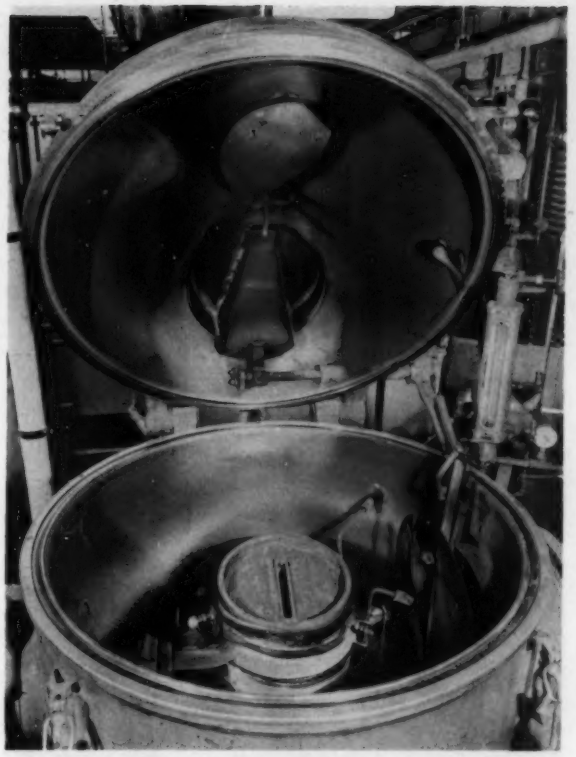


**Fig. 3 — Vacuum induction melting furnace (5-50 lb capacity).**

one consumable electrode vacuum arc melting furnace capable of making up to an 8 in. diameter ingot, one multi-hearth inert arc electrode furnace containing nine hearths and making use of a thoriated tungsten electrode, and one six-hearth furnace which can be used either as a consumable or an inert electrode furnace. The nine-hearth furnace is illustrated in Fig. 5.

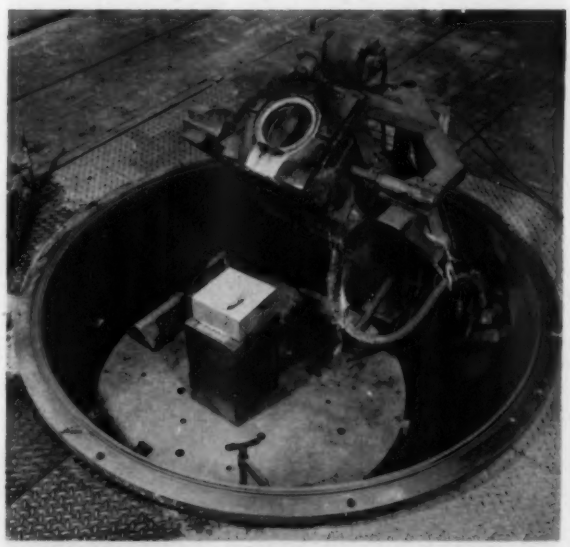


**Fig. 4 — Vacuum induction melting furnace (10-30 lb capacity).**

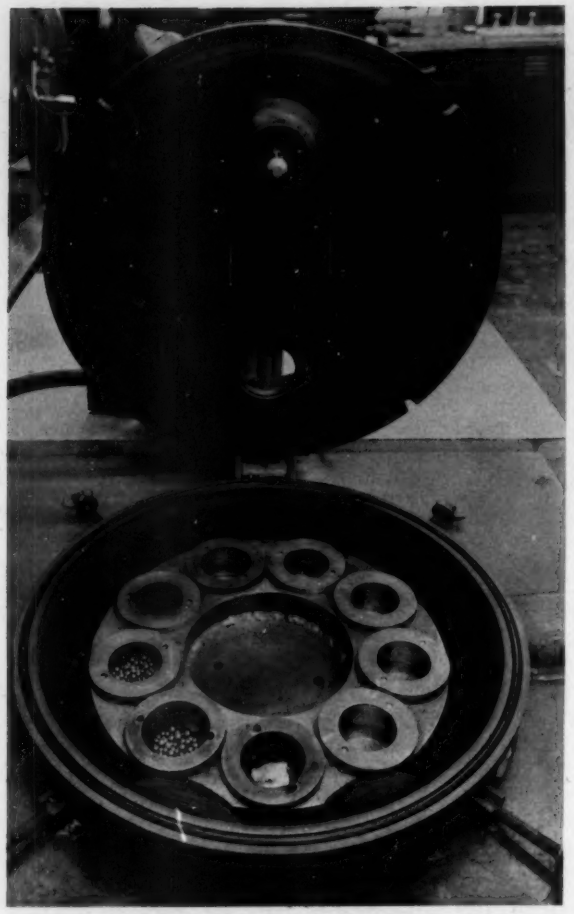


Fig. 5 — Multi-hearth inert arc melting furnace.

### *Atomic Hydrogen Melting*

High temperatures can be reached when atomic hydrogen recombines to form molecular hydrogen. Advantage is taken of this in the atomic hydrogen welding process. It also provides an excellent tool for melting metals and alloys, and comes equipped with its own built-in deoxidizer in the form of hydrogen.

### *Atomic Hydrogen Melting Equipment*

Atomic hydrogen melting in the laboratory is possible on 0-20 gram samples in the furnace illustrated in Fig. 6. In this furnace the ingredients of the melt are placed on a copper hearth. Making use of an atomic hydrogen arc welder, an arc is formed between two tungsten electrodes located above the charge. Hydrogen is admitted to the chamber through orifices around the periphery of the tungsten electrodes. The hydrogen passes through the arc and is dissociated.

It recombines in the vicinity of the metal charge with the evolution of sufficient heat to form a molten bath. This furnace is particularly useful in making small heats where raw material is expensive or not available in large quantities.

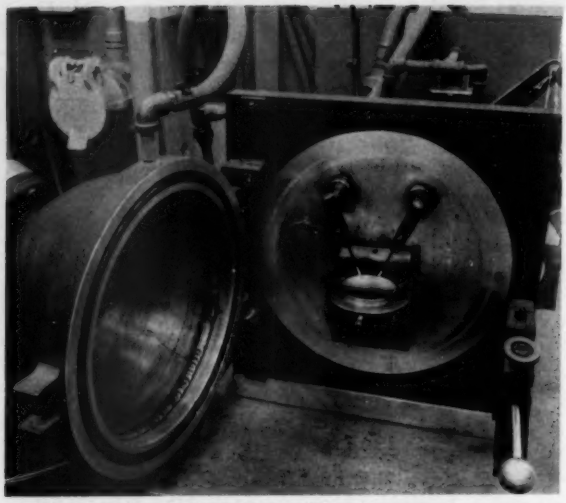


Fig. 6 — Atomic hydrogen melting furnace.

### *Zone Melting*

Zone refining consists of traversing a narrow molten zone of metal slowly from one end of a bar to the other end. At the melting interface, solid metal is melted and added to the molten zone. At the freezing interface, if an impurity lowers the melting point of the base material its concentration in the solid will be less than in the liquid, and hence it will be rejected to the liquid and will accumulate there. If the impurity raises the melting point of the base material its concentration in the solid will be greater than in the liquid, and hence it will freeze out of the liquid.

In this manner some impurities are moved with the molten zone to the end of the bar, and others are left behind as the metal freezes behind the molten zone. Repeated passes of the molten zone through the bar in the same direction will further refine the material until a section somewhere near the center of the bar will be extremely pure. Most impurities will be concentrated at the finish end of the bar with some at the starting end.

Materials refined in this manner reach purity levels which might be measured in parts per billion rather than the usual parts per million, if it were possible to measure them. However, it is often necessary to employ physical measurements such as resistivity as a measure of the purity, because of the inadequacy of the usual chemical or spectrographic methods at these low impurity levels.

Materials which are not contaminated by contact with some crucible material may be zoned horizontally, while contained in a boat. Those which will react with crucible materials may be zoned vertically, without benefit of container, assuming that conditions of surface tension, density of material and any levitating forces that might be applied, are favorable. Obviously this is a slow process and consequently a costly one. Unless there is a need for extreme purity that cannot be obtained in any other manner, this process is to be avoided.

Zone leveling is a term applied to a zone melting technique which almost completely eliminates segregation due to the change in composition of a melt as it freezes. There are a number of different techniques by which this may be accomplished. However, the simplest method is to make several fairly rapid molten zone passes through a charge in each direction.

Zone melting may be accomplished by induction melting, electron beam melting, resistance melting or arc melting. The type material which is to be refined will to a large extent determine the type of equipment needed. For example, it is difficult to melt the highly refractory metals by induction means, so it is necessary to melt these materials by electron beam or arc melting. On the other hand, electron beam melting requires a fairly high vacuum, and this would not be suitable for some high vapor pressure materials.

Induction melting depends upon the material itself being a conductor for its heating ability, but melting in a resistance type furnace does not. On the other hand, induction melting provides a beneficial stirring action to the melt which resistance heating does not. Induction melting also provides some levitating action, which aids considerably in maintaining a floating molten zone without benefit of container.

### *Zone Melting Equipment*

Two pieces of equipment are available in the laboratory for zone melting. Both are powered by a single 450 kc, 25 kw electronic induction heater. Horizontal melting in a crucible or boat is accomplished by passing a quartz tube containing the boat through a fixed induction coil, using a lathe bed and appropriate gears and pulleys as a means of controlling travel speed.

Vertical (floating zone) melting equipment consists of a fixed coil (usually single turn) with a means of passing a bar through the coil vertically. The bar, coil and necessary clamps and chucks are enclosed in a pyrex bell jar. Both pieces of equipment can be pumped to a pressure of $10^{-6}$ of mercury and operated either in vacuum or inert or deoxidizing atmosphere.

The zone refining equipment is illustrated in Figs. 7 and 8.

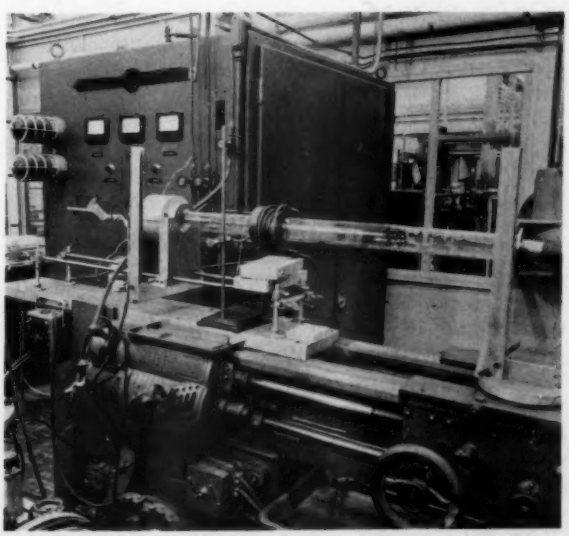


Fig. 7 — Horizontal zone melter.

## MELTING OPERATIONS

### *How to Choose a Melting Process*

Why is it necessary to have a variety of types of melting equipment? Each type of melting has its own advantages and disadvantages compared to each other type. Some melting processes are capable of attaining higher temperatures than others, some can be used with more reactive metals than others, some are limited in the size and shape of product that can be made, some are limited more than others in the amount of purification that can be accomplished and some are inherently too expensive for the average job.

It certainly is not always possible to find the perfect melting process for each individual melting problem that arises. Compromises must be made.

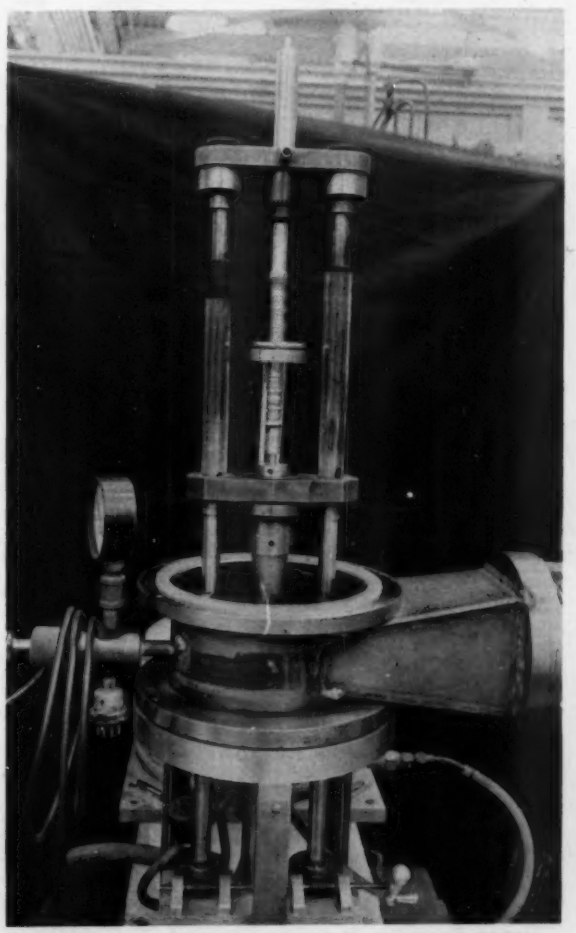


Fig. 8 — Floating zone melter.

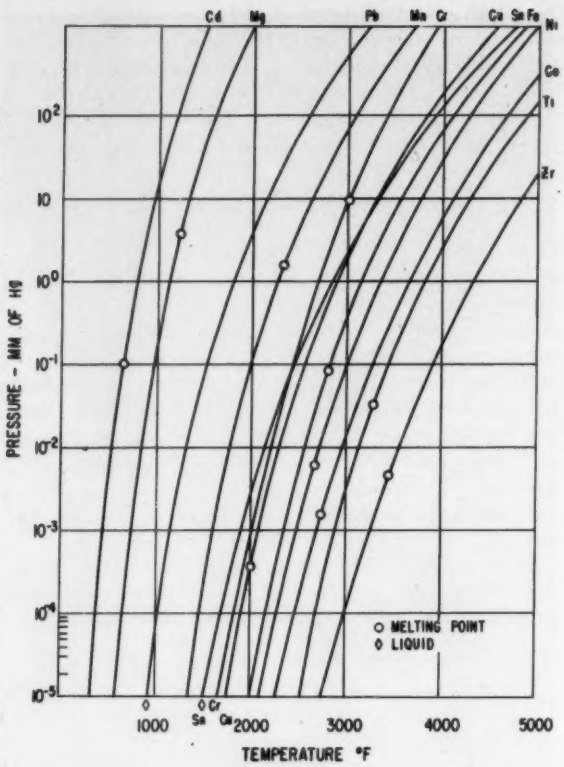



**Fig. 9 — Curves of vapor pressure as a function of temperature for some metals.**

*Questions to Be Answered*

In order to determine how a particular metal or alloy is to be processed, a number of questions must first be answered:

1. Is the material a pure metal or an alloy? If an alloy, is phase diagram information available?
2. What is the melting point of the metal or alloy and of each constituent?
3. What is the vapor pressure of each constituent?
4. How reactive is each constituent of the alloy?
5. What purity is desired? How much purification must be accomplished in the melting operation?
6. Is high vacuum desirable or necessary?
7. Are there large differences in density in the alloying ingredients?
8. In what form are the raw materials available?
9. What purity raw materials are available?
10. What shape product is desired?
11. Is slow cooling or chill casting required?

*Tools*

Some of these questions are answered by the customer. For the answers to others, it is necessary to have available some basic tools. These should include:

**Phase Diagram Information.** An extensive collection of phase diagrams is invaluable in determining melting point data, solubility data, presence of compounds, eutectics, etc.

**Vapor Pressure Data.** Vapor pressure versus temperature data for all metals is necessary to determine what degree of vacuum can be utilized in each particular application. It is possible that some impurities, by virtue of vapor pressures higher than the base materials, may be removed from the alloy without loss of base metal at pressures intermediate between the vapor pressure of the base material and the impurity. On the other hand, if melting is done at too low a pressure, it is possible to concentrate low vapor pressure impurities while evaporating base metals. Figure 9 is a curve of vapor pressure versus temperature for a number of metals.

**Thermodynamic Data.** Thermodynamic data on formation and dissociation of oxides, nitrides and hydrides of metals is invaluable in determining temperatures and pressures at which reactions will proceed. Figure 10 shows the standard free energies of formation of some metal oxides as a function of temperature.[1] Superimposed on this chart are isobars of the reaction of carbon with oxygen as a function of temperature. This is a particularly useful tool in the specialty melting area. It enables a prediction to be made whether crucible reaction will be a limiting factor in the melting of a particular metal or alloy.

It aids in deciding whether deoxidation is necessary and in picking the most effective deoxidation technique. For example, in using this diagram, if a line is drawn from the *O* index at the left of the figure to intersect the line representing the FeO heat of formation at 3000 F (1650 C), the extension of this line will intersect the $Po_2$ scale at a value of $10^{-6}$ mm of mercury, indicating that FeO will be dissociated under these conditions of temperature and pressure. This reaction does not help much in an iron alloy because iron itself vaporizes at approximately 300 microns of mercury at this temperature.

Thus too much iron would be lost by vaporization if the heat were to be made at a low enough pressure to dissociate the FeO. Obviously then, in order to produce an iron alloy with low oxygen content something more than low pressure is necessary. Some form of deoxidation is needed. If a line is drawn from the *C* index at the left of the diagram to intersect the FeO heat of formation line at 3000 F (1650 C), the extension of this line intersects the $CO/CO_2$ Ratio scale at a value of 7/1, indicating that FeO will be reduced by carbon at 3000 F (1650 C) in an atmosphere with the $CO/CO_2$ ratio greater than 7/1.

Similarly, using the *H* index and the $H_2/H_2O$ Ratio scale, we find that FeO will be reduced at 3000 F (1650 C) in an atmosphere with the $H_2/H_2O$ ratio greater than 1/1. Therefore, either carbon or hydrogen can be used for deoxidation.

It should be remembered in using this diagram that no indication of rates of reaction is given.

**Raw Material File.** A complete up-to-date file on available sources of raw materials including purity and price must be maintained. There is usually a choice of raw materials available for most jobs. Purities vary from the ferroalloy types to ultra high purity zone refined materials. Prices vary from cents/lb to hundreds of dollars/lb. It is often nec-

essary to compromise on purity for the sake of economy. It is always necessary to use common sense in choosing the purity of raw materials.

In vacuum melting gases are removed from the melt. If the raw materials are extremely high in gas content, the process of their removal may be quite violent. Gas content is therefore an important factor in the choice of raw materials. In considering purity of raw materials we must also keep the application in mind. What might be considered high purity in one application would be impure in another. Using silicon as an example, there are various grades of ferrosilicon which are available.

A 98 per cent silicon at a cost of 29 cents/lb is

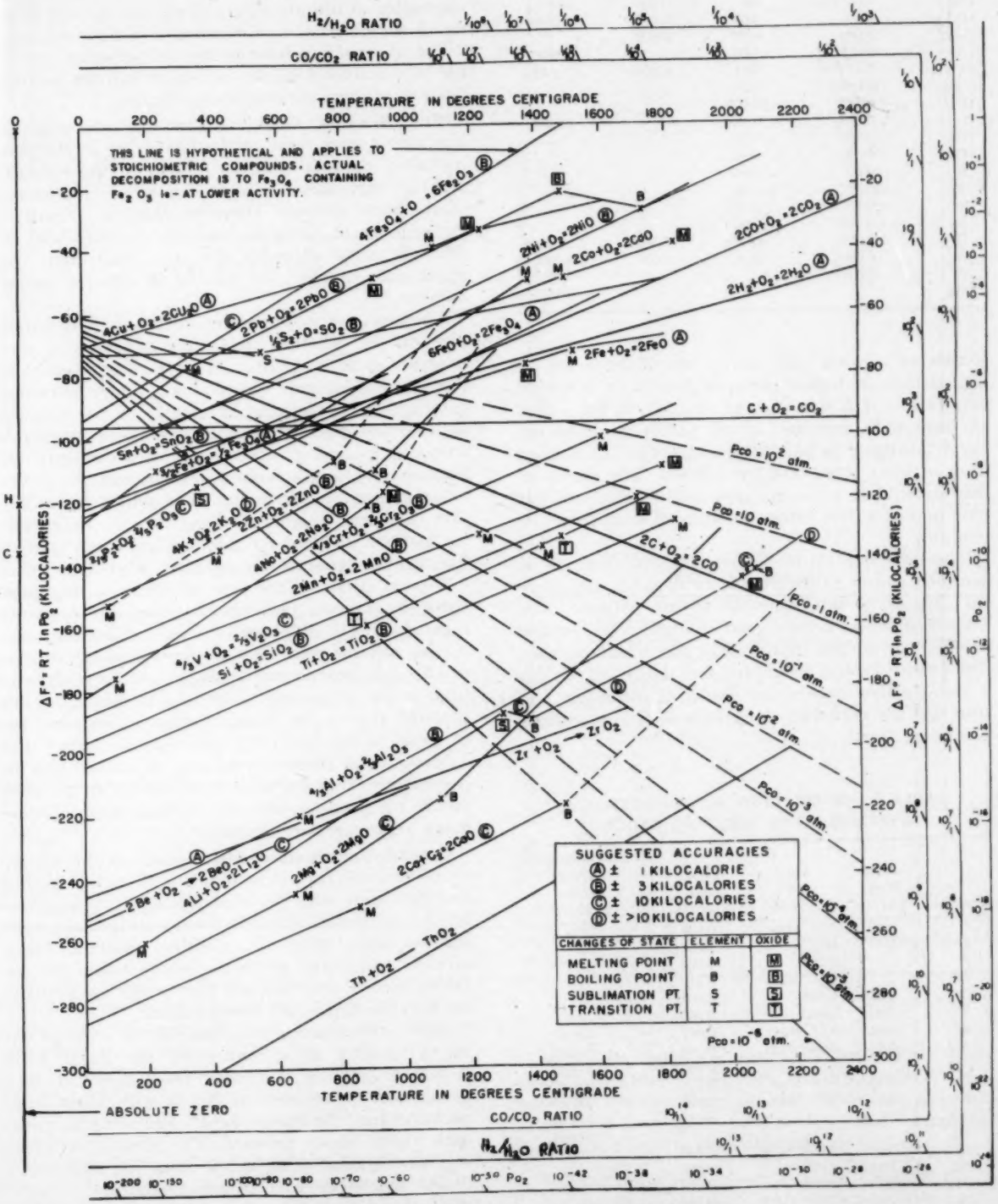



Fig. 10 — The standard free energy of formation of many metal oxides as a function of temperature.

TABLE 2 — SOME CHROMIUM RAW MATERIALS AVAILABLE FOR MELTING STOCK

| Element | Elect. Chromium Metal, Regular $1.16/lb | Elect. Chromium Metal, Vac. Grade $1.19/lb | Elect. Chromium Metal, Grade A $28.00/lb | Iodide Chromium, $100.00/lb |
|---|---|---|---|---|
| Cr | 99.85 | 99.35 | 99.95 | 99.997 |
| Fe | 0.13 | 0.30 | 0.009 | 0.0002 |
| C | 0.01 | 0.05 | 0.006 | 0.001 |
| O | 0.45 | 0.04 | 0.008 | 0.0008 |
| Si | 0.004 | 0.04 | 0.009 | |
| Al | 0.002 | 0.05 | | 0.0001 |
| S | 0.025 | 0.01 | 0.004 | |
| Mn | 0.002 | 0.01 | | |
| Pb | 0.002 | <0.001 | | |
| Cu | 0.006 | 0.003 | | 0.0001 |
| Ni | <0.001 | <0.001 | | |
| Co | <0.001 | <0.001 | | |
| B | <0.001 | <0.001 | | |
| Sn | <0.001 | <0.001 | | |
| Bi | <0.001 | <0.001 | | |
| Sb | <0.001 | <0.001 | | |
| H | 0.008 | 0.001 | 0.0003 | 0.00008 |
| N | 0.035 | 0.01 | 0.007 | 0.0003 |
| P | 0.002 | | | 0.0003 |
| Mg | | | | 0.0001 |

considered a good high purity material for an air melted heat. If higher purity is desired in a special application, it is necessary to pay $17/lb for a 99.9 per cent pure material. A still higher grade of purity is available in smaller quantities in a densified form of solar cell grade for $100/lb. The purity of this material is no longer measured in terms of per cent impurity, but rather in units of resistivity (3-5 ohm-cm).

The next step up in purity is a $300/lb grade of densified silicon whose resistivity is about 100 ohm-cm. To carry this example to its extreme, this $300/lb material is only good enough to be the raw material in further purification for transistor uses. Two other typical examples are shown in Tables 2 and 3 where some of the grades of chromium and iron that are available, along with their approximate prices, are listed.

TABLE 3 — SOME IRON RAW MATERIALS AVAILABLE FOR MELTING STOCK

| Element | Ingot Iron, $.07/lb | Elect. Iron, $.29/lb | Elect. Iron, $1.65/lb | Spectrographic Grade Iron, $130.00/lb |
|---|---|---|---|---|
| C | 0.02 | 0.01 Max. | 0.004 | 0.03 |
| O | 0.07 | 0.07 | 0.05 | 0.01 |
| N | | 0.004 | 0.004 | 0.01 |
| H | | 0.10 Max. | 0.01 | |
| S | 0.012 | 0.004 | 0.004 | |
| P | 0.010 | 0.003 | 0.003 | n.d. |
| Si | 0.04 | 0.01 Max. | 0.001 | 0.0007 |
| Mn | 0.05 | 0.005 Max. | <0.001 | 0.0003 |
| Mg | | | <0.0005 | 0.0002 |
| Ni | 0.01 | 0.05 Max. | 0.02 | 0.0002 |
| Cu | 0.07 | 0.01 Max. | <0.001 | 0.0001 |
| Na | | | | 0.0001 |
| Ag | | | | <0.0001 |
| Cr | 0.01 | 0.005 Max. | <0.003 | n.d. |
| Co | 0.007 | 0.007 | 0.005 | n.d. |
| Mo | <0.01 | 0.002 | <0.004 | n.d. |
| Al | 0.002 | 0.002 | <0.003 | n.d. |
| V | <0.01 | | | |

**Physical Property Data.** Complete and up-to-date information on the physical properties of the elements including melting point, density, coefficients of thermal expansion and values of electrical resistivity are particularly important. For example, a metal which is not an electrical conductor will not be heated in an induction coil at the frequencies normally used, i.e., 450 kc and below. It will also be impossible to initiate an arc in the arc melting process, if the material to be melted is a nonconductor. Use of resistance heating may be indicated, or the use of a susceptor in the induction coil for indirect heating of the non-conducting charge.

**Melt Records.** Complete melt records must be kept of all work done in the laboratory in the past, including chemical analysis results. In the specialty melting field there is a minimum of duplication of the same analysis. However, there is a limit to the number of elements available to work with so that the same elements are used over and over again, even though they may be in different combinations.

Therefore, it is important that any information that can be obtained about the behavior of each element under a specific set of melting conditions should be accurately recorded for future reference. Information concerning reactivity of an element with certain crucible materials can be of invaluable service in planning future work with this material. Records of chemical analysis on all metals provide a background of information that allows the calculation of charge weights resulting in accurate chemistry without the need for intermediate check analyses during the course of the melting process.

Typical curves of recovery of metals during processing by vacuum induction melting are shown in Figs. 11 and 12. These curves are reliable for most of the melting and deoxidizing techniques in use in the laboratory, and apply to all sizes of heats from a few grams up to 500 lb. It should be emphasized that in the vacuum melting processes, loss of metal is due mainly to vaporization rather than oxidation and formation of slag. It should also be remembered that losses of some metals in an alloy due to vaporization will cause concentration of other lower vapor pressure elements.

**Crucible Information.** Information on the reactivity of various crucible materials with metals, carbon and hydrogen must be available. Crucible materials which are readily available for use in specialty melting are silica, magnesia, alumina, zirconia, silicon carbide, zircon and graphite. Others which are useful in special cases but are not as readily available are beryllia, thoria and boron nitride.

Most oxide crucible materials will be reduced during a melting cycle, either by the metal being melted or by the deoxidant used. Some of these oxides will be reduced to metals with lower vapor pressure than the charge metals and some to metals with higher vapor pressure. The chances of removing this contaminating metal from the melt are of course much better if its vapor pressure is higher than that of the charge materials. On the other hand, the more resistant the oxide is to reaction

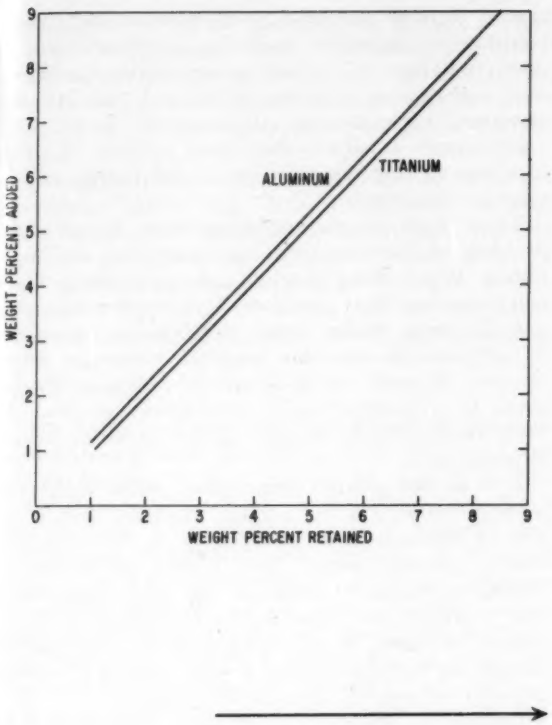



Fig. 11 — Recovery curves for titanium and aluminum in vacuum induction melting.

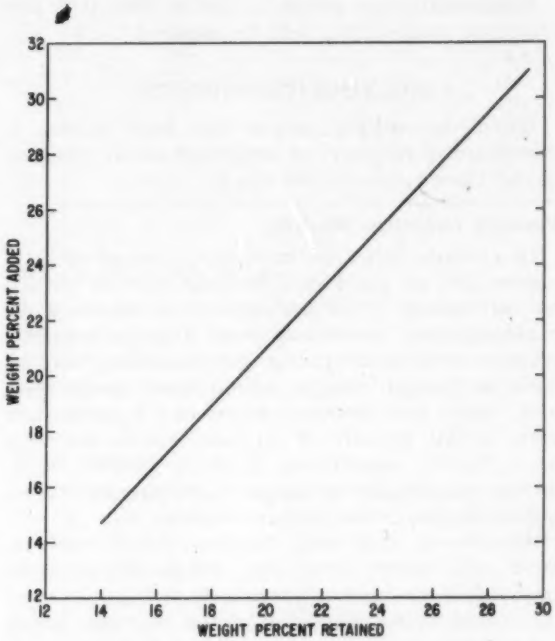



Fig. 12 — Recovery curve for chromium in vacuum induction melting.

during melting, the less chance there is for contamination at all.

In general, stability of the oxides used in crucible manufacture, in contact with metals at high temperatures in order of decreasing reactivity is:[2] $ThO_2$, BeO, $ZrO_2$, $Al_2O_3$ and MgO. The temperatures at which some of these crucible materials show appreciable volatility in vacuum are:[3]

| Material | Temperature, F (C) |
|---|---|
| MgO | 3092 (1700) |
| BeO | 3812 (2100) |
| $ZrO_2$ | 4172 (2300) |
| $ThO_2$ | 4172 (2300) |
| Graphite | 4172 (2300) |

Of the oxide materials, thoria, zirconia and beryllia appear to be particularly useful in vacuum applications. However, of these, thoria and beryllia have some disadvantages. They are both quite expensive, thoria is radioactive and beryllia has a toxicity problem. In actual practice in the laboratory, magnesia, alumina and zirconia have proved to be satisfactory crucible materials in all but a few exceptional cases. Table 4 is a list of the metals with which the laboratory has become somewhat familiar, and the types of crucibles in which they can be melted with a reasonable chance of success.

TABLE 4 — CRUCIBLES USED FOR MELTING METALS

| Graphite | $Al_2O_3$ | MgO | $ZrO_2$ | Zircon | $SiO_2$ | BeO | $ThO_2$ | SiC | Water Cooled Cu | Ta |
|---|---|---|---|---|---|---|---|---|---|---|
| Al* | Al | Al | Cr | Al | Si* | Cr | Ti | Au* | Cr | Gd* |
| Sb* | Sb | Sb | Co | Co | | | Zr | Mg | Cb | Y* |
| Bi* | Bi | Bi | Au | Fe | | | | | Mo | |
| Cd* | Cd | Cd | Fe | Ni | | | | | Ta | |
| Cu* | Co | Co* | Ni | | | | | | Ti | |
| Ga* | Cu | Cu | | | | | | | V | |
| Au | Ga | Ga | | | | | | | W | |
| In* | Au | Au | | | | | | | Zr | |
| Pb* | In | In | | | | | | | | |
| Mg* | Fe | Fe* | | | | | | | | |
| Ag* | Pb | Pb | | | | | | | | |
| Sn* | Ni | Mn* | | | | | | | | |
| Zn* | Pd | Ni* | | | | | | | | |
| | Ag | Ag | | | | | | | | |
| | Sn | Sn | | | | | | | | |
| | Zn | Zn | | | | | | | | |

*Indicates crucible with which best results have been obtained in this laboratory.

When the questions previously posed have been answered satisfactorily, the melting process best suited to the problem can be chosen. In general, if the metal or alloy falls within these categories it may be vacuum induction melted:

1. Melting point not greater than 3272 F (1800 C).
2. No harmful reaction with either graphite or oxide crucible.

3. The purity required does not exceed 99.99 to 99.999 per cent.

If categories 1 and 2 cannot be met and number 3 is the same as above, then a need for cold crucible arc melting is indicated.

If extremely high purity is desired, then it is possible that zone refining will be required.

## MELTING TECHNIQUES

When the melting process has been chosen, it then becomes necessary to determine which variation of the chosen process will apply.

### *Vacuum Induction Melting*

In vacuum induction melting a number of techniques can be employed. Melting may be carried out in vacuum or in atmospheres of inert or deoxidizing gases. Gases may be at a full atmosphere pressure or at some partial pressure. They may be static or dynamic flowing atmospheres. Metals with high vapor pressures may necessarily be melted at some partial pressure of an inert gas to minimize losses due to vaporization. Either a carbon or hydrogen deoxidation technique will normally be required to insure low oxygen content.

The use of solid deoxidants is not desirable as these will usually form slag, which is not easily removed in the vacuum furnace. If hydrogen is used, it is necessary to remove all of the hydrogen before pouring an ingot. If carbon deoxidation is used it is necessary to complete the carbon boil reaction before the pour, as a resumption of the boil in the mold will result in gassy metal.

Even though it is impossible to make high quality material when reactive, pure metals are melted in oxide crucibles, it is possible to add considerable quantities of these metals to form alloys. For example, 80 per cent chromium alloys and 10 per cent titanium or zirconium alloys are possible in oxide crucibles. Ideally, the metals in the initial charge, which normally consist of such metals as iron, cobalt, nickel, chromium, tungsten and molybdenum, should be completely deoxidized before making any late, reactive additions. In this manner complete utilization of the alloying effects of these additions can be realized.

When an inert gas or hydrogen is used, whether it is in induction or arc melting, it is important that the gas be pure and dry. Dew points approaching −100 F are desirable. Purifiers for removing oxygen and dryers for removing water from hydrogen are available. Passing argon over a bed of hot titanium chips will getter most of the oxygen and nitrogen from it, and water may be removed further by trapping. Molecular sieves can also be used to accomplish the same purpose.

### *Susceptor Use*

Not all materials can be melted by direct coupling in an induction furnace. In this case it may be necessary to use a susceptor around the outside of the melting crucible. Heating of the charge is now accomplished by conduction of heat from the susceptor, through the melting crucible to the charge. Graphite is normally used for susceptors, but if there is danger of carbon contamination, molybdenum, tantalum or tungsten can be used. One problem associated with the use of susceptors is the high temperatures at which they must operate in order to reach melting temperatures in the charge in reasonably short times.

These high temperatures aggravate the reaction problem of the susceptor with adjoining oxide ceramics. When using graphite susceptors, lamp black makes an excellent insulator between the susceptor and the coil. When using molybdenum, tantalum or tungsten as susceptor material, either an oxide ceramic material or a series of radiation shields made of a material such as molybdenum may be used. Boron nitride is also a useful material in this application.

It is of the greatest importance, when using any ceramic material in a vacuum furnace, that it be free of moisture and adsorbed gases. All crucibles, molds, pouring basins, etc., should be stored in heated or evacuated areas. It has been experienced that a preliminary conditioning of melting crucibles greatly decreases oxygen pickup from the crucible during the melting process. A wash heat or a furnace fire will accomplish this conditioning.

Casting to shape in a vacuum furnace can be successful only if the mold has been completely dried and has no organic binders to burn out. Of course, casting can be done in organic molds if the furnace chamber is brought to atmospheric pressure before the pour. However, the advantage of pouring into an evacuated mold is lost, with a consequent decrease in yield of good castings. A material made up of various proportions of zircon sand and flour with an aluminum phosphate bond has been used extensively in the laboratory, and has been found to be a universal material for molds, pour basins and pour lips.

A preliminary bake at 392 F (200 C) followed by a firing treatment at 1832 F (1000 C) will prepare this material for use in vacuum. Investment type molds can also be used successfully, including the glass-cast types, if they are properly conditioned.

### *Cold Crucible Arc Melting*

If it is not possible to melt a material in the vacuum induction furnace because of high melting point or reactivity with crucible materials, then it is possible that the material may be successfully made by arc melting in a water cooled copper mold. The greatest degree of purification will be accomplished in the consumable electrode vacuum arc process. Here gases and volatile impurities are removed in vacuum at high temperature. If the vapor pressure of the material is excessively high, partial pressures or full atmospheres of inert gases may be used.

Consumable electrode melting requires that an electrode be formed, either by powder methods (hydrostatic pressing works well) or by fabricating from an ingot or sheet material. If the material to be melted is expensive or not available in a form

easily converted to an electrode, or if only a small quantity is required, then it is possible to melt small quantities (25 to 150 grams) in the button furnace using a tungsten electrode and an inert atmosphere. A static atmosphere of purified argon is used.

This atmosphere is further gettered by melting a zirconium button first and if necessary alternate alloy buttons and zirconium getter buttons. Another method of gettering the atmosphere would be to install a resistance heated strip of titanium or zirconium in the furnace chamber and maintain it at a temperature of about 1382 F (750 C) during melting.

### *Hot Topping*

Hot topping of inert arc melted buttons is accomplished by gradually decreasing the power at the end of the melting cycle. This can be done with consumably melted materials if the adapter to which the electrode is attached is made of tungsten and can be used as a hot topping electrode. In arc melting it is a little more difficult to predict recoveries of various elements in an alloy than in vacuum induction melting because of the higher temperatures involved. This fact makes it even more important that good records be kept on all heats.

Button melting of good heat conductors or high melting point materials is sometimes difficult because in contact with a cold, water cooled copper mold, the heat tends to be withdrawn from the button faster than the arc can replace it. This may result in only one small portion of the button being molten at any one time. This problem is alleviated by rotating the arc around the surface of the button, by stirring the melt with a stirring coil and by melting the button from both top and bottom by turning it over at least once.

Using these precautions it is possible to make sound, homogeneous materials, in quantities sufficient for preliminary evaluation. If results warrant further work, larger ingots can be made.

Kinds of materials which are particularly troublesome in arc melting are alloys in which a high melting reactive material is combined with a low melting, high vapor pressure material. An extreme example of this would be an alloy of titanium and cadmium. The boiling point of cadmium is lower than the melting point of titanium. Tin has a much lower vapor pressure than cadmium, but the problem of a combination of low and high melting point materials is still present in an electrode made up of titanium and tin.

The heat conducted and radiated to the part of the electrode not yet melting in the arc, would heat this portion of the electrode enough to cause the low melting tin to melt and either break the electrode or diffuse out and drop into the melt. Even making a button of this type of alloy in an argon atmosphere, it would be difficult to predict losses of the more volatile element. However, it is surprising the number of different types of alloys that have been made successfully in these arc melting furnaces. Some of the metals that have been melted in one combination or another are listed in Table 5.

TABLE 5 — VARIETY OF METALS MELTED IN SPECIALTY MELTING FACILITY ARC MELTING FURNACES

| | | | | |
|---|---|---|---|---|
| W | Y | Mn | Ge | Sr |
| Mo | Gd | Mg | Cu | Mn |
| Ti | Si | Li | Al | Pd |
| Ta | Ni | Ba | Ag | Th |
| Zr | Fe | La | Ca | Pb |
| V | Co | Hf | B | Nd |
| Cb | Cr | Ce | Sn | Pr |
| As | Te | Dy | Pt | Er |
| Be | C | S | Ga | Au |
| Tl | Ir | Bi | | |

### *Atomic Hydrogen Melting*

There are no special techniques involved in the atomic hydrogen melting process except that a gradual decrease in power is necessary at the end of melting in order to hot top the button and keep the shrink up at the top surface.

### *Zone Refining*

In zone refining, the same principles governing the use of high vacuum or atmosphere pressures apply as have been previously discussed for the other melting techniques.

In general narrow molten zones and slow zone travel will effect greater purification than broad zones and rapid zone travel. One method of speeding up this process in the case of horizontal zoning in a crucible is to have a number of molten zones passing through the bar at the same time.

Occasionally requests for single crystals or large crystals of some specific metal or alloy are received. The zone refining apparatus and process are ideal for this purpose. However, a disadvantage is introduced by use of our 450 kc induction source in single crystal growing, since the amount of stirring which is induced in the charge is excessive. This agitation is more conducive to the nucleation and growth of fine grains than large grains.

## SUMMARY

Several specialty melting processes and the methods for determining which of these processes will best suit a particular melting job have been discussed. Some of the necessary tools that are used in making this determination are described. Using all of the available data it is possible to determine theoretically how a heat should be processed. However, there is still no substitute for practical experience to back up the theory in specialty melting.

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# HIGH STRENGTH ALUMINUM ALLOY AIRFRAME CASTINGS

by W. A. Bailey and E. N. Bossing


## ABSTRACT

**There is a growing trend toward the use of high strength light metal alloy castings as replacement for weldments, machined from bar parts and subassemblies. In the airframe industry this trend has produced considerable savings in cost, time and in facilities necessary for the production of components, as well as savings in weight. This has come about because the cast structure is stronger and more reliable than formerly. High strength and confidence levels can be obtained in almost any properly engineered configuration, and this is of great importance to the airframe industry. These demands must be met or other production methods could replace even the presently cast configurations.**


## INTRODUCTION

The popularity of high strength, light metal casting alloys, apparently passes through a number of historically recognizable phases, as we can roughly identify a time with a given high strength casting alloy. With magnesium the progression has been from AZ63 to AZ91 to the edge of the ZK51-ZK61-QE22 era. With aluminum it has been from the brittle epoch of 195 through a brief and mostly painful 220 period and then, rapidly into the aluminum-magnesium silicide alloys. Today could be called, according to these chronological systematics, mid 356.

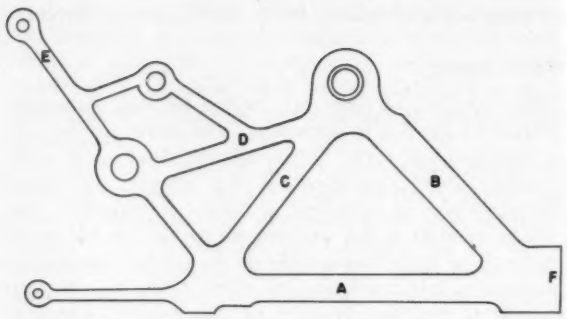
To all the various modifications of the parent 356 aluminum alloy—A356, 356 CMB, X357, 356Be, Tens-50—it is thought that much of the current increased emphasis on cast structure in airframes is owed.

Within the authors' company, as elsewhere, there is a growing trend toward the use of high strength light metal alloy castings as replacements for weldments, machined from bar parts and subassemblies. To those in the airframe industry, the consequence of this trend seems increasingly attractive, for such use has already produced considerable savings in cost, in time, in the facilities necessary for the production of components and, happily, in weight. The origin of this trend must be in a generally held opinion that cast structure is stronger and more reliable today than it was formerly. This opinion appears to be fairly well founded in fact.

## LANDING GEAR ASSEMBLY

As an example, there was a cast 356 aluminum alloy bracket in a landing gear assembly that failed.

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356-T6 ALUMINUM ALLOY

| Location | Purchased Per QQ-A-601 FTU | FTY | E,% | Purchased Per DMS 1561-A FTU | FTY | E,% |
|---|---|---|---|---|---|---|
| A | 20.9 | 16.5 | 3.0 | 48.9 | 39.5 | 8.0 |
| B | 25.5 | 18.7 | 2.0 | 49.5 | 40.1 | 8.0 |
| C | 24.3 | 16.0 | 4.0 | 49.5 | 38.5 | 10.5 |
| D | 26.6 | 20.6 | 2.0 | 49.4 | 38.9 | 10.0 |
| E | 30.9 | 20.6 | 5.0 | 50.4 | 38.9 | 16.0 |
| F | 30.6 | 20.8 | 5.0 | 50.9 | 37.9 | 17.0 |

**Fig. 1 — Bracket in landing gear assembly. Letters indicate test coupon locations.**

The casting was purchased under the Federal Specification QQ-A-596, and is diagramed in Fig. 1. The mechanical properties obtained from a similar casting of the same heat, which incidentally was filled with gas porosity, are listed in the left hand column. The letters on the diagram show the location of the test coupons. This part according to the Federal Specification is a legal part despite the low mechanical properties and despite the fact that the casting failed in use. It required only 1000 lb to break a similar test casting statically loaded to failure.

Legal part or no legal part, the authors' company grew discouraged with both the producing foundry and some aspects of the Federal Specification. It was tried again using a new foundry and a new specification (company Material Specification 1561). This requires both a high purity A356 aluminum alloy chemical composition and guaranteed minimum mechanical properties in a specified highly stressed area of the casting. What was learned about its mechanical properties, is listed in the right hand column of Fig. 1. A similar part failed on static test at about a 3000 lb load, approximately three times the loading of the first casting. There has never been a service failure of these castings.

This example of the basic improvement in mechan-

ical properties of coupons cut from castings and in the static strength of cast structure is typical of many high strength castings now in use at the authors' company. It is due primarily, not to a new and startling alloy but to the better application of well known foundry principles, principles which have been recognized for years. Unfortunately, most of the attention which has been paid to them is paid in lip service, and lip service does not improve the quality of a given casting anymore than does shining x-rays through it.

Of course, some of this improvement has come through alloy modification, as in the case of 356 aluminum alloy, where apparently a decrease in iron content is reflected by a corresponding increase in resultant elongation. This increased elongation, as you know, can be taken advantage of by increasing artificial aging times and temperatures to higher yield strengths while maintaining a given ductility. But uncomplicated by chill, by tailored heat treat, how much of this is due to actual reduction of iron content? Rather less than initially thought.

## TENSILE DATA

The origin of the data, presented in the table, are from a number of tensile coupons cut from several castings poured in each of three aluminum alloys. The castings were poured into green sand and were identically gated. The differences here should be principally due to alloy. As can be seen, the major variation in chemistry between the 356 and A356 aluminum alloys is one of iron content, the iron in the standard 356 being approximately double that contained in the high purity material.

As can be seen the A356-T6 aluminum alloy material has an advantage over standard 356 in elongation regardless of section thickness—some of this elongation has already been taken advantage of by an apparent increase in artificial aging to a higher yield. The results from Tens-50-T6 aluminum alloy, also shown in the table, are to demonstrate what can be done to mechanical properties by a simple composition change. The beryllium in part negates the embrittling effect of iron, but only in part as witnessed by lower elongation.

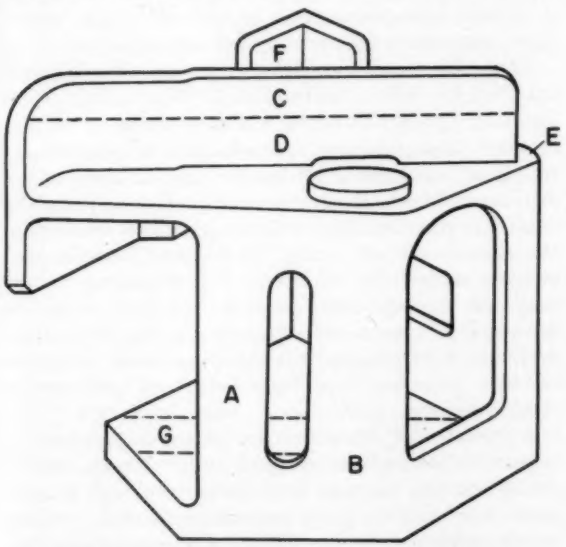
With these results, the extent of the simple effect of composition variation in 356 type aluminum alloys has roughly been demonstrated. The improvement in mechanical properties is significant, but nothing approaching the advance in tensile strengths shown in Fig. 1, especially in elongation.

**COMPARISON OF MEAN MECHANICAL PROPERTIES FOR SIMILAR SECTIONS OF THREE ALUMINUM CASTING ALLOYS**

| Section Thickness, in. | Test | Aluminum Alloy 356 | A-356 | TENS-50 |
|---|---|---|---|---|
| 1¼ | FTU | 28,140 | 32,995 | 38,620 |
| | FTY | 20,525 | 25,715 | 32,850 |
| | E,% | 4.0 | 4.4 | 2.5 |
| ¾ | FTU | 28,800 | 34,115 | 41,330 |
| | FTY | 21,370 | 26,850 | 35,650 |
| | E,% | 4.0 | 5.1 | 4.0 |
| ⅜ | FTU | 30,640 | 35,345 | 42,875 |
| | FTY | 22,110 | 26,910 | 36,880 |
| | E,% | 4.0 | 5.6 | 5.0 |
| ¾ | Bending Modulus | 56,160 | 69,325 | 74,475 |
| ⅜ | Bending Modulus | 61,935 | 73,785 | 80,570 |

**Chemistry**

| Element | Aluminum Alloy, Composition, % 356 | A-356 | TENS-50 |
|---|---|---|---|
| Copper | 0.015 | 0.015 | 0.033 |
| Magnesium | 0.30 | 0.32 | 0.42 |
| Silicon | 7.5 | 7.4 | 8.2 |
| Iron | 0.34 | 0.18 | 0.36 |
| Titanium | 0.11 | 0.11 | 0.13 |
| Beryllium | — | — | + |




**HINGE CASTING — A-356 ALUMINUM ALLOY**

| Location | Foundry A FTU | FTY | E,% | Foundry B FTU | FTY | E,% |
|---|---|---|---|---|---|---|
| A | 32.5 | 27.6 | 1.9 | 46.0 | 32.0 | 13.7 |
| B | 32.1 | 28.1 | 1.6 | 46.2 | 31.7 | 14.8 |
| C | 35.2 | 29.5 | 3.6 | 46.2 | 31.2 | 15.3 |
| D | 35.2 | 29.9 | 2.9 | 45.9 | 31.7 | 14.8 |
| E | 34.4 | 29.2 | 2.3 | 45.8 | 32.1 | 14.8 |
| F | 31.8 | 27.7 | 2.1 | 45.4 | 33.3 | 14.0 |
| G | 33.3 | 29.0 | 2.4 | 45.8 | 32.1 | 13.3 |

**Fig. 2 — Hinge casting produced in sand by two different foundries.**

## HINGE CASTINGS

So there is still left much of the question "why?" The answer perhaps can be found in the data presented in Figure 2. These are data from a number of DC-8 hinge castings found in A356 aluminum alloy. The castings were produced in sand by two different foundries. Both compositions fall within the specification for high purity A356 aluminum alloy—silicon is between 6.5 and 7.5 per cent, iron lies below 0.20 per cent and the magnesium between 0.20 and 0.40 per cent. Both foundries have produced castings in the solution heat treated and aged condition.

Radiographically the castings are comparable. And yet foundry *B* has produced mean mechanical properties of 46.0 ksi ultimate strength, 32.0 ksi yield strength and 14.5 per cent elongation as opposed to 33.0 ksi ultimate, 29.0 yield and perhaps 2.5 per cent

elongation from foundry *A*. Foundry *A* has produced a fairly good average casting, Foundry *B* has produced a superior casting. How did foundry *B* manage this?

**First,** by chilling heavily. Chill, besides disguising the presence of gas does have some function other than moving around radiographically visible defects. This, in conjunction with optimum gating and a sound knowledge of progressive solidification, is to a large extent responsible for the superior mechanical properties.

**Second,** by keeping the magnesium content of this alloy near the upper reaches of the specification—0.35 per cent as opposed to 0.20 per cent in the poorer castings.

**Third,** by adequate heat treatment fashioned to the requirements of the part. Solution heat treat temperatures of 980-1000 F are used standardly on 356 aluminum alloy. The high purity material should in theory contain a smaller quantity of low melting point eutectic compounds. This reduced amount of eutectic would permit the use of a higher temperature in solution heat treatment, resulting ultimately in higher and more reliable mechanical properties. In all probability foundry *B* used a slightly higher solution temperature coupled with an extremely rapid water quench.

This rapid quenching is extremely important. And this fact unfortunately leaves many foundries literally at the mercy of their heat treaters. These heat treaters, it would seem, are often aware enough of the need for rapid quenching to fix the separately cast test bars representing a heat treat load to the periphery of that load, but are not aware enough to similarly quench the castings so represented.

Using a low iron, high magnesium A356 aluminum alloy ingot, with little scrap in the melt, with adequate degassing, with heavy chilling and good gating, coupled with adequate solution heat treatment, a rapid quench and judicious aging, a part is produced with superior mechanical properties, properties which a few years ago would have been considered impossible or at least improbable for production. The high mechanical properties, shown in Fig. 2, were from production castings. The things combining to produce a superior part are not new foundry techniques; they are old foundry techniques at last being used.

## CASTING TECHNOLOGY IMPROVEMENTS

Thus, much of the so-called improvements in casting technology and in alloys and their resultant reflection in part strength have originated not from revolutionary foundry techniques, but from the application of well known principles to existing alloys. The greater part confidence has come not from magic, but from more extensive use of radiographic techniques and from sampling directly the mechanical properties of castings randomly selected from a submitted casting lot – no longer depending upon a separately cast test bar reputedly representative of a given production order. Apparently, awareness is coming that castings are flown, not separately cast test bars.

Recently, of course, there have been many alloy developments in both aluminum and magnesium. These alloy developments most foundrymen grasp eagerly in their constant search for a panacea, a cure-all, a crutch which will enable them to pour sounder, stronger parts with no more effort and no more cost than they are accustomed to expending. To date there is no such alloy now anymore than there was back in 1500 when a gentleman named Theophrastus Paracelus also searched for it unsuccessfully. Foundries in 1961 will produce stronger more reliable parts when they actually apply those principles which they profess to understand.

It would seem that a thesis has been developed – that much of the apparent improvement in casting technology, and its resultant reflection in part strength, has come about almost exclusively from the strength and confidence demands of foundry customers, ably assisted by those foundrymen owning curiosity and technical competence and who are willing to use them. For foundries will, in general, produce no better parts than the poorest part that will be accepted. This is economics, but it is neither science or progress.

## CUSTOMER SATISFACTION

It seems more foundry customers should turn into benevolent despots and continue to force the bulk of founders to advance. To demand high mechanical properties for castings themselves, not in little bars of metal ideally gated. To sample statistically lots of castings to insure themselves of part confidence, not to grossly over-design, and in over-designing, knowingly compensate for the expected foundry error.

That this road has been traveled is evidenced tangibly by MIL specification 21180, by the A revision of QQ-M-56 and by a host of company specifications. However, still a considerable amount of prejudice exists concerning the use of castings in airframes. Much of this prejudice is based on the historical unreliability of castings. It has resulted in the present use of casting factors that produce part weights which are excessive. These casting factors may not be ultimately realistic in light of recent advances in alloys and especially in the application of foundry techniques, but they must serve until the old-new processes have been evaluated and demonstrated reliable.

A handful of foundries are at present producing massive aluminum alloy castings for certain missile configurations, castings with extremely high guaranteed mechanical properties, complicated castings which demonstrate the authors' contention that high strength and confidence levels can be obtained in almost any properly engineered configuration. That this is often difficult is not doubted, but the aim is of paramount importance to the airframe industry. Certainly such strength and confidence level requirements are steadily increasing.

They may well ultimately become mandatory for almost all cast airframe structure. In order for the foundry industry to survive, such demands must somehow be met, or other production methods will replace even those configurations presently cast. An industry that suffers from stasis cannot long survive.

# HIGH STRENGTH YELLOW BRASS CASTINGS WELD REPAIR

by Donald K. Fox


## ABSTRACT

**An investigation was conducted on the repair welding of high strength yellow brass castings. Test castings oxyacetylene gas and inert-gas-shielded-arc welded with several types of fillers were evaluated. Tensile and bend tests on samples removed from these castings showed that satisfactory high strength, high ductility welds could be obtained by inert-gas-shielded-arc welding with base metal composition filler and high manganese aluminum bronze alloy filler. Several production castings welded in this manner supported the experimental results. Additional information is presented on microhardness tests and metallographic interpretation of the weld deposits.**


## INTRODUCTION

The cast high strength yellow brasses, known also as the manganese bronzes, are employed extensively throughout industry for a wide range of applications. The popularity of this class of materials with designers and foundrymen alike is due, to a great extent, to the high mechanical strengths developed in the as-cast condition and the lower cost per pound of finished casting compared to other nonferrous casting alloys providing similar mechanical properties. In our own corporation, the cast manganese bronzes are employed for grid supports in oil circuit breakers, bushings in turbine generators and resistance rings in large motors. Other industrial applications include marine propellers, heavy duty gears, valve and rolling mill bearings.

For the most part, the casting of these high strength alloys presents no difficulty, providing good melting practices are followed and careful consideration is given to gating and risering systems. Occasional difficulties, however, with misruns, shrinkage and dross inclusions do occur in complex castings where adequate feeding, directional solidification and nonturbulent metal flow is more difficult to achieve than with the simpler shapes. In these instances, it becomes advantageous from both the viewpoints of cost as well as of production scheduling to be able to salvage these defective castings by weld repairing.

The amount of commercial as well as technical literature devoted to the welding of copper-zinc alloys, and particularly the high strength yellow brass casting alloys, is not extensive. Although gas welding is normally reported to be the most common technique used for the brasses, and some detailed information is often given to provide a procedure for welding, little has been devoted to the evaluation of welded castings and test bars removed from these castings. In order to obtain such information, which the author's company believes paramount to the acceptance of repair welding as a salvage procedure for high strength yellow brasses, an investigation was initiated on experimental castings oxyacetylene gas welded and inert-gas-shielded-arc welded utilizing several different filler metals. The results of this investigation are described herein.

## TEST CASTINGS PREPARATION

For this investigation, it was decided to conduct mechanical tests and metallurgical examinations on welded high strength yellow brass castings cast to the shape of the test plate, described in the A.S.T.M. Specification B225. A sketch of this plate and location of test specimens is shown in Fig. 1.

In collaboration with the author's company's brass foundry, a rigging which was anticipated to provide sound, dross-free castings was designed and employed in a matchplate pattern for producing the high strength yellow brass test castings in green sand molds. This rigging design utilized a short ½-in. diameter downsprue, 1½-in. diameter by one in. deep sprue well, 1½-in. wide by one in. deep tapered runner and ⅛-in. thick flash gate to provide a gating ratio of 1 : 6½ : 4.

To assure adequate feeding during solidification, four 1½-in. diameter by 4 in. high, plaster insulated risers were provided on the cope side of the castings. A control casting, not to be welded, was cast solid by removing the section of the pattern which provided the 60 degree groove shown in Fig. 2. Radiographic examination of sample plates showed the rigging design to be satisfactory in providing castings suitable for the welding experiments.

## WELDING TEST CASTINGS

The nominal chemical compositions of the test castings and filler metals used in the experiment are shown in Table 1. The high strength yellow brass castings were made of an alloy which is similar in

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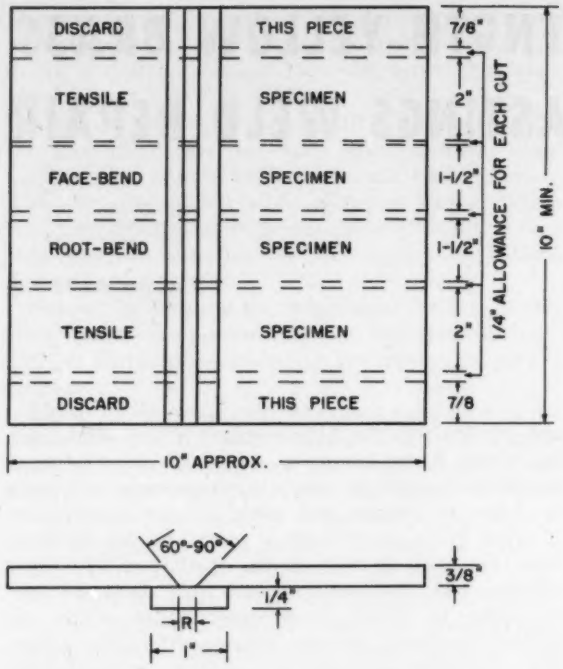



**Fig. 1 — Welding mechanical test assembly.**

composition and mechanical characteristics to the A.S.T.M. 8B manganese bronze sand casting alloy. In the tables, as well as the following discussion, castings and filler metals made of this alloy will be referred to as "manganese bronze 8B alloy."

Three of the test castings were oxyacetylene gas welded employing in two instances filler metals similar in composition to the base metal, and in the other a low fuming manganese bronze filler corresponding to the A.W.S.-A.S.T.M. classification RCuZn-C. The castings were wire brushed in the groove area, preheated to a temperature of about 750 F (400 C) and welded using a slightly oxidizing flame. Fluxing was accomplished by frequently dipping the welding rod into a pool of molten borax. Four passes of 3 min duration each were needed to complete each weld. After welding and cooling to a somewhat lower temperature, the castings were stress relieved at 1022 F (550 C) for one hr and furnace cooled.

Six additional castings were inert-gas-shielded-arc welded with six different filler metals. The fillers were:

1) Low fuming manganese bronze (A.W.S.-A.S.T.M. RCuZn-C).
2) Aluminum bronze (A.W.S.-A.S.T.M. RCuAl-A2).
3) Medium strength aluminum bronze alloy (A.S.T.M. 9A aluminum bronze alloy).
4) High strength aluminum bronze alloy (A.S.T.M. 9D aluminum bronze alloy).
5) High strength manganese bronze alloy (A.S.T.M. 8B manganese bronze alloy).
6) High manganese aluminum bronze alloy.

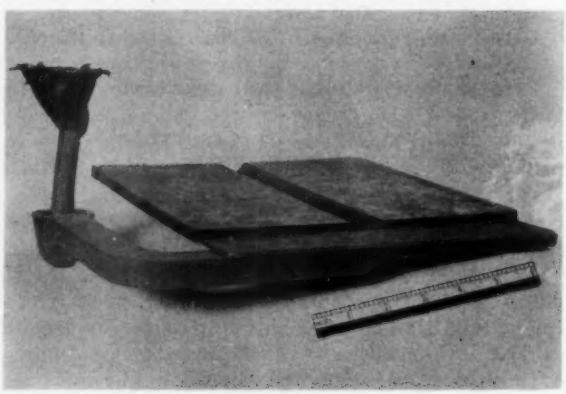


**Fig. 2 — Cast test plate with attached gating system. Gating ratio 1 : 6.5 : 4.**

**TABLE 1 — CHEMICAL COMPOSITIONS OF TEST CASTINGS AND FILLER METALS**

| Welding Process | Control Casting | Oxyacetylene | | | Shielded Arc | | | | | |
|---|---|---|---|---|---|---|---|---|---|---|
| Filler Metal | | Manganese Bronze (8B) | Low Fuming Manganese Bronze | Manganese Bronze (8B) | Low Fuming Manganese Bronze | Aluminum Bronze | Aluminum Bronze (9A) | Aluminum Bronze (9D) | Manganese Bronze (8B) | High Manganese Aluminum Bronze |
| A.W.S.-A.S.T.M. Classification | — | — | RCuZn-C | — | RCuZn-C | RCuAl-A2 | — | — | — | — |
| **Nominal Comp. Test Castings** | | | | | | | | | | |
| Copper, % | 60 | 60 | 61 | 63 | 58 | 58 | 61 | 63 | 58 | 58 |
| Iron, % | 1.0 | 1.0 | 1.0 | 1.0 | 1.5 | 1.5 | 1.0 | 1.0 | 1.5 | 1.5 |
| Manganese, % | 3.0 | 3.0 | 3.0 | 2.0 | 3.5 | 3.5 | 3.0 | 2.0 | 3.5 | 3.5 |
| Aluminum, % | 3.0 | 3.0 | 3.0 | 2.0 | 3.0 | 3.0 | 3.0 | 3.0 | 5.0 | 5.0 |
| Zinc, % | 33 | 33 | 33 | 32 | 34 | 34 | 32 | 31 | 32 | 33 |
| **Nominal Comp. Filler Metal** | | | | | | | | | | |
| Copper, % | — | 60 | 56 | 60 | 56 | 88 | 88.7 | 78 | 60 | 75 |
| Iron, % | — | 1.0 | 0.6 | 1.0 | 0.6 | 1.5 | 2.3 | 3.7 | 1.0 | 3 |
| Manganese, % | — | 3.0 | 0.25 | 3.0 | 0.25 | — | — | — | 3.0 | 12 |
| Aluminum, % | — | 3.0 | — | 3.0 | — | 10 | 9.0 | 10.8 | 3.0 | 8 |
| Zinc, % | — | 33 | 41 | 33 | 41 | — | — | — | 33 | — |
| Tin, % | — | — | 0.75 | — | 0.75 | — | — | — | — | — |
| Nickel, % | — | — | — | — | — | — | — | 4.0 | — | 2 |

TABLE 2 — MECHANICAL TEST RESULTS ON WELDED HIGH-STRENGTH YELLOW BRASS CASTINGS

| Welding Process | Control Casting | Oxyacetylene | | | Shielded Arc | | | | | |
|---|---|---|---|---|---|---|---|---|---|---|
| Filler Metal | | Manganese Bronze (8B) | Low Fuming Manganese Bronze | Manganese Bronze (8B) | Low Fuming Manganese Bronze | Aluminum Bronze | Aluminum Bronze (9A) | Aluminum Bronze (9D) | Manganese Bronze (8B) | High Manganese Aluminum Bronze |
| A.W.S.-A.S.T.M. Classification | — | — | RCuZn-C | — | RCuZn-C | RCuAl-A2 | — | — | — | — |
| Flux | — | Borax | Borax | Borax | — | — | — | — | — | — |
| Stress Relief | Yes | Yes | Yes | Yes | Yes | Yes | No | No | Yes | No |
| Tensile Strength, psi | 93,300 | 26,900 | 34,300 | 42,400 | 47,400 | 68,500 | 74,000 | 76,200 | 70,800 | 90,000 |
| Yield Strength, psi | 50,000 | 24,100 | 18,200 | 34,600 | 23,000 | 52,000 | 30,000 | 35,000 | 27,000 | 47,000 |
| Elong., % | 18.7 | 4.5 | 10.1 | 4.3 | 10.8 | 3.7 | 12.0 | 8.5 | 16.7 | 6.7 |
| Brinell hardness, 3000 Kg. Load | 185 | — | — | — | — | — | — | — | — | — |
| Fracture | — | Weld | Weld | Weld | Weld | Weld | Weld | Weld | Weld | Outside Weld |
| Root Bend, degrees | 180 | 17.5 | 180 | 55 | 139 | 9.5 | 43 | 25 | 48 | 72 |
| Face Bend, degrees | 180 | 16 | 180 | 21 | 111 | 23 | 87 | 30 | 88 | 81 |
| Fracture | — | Weld | — | Weld | Weld | Weld | Weld | Weld | Weld | Outside Weld |

A conventional tungsten nonconsumable electrode, helium atmosphere and 150 ampere d.c. current (straight polarity) were employed during welding along with approximately ¼-in. thick uncoated filler rods. The castings were preheated to a temperature of about 750 F (400 C) and wire brushed between each pass to remove any deposited zinc oxide resulting from volatilization of zinc from the base metal. Four passes of 3 min duration each were required to complete each weld. The castings welded with the low fuming manganese bronze and aluminum bronze RCuAl-A2 filler metals were stress relieved at 1022 F (550 C) whereas the remaining 4 castings were tested as-cast.

## WELDED TEST CASTINGS EVALUATION

Transverse tensile and guided bend test specimens of the dimensions, shown in Fig. 3, were removed from both the welded and nonwelded test castings. The results of these tests are shown in Table 2. Compared to the properties of the nonwelded high strength yellow brass control casting, the mechanical characteristics of castings oxyacetylene gas welded with manganese bronze 8B alloy and low fuming manganese bronze alloy fillers were inferior. In the case of the manganese bronze 8B filler, the appearance of the test bars before and after testing indicated that excessive porosity caused by zinc volatilization from the casting and filler contributed to the poor properties. All test bar fractures occurred through the welds.

Inert-gas-shielded-arc welding with the several different fillers resulted in test bar values that in some instances approached those of the high strength yellow brass control casting. Of particular note are the results on the casting welded with the high manganese aluminum bronze alloy. Fracture occurred outside the weld at a load near the typical ultimate strength of the high strength yellow brass parent metal. The rather low 6.7 per cent elongation exhibited by the parent metal is attributed to the somewhat higher than desired aluminum content of the test casting (Table 1).

High ductility (16.7 per cent) was obtained from the casting welded with parent metal composition (manganese bronze 8B alloy) but tensile properties were lower than expected. This circumstance is believed to have been caused by a loss of zinc during welding. The other welded castings, except that welded with aluminum bronze RCuAl-A2 filler, showed good combinations of strength and ductility, but not of the level desired for the high strength yellow brasses.

Originally it was intended to stress relieve all castings after welding, however, the ductility of the casting welded with the RCuAl-A2 filler was so unusually low it was concluded that the stress relief treatment had resulted in an undesirable eutectoid transformation in the weld desposit causing the loss in ductility. For this reason, the remaining castings welded with aluminum bronze fillers were not stress relieved following welding.

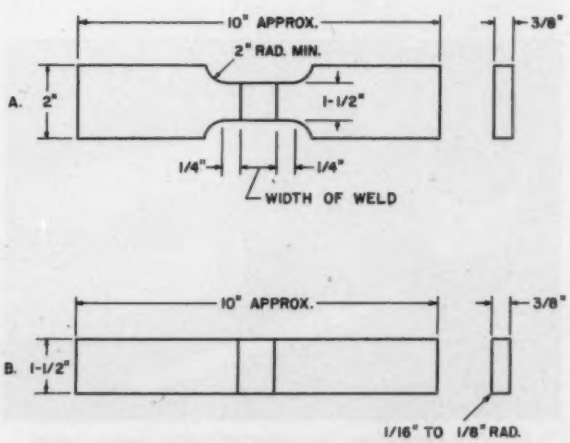



Fig. 3 — Transverse tension (A) and guided bend (B) test specimens.

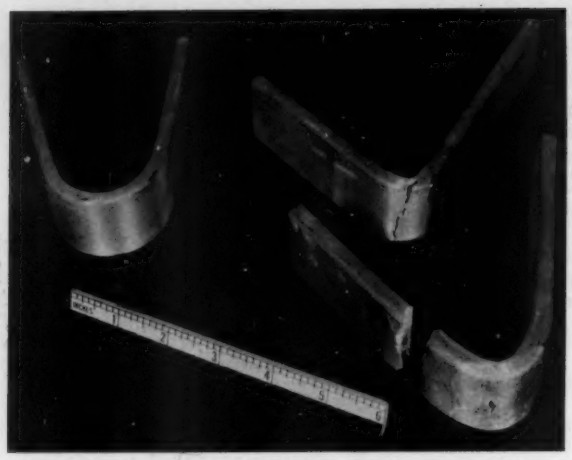


Fig. 4 — Bend tests on high strength yellow brass specimens as welded with low fuming manganese bronze filler rod (A.W.S.-A.S.T.M. RCuZn-C).

### *Bend Test Specimens*

Guided bend test specimens were placed in a die-plunger jig, and bent until the specimens conformed to a U-shape or cracks appeared on the convex surfaces. One face-bend and one root-bend test was performed on each of the test castings. The face-bend samples were placed in the die member with the faces of the welds toward the die gap, whereas the root-bend samples were placed with the roots of the welds toward the gap. As shown in Table 2, aside from the casting oxyacetylene gas welded with low fuming manganese bronze, none of the welded test specimens were able to sustain 180 degree bends as had the nonwelded control casting.

A considerable amount of ductility in bending was revealed by most of the inert-gas-shielded-arc welded specimens, especially those welded with low fuming manganese bronze and high manganese aluminum bronze. In the latter instance, cracks initiated in the parent metal rather than in the weld deposit. Figures 4 and 5 show several of the bend test specimens compared to a sample from the control casting.

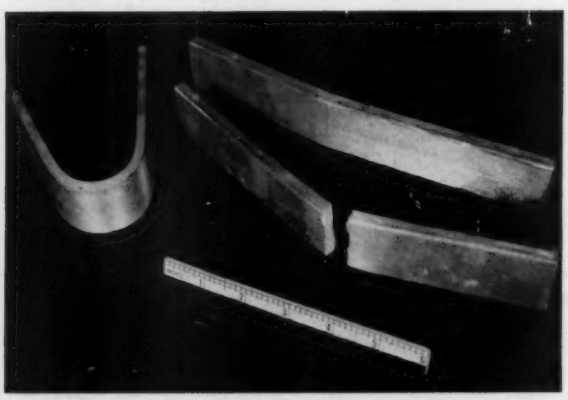


Fig. 5 — Bend tests on high strength yellow brass specimens arc welded with aluminum bronze filler rod (A.W.S.-A.S.T.M. RCuA1-A2).

To determine if there were variations in hardness in the weld deposits or a brittle fusion zone at the weld-parent metal interfaces, microhardness tests were conducted on sections of the inert-gas-shielded-arc welded castings normal to the direction of welding. Before testing, the samples were etched with a solution of potassium dichromate to reveal the structure of both the weld and parent metal. Figure 6 shows one of the samples and location of Vickers microhardness impressions taken every 0.1 in. across and down the weld.

The numbers below the photograph correspond to the transverse hardness values on the several welded castings. There was no evidence of any brittleness in the welds or parent metal adjacent the welds. The somewhat lower hardness in the manganese bronze 8B filler deposit supports the previously stated opinion that some zinc was lost during welding. Spectrochemical analysis of the deposit also showed zinc to be below that in the original filler rod. The highest weld hardnesses were obtained with the aluminum bronze 9D alloy and the high manganese aluminum bronze alloy, whereas the lowest hardnesses were realized with the low fuming manganese bronze filler.

## METALLOGRAPHIC EXAMINATION

Figures 7 through 12 show portions of the weld-base metal interfaces in the several inert-gas-shielded-arc welded castings at 100 ×. The base metal is shown at the top portion of each photograph, all consists of an all beta solid solution matrix through which is dispersed a star-like iron-rich compound. Variations in the shading of grains is the result of preferential attack by the etching solution. Figure 7 reveals the duplex alpha plus beta structure of the low fuming manganese bronze filler. Although somewhat difficult to discern, the alpha constituent appears plate-like, whereas the beta constituent is a continuous matrix. The weld is sound and well fused to the base metal. Alloying appears to have been limited to about one grain diameter.

Figure 8 indicates that stress relieving welded high strength yellow brasses at a temperature of 1022 F (550 C) can cause an appreciable amount of brittle eutectoid to form in aluminum bronze weld deposits. This circumstance, as stated previously, is believed to have caused the low ductility in the tensile and bend specimens of the sample shown. The white, needle-shaped constituent in the weld is the copper-aluminum solid solution alpha phase, whereas the dark constituent is finely divided eutectoid (alpha plus gamma phase).

Figure 9 is interesting in that a dark continuous band is readily observed between the base metal and weld deposit. The nature of this fusion zone is similar to the eutectoid viewed in the previous photograph. The remainder of the weld consists essentially of a copper-aluminum alpha matrix with small patches of eutectoid. A narrow portion of the heat affected zone adjacent the eutectoid band is duplex alpha plus beta which fuses within about two grain diameters into the all beta matrix. Figure 10 shows a portion of the weld-base metal interface of



Fig. 6 — Etched cross-section of welded specimen with hardness indentations (see table).

HARDNESS MEASUREMENTS ON WELD DEPOSIT AND PARENT METAL

Bhn 3000 Kg. converted from Dph (see Fig. 6).

| Low Fuming Manganese Bronze (RCuZn-C) | Aluminum Bronze (RCuAl-A2) | Aluminum Bronze (9A) |
|---|---|---|
| 175 | 175 | 182 |
| 177 | 178 | 178 |
| 165_____ | 179_____ | 175_____ |
| 83 | 137 | 138 |
| 84 | 138 | 137 |
| 69 Weld | 138 Weld | 131 Weld |
| 79 Metal | 138 Metal | 135 Metal |
| 74 | 135 | 135 |
| 79 | 135 | 137 |
| 84_____ | 135_____ | 140_____ |
| 172 | 175 | 167 |
| 173 | 165 | 172 |
| 175 | 165 | 175 |
| **Aluminum Bronze (9D)** | **Manganese Bronze (8B)** | **High Manganese Aluminum Bronze** |
| 175 | 173 | 173 |
| 175 | 173 | 165 |
| 178_____ | 169_____ | 162_____ |
| 194 | 148 | 173 |
| 199 | 148 | 175 |
| 195 Weld | 148 Weld | 173 Weld |
| 193 Metal | 151 Metal | 175 Metal |
| 194 | 145 | 184 |
| 192 | 153 | 175 |
| 190_____ | 155_____ | 175_____ |
| 175 | 178 | 167 |
| 184 | 173 | 163 |
| 184 | 173 | 167 |

the casting welded with aluminum bronze 9D alloy. In this case, the weld consists primarily of small copper-aluminum beta grains surrounded by an alpha constituent. At the point of fusion, the phases present in the heat affected zone are equiaxed beta and alpha plus beta.

The manganese bronze weld deposit, shown in Fig. 11, consists of a mixture of small copper-zinc alpha plus beta phases. The loss of zinc during welding and illustrated by the low hardness values and low tensile and yield strengths of the test bars undoubtedly accounts for the lack of a complete beta structure in the weld. A rather sharp fusion line defines the weld-base metal interface. Figure 12 presents a photomicrograph of the weld consisting of the high manganese aluminum bronze alloy. Fusion is quite extensive, and several different structures can be observed in the heat affected zone. The high manganese aluminum bronze consists primarily of copper-aluminum alpha phase with some dispersed beta.

## WELDING PRODUCTION CASTINGS

As a further step in this experimental program, several castings from a production lot of high strength yellow brass were purposely slotted and routed, as shown in Fig. 13. Utilizing the same techniques employed on the plate-shaped castings, these production castings were inert-gas-shielded-arc welded with either base metal composition filler rod (manganese bronze 8B alloy) or high manganese aluminum bronze alloy filler. Figure 14 shows two of these castings after welding. The casting to the left in the photograph has been welded with base metal composition filler.

Some zinc splatter can be observed as a white coating on the weld and part of the casting. The casting to the right has been welded with high man-

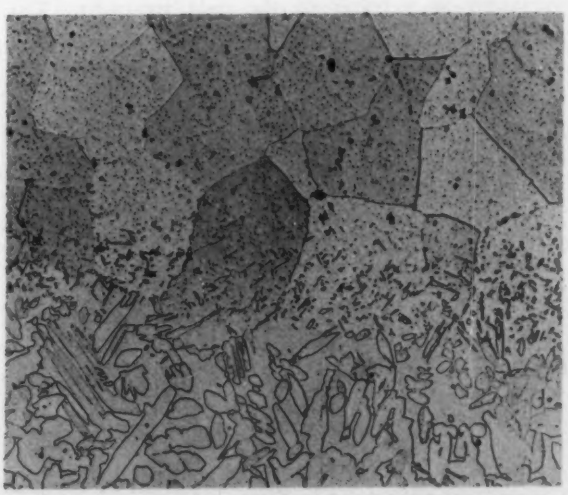


Fig. 7 — Arc welded high strength yellow brass. *Top* — base metal; *bottom* — low fuming manganese bronze filler (RCuZn-C). $K_2Cr_2O_7$ etch. 100 ×.

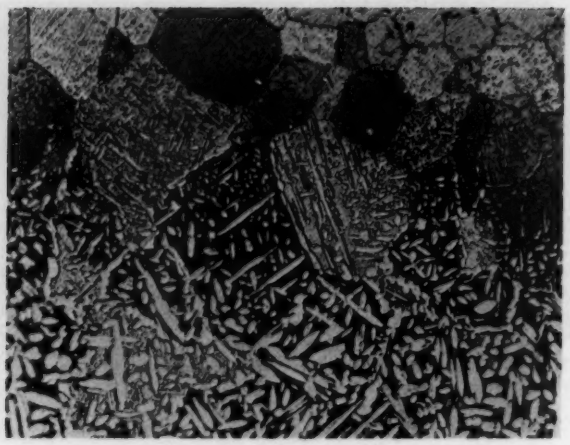


Fig. 8 — Arc welded high strength yellow brass. *Top* — base metal; *bottom* — aluminum bronze (RCuA1-A2). $K_2Cr_2O_7$ etch. 100 ×.

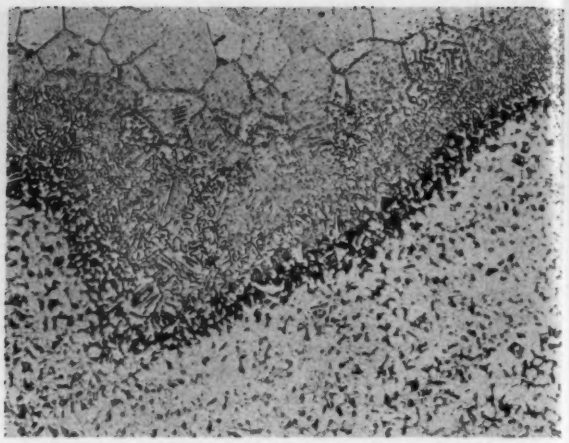


Fig. 9 — Arc welded high strength yellow brass. *Top* — base metal; *bottom* — aluminum (9A). $K_2Cr_2O_7$ etch. 100 ×.

ganese aluminum bronze filler. A smaller amount of zinc was volatilized and the bead appeared cleaner and more uniform than its counterpart in the casting to the left.

The welded high strength yellow brass castings are shown again in Fig. 15 (position has been switched from Fig. 14) after removing the weld flash from the tongue portions of the castings. The color match was good in the casting welded with the base metal composition filler. The color match in the casting welded with high manganese aluminum bronze was not quite as good because of the silvery appearance of the weld deposit. Fusion was good in both samples, and both the welds and base metals showed complete absence of surface porosity and other discontinuities.

To investigate the effect of stress relief treatment on the welded castings, several of them were heated to a temperature of 102 F (550 C) for one hr and furnace cooled. Brinell hardness values on these castings before and after stress relieving showed little variation as a result of the heat treatment. The high manganese aluminum bronze is reportedly not prone to a brittle eutectoid transformation at temperatures of about 1022 F (550 C).

This circumstance makes it possible to stress relieve high strength yellow brass castings welded with the high manganese aluminum bronze at the conventional stress relieving temperatures without concern for developing undesirable structures in the weld. The Brinell hardness values on the deposited manganese bronze 8B composition were on the order of about 180 (3000 kg. load). This indicates that properties close to that of the control test casting previously discussed could be obtained from the weld.

## CONCLUSION

Based on the results obtained in this investigation, it is concluded:

1. High strength yellow brass castings can be successfully repair welded using inert-gas-shielded-arc welding techniques.

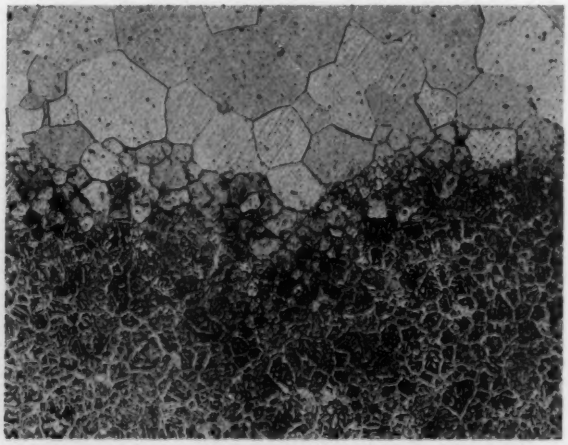


Fig. 10 — Arc welded high strength yellow brass. *Top* — base metal; *bottom* — aluminum bronze (9D). $K_2Cr_2O_7$ etch. 100 ×.

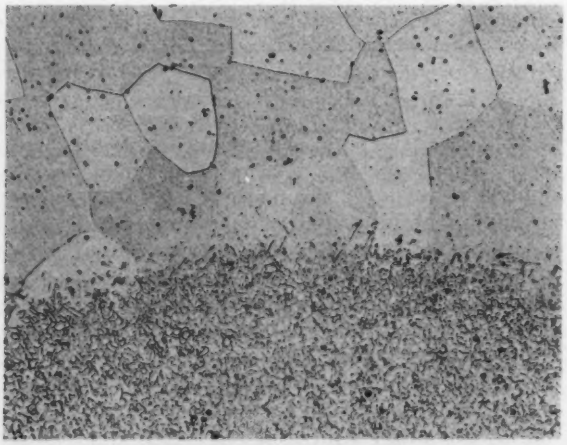


Fig. 11 — Arc welded high strength yellow brass. *Top* — base metal; *bottom* — manganese bronze (8B). $K_2Cr_2O_7$ etch. 100 ×.

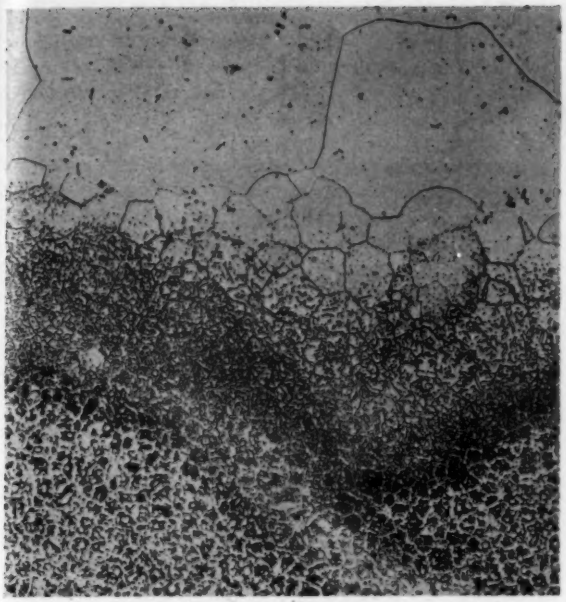


Fig. 12 — Arc welded high strength yellow brass. *Top* — base metal; *bottom* — high manganese aluminum bronze. $K_2Cr_2O_7$ etch. 100 ×.

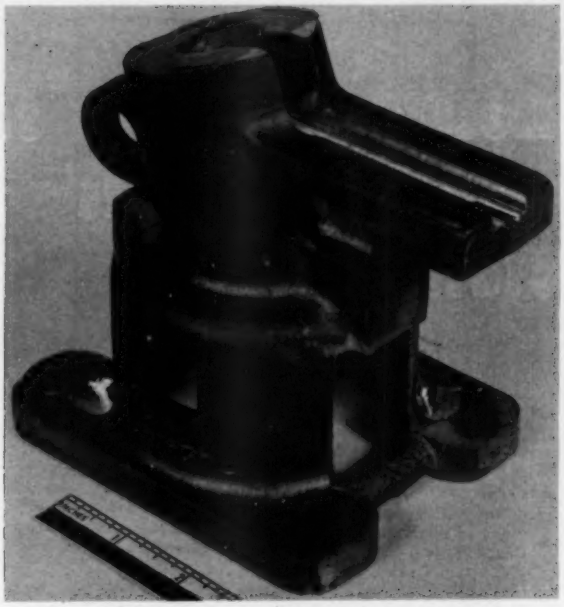


Fig. 13 — High strength yellow brass casting with machined slot and routed corners.

2. Satisfactory high strength, high ductility welds can be obtained with both high manganese aluminum bronze filler metal as well as manganese bronze 8B alloy. A greater amount of variation in the properties of welds made with the latter material would be expected as a result of the zinc volatility problem.

3. Color match is better in high-strength yellow brasses welded with the manganese bronze 8B alloy than those welded with high manganese aluminum bronze. Where color match is not important, however, either material can be used.

4. High strength yellow brasses repair welded with either of the two filler metals can and should be stress relieved at a temperature of about 1022 F (550 C) to minimize residual stresses caused by welding.

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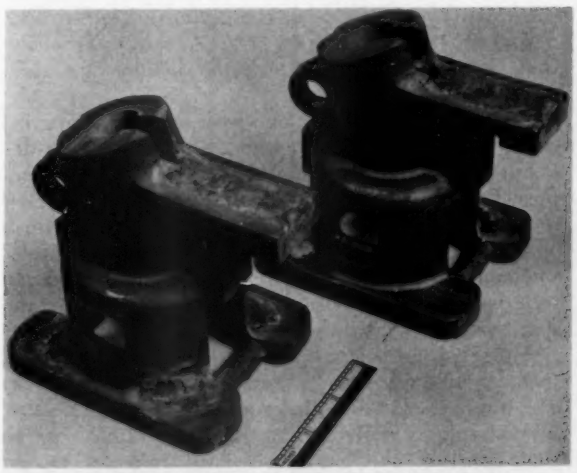



Fig. 14 — Welded high strength yellow brass castings.

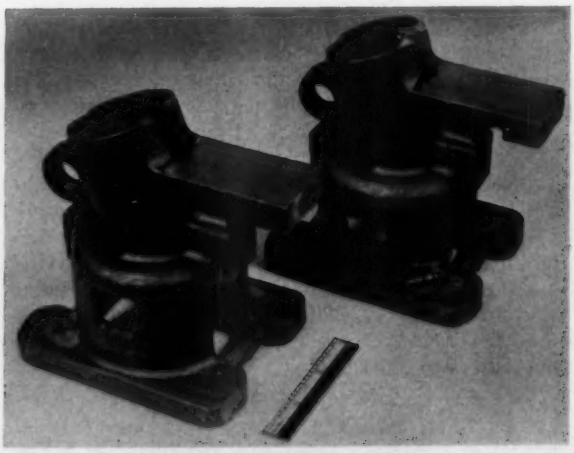


Fig. 15 — Welded high strength yellow brass castings after removal of weld flash from tongues.

# MOLDING AND CORE SAND HIGH TEMPERATURE STRENGTH TESTING

## *A New Approach*

by R. W. Heine, J. S. Schumacher and E. H. King


### ABSTRACT

**A method of testing is described using the dilatometer which shows how compressive strength changes with time of exposure to elevated temperatures. A hollow cylindrical sand specimen is used that has a heating rate equivalent to that obtained in sand layers of similar thickness adjacent to the mold-metal interface. The strength vs. time of exposure curves for molding and core sand is shown and related to their behavior in casting. The variables of clay content and type and moisture content in molding sand are investigated by this method. Comparison is made of the behavior of new and used sand mixtures. The testing method provides a new tool for revealing the behavior of sands during casting of the metal.**


### INTRODUCTION

A method of studying the change of the compressive strength property of a molding or core sand during heating such as occurs when molten metal is cast against the sand is offered. The elevated temperature properties of foundry sands have been studied under conditions approaching equilibrium where a sample is heated to a certain temperature, held long enough at that temperature to approach equilibrium and then tested. Most of the past studies of high temperature sand properties reported in the literature have been of this type. A reference[1] presents the results of this type of a study by the authors, and other references[2-16] report the work of others who have also used this method.

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Objection to this method of testing has been voiced by a number of researchers.[17,18] It is frequently stated that a method of testing should be used that involves rapid heating of the sand sample at rates such as occur in the sand when molten metal is poured into a mold. Such a testing method is presented.

### TESTING METHOD

The sample employed is a hollow cylinder 2.0 in. long with 1⅛-in. O.D. and 0.50 in. I.D. Wall thickness of the sand is thus $\frac{5}{16}$-in. It is double-end rammed three times using a rammer of the type, shown on p. 108 (Fig. 47) of the AFS FOUNDRY SAND HANDBOOK. The specimen tube is shown schematically in Fig. 1. Figure 1 shows a 0.500 in. D. steel rod supported vertically by pin *D* and two centering steel discs *A* and *B*. Disc *A* and pin *D* are placed in position first. The rod and the specimen tube are then positioned.

Sand is then charged into the annular space between the rod and the tube and uniformly distributed. Centering disc *B* is then positioned and pushed down onto the sand until it binds. Pin *D* is removed and rod C is pushed downward until flush with the top of disc *B*. The post is then placed under the rammer and rammed three times. The specimen tube is inverted and rod *C* is withdrawn before the specimen is stripped from the tube.

The sample is then placed in the dilatometer furnace held at some preselected temperature. An own-atmosphere hood shrouds the sample. The hollow sample is seated on a vented ceramic disc so that moisture and gases may pass from the heated O.D. into the hollow center and cut through the

vented disc. This simulates the movement of gases away from the mold-metal interface toward the interior of the mold (or core). The vented disc is shown as part of the setup in Fig. 3.

### *Compressive Strength*

A series of the specimens is broken after increasing time intervals of exposure to the elevated temperature of the dilatometer. For example, with the furnace at 1750 F (954 C), a series of samples is tested for compressive strength after a time interval in the furnace from 20 sec to 10 min. The results permit a graph, such as Fig. 2, to be constructed relating compressive strength of the sand to time of exposure in the furnace.

The hot compressive strength and time of rupture data are plotted as in Curve *A* in Fig. 2, which

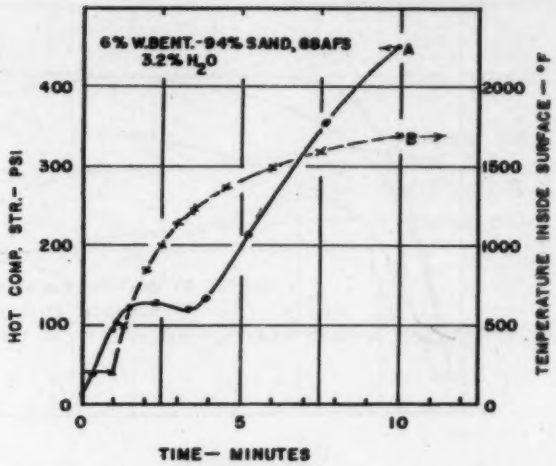



**Fig. 2 — Hot compressive strength vs. time in the furnace, and temperature at inside surface vs. time in the furnace for a furnace temperature of 1750 F (954 C).**

shows the change of compressive strength with the time of a 6 per cent W. bentonite–94 per cent silica sand mixture with 3.2 per cent moisture exposed to a furnace temperature of 1750 F (954 C). Also shown in Fig. 2 is the rise in temperature with time in the furnace of the inside surface of the hollow specimen at the midpoint of its length, curve *B*. The setup for determining the temperature-time curve is illustrated in Fig. 3.

The hot compressive strength of the sand, and the temperature at the I.D. at any moment of time after the sample is placed in the furnace, is then established in Fig. 2. For instance, 2 min after the specimen is placed in the furnace, the hot compressive

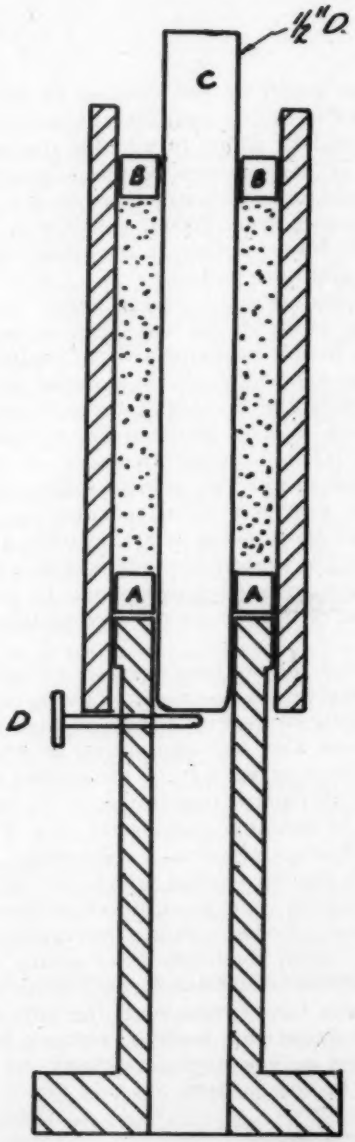



**Fig. 1 — Schematic drawing of modified ramming post and sleeve for ramming hollow specimens.**

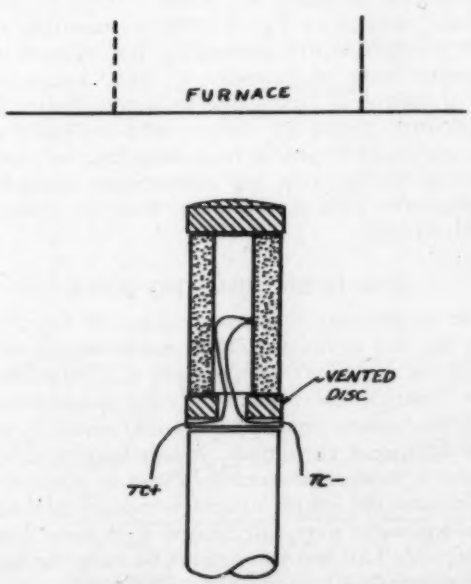



**Fig. 3 — Location of thermocouple in hollow specimen for determining temperature of inside surface as a function of time in the furnace.**

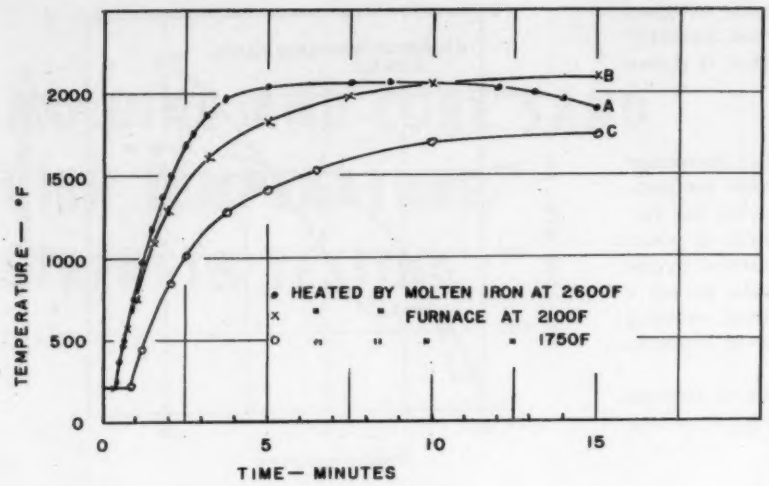



Fig. 4 — Temperature curve of inside surface of hollow specimen vs. time after casting iron around the sample. Curves of temperature vs. time after placing a sample in the furnace at a temperature of 2100 F (1149 C) and 1750 F (954 C) are also shown, for comparison.

strength of the sample is 125 psi and the temperature at the I.D. wall is 850 F (454 C). The plateau in curve *A* at one to 3 min corresponds to the dry compressive strength of the sand. Moisture is evaporated from the sample during the first one min in the furnace, as shown by the temperature curve plateau at 212 F (100 C).

Compressive strength rises from the green strength to the dry strength level during this period. The dry compressive strength level prevails until temperature at the I.D. rises above about 1200 to 1300 F (649-704 C) in about 3.5 to 4.0 min. Beyond this time and temperature, hot compressive strength develops in an increasing amount as temperature and time increase. Hot compressive strength develops mainly in the temperature range 1200 to 1800 F (649-982 C) for bentonite bonded sands, according to the equilibrium method of study.

Thus, curve *A* in Fig. 2 shows a transition from green strength to dry strength to hot strength with increasing time of exposure to high temperature. This transition is predictable from equilibrium high temperature testing of simple sand-clay-water mixture, and could be shown from these data by plotting the data of Fig. 2 as hot compressive strength vs. temperature. This is not done here, as the curve is well known.

## SPECIMEN HEATING RATE

The temperature time curve, shown in Fig. 2, provides the rate of heating of the inside surface of the sample for a furnace temperature of 1750 F (954 C) (rate is actually the slope of the curve at any instant). A higher furnace temperature would result in more rapid heating of the sample. It is important to know whether a furnace temperature could be selected that would cause the sample to heat as rapidly as it would if molten metal were cast around it in some specific casting size. This was investigated by using the hollow test specimen as a core in a green sand mold.

The casting consisted of a cylinder of iron 1⅝-in. in thickness by 4¾-in. in diameter weighing about 7 lb. The test specimen was placed vertically in the mold at the center of the cylinder of iron. This particular size of casting was used, as it was calculated that it could take about 10 min for the center of the casting, i.e., the specimen location, to cool through the final eutectic solidification temperature of gray cast iron poured from 2600 F (1426 C). A heating curve at the inside surface at the point, shown in Fig. 3, was again determined.

The temperature time curve is shown in Fig. 4. The heating curve *A*, for the specimen heated by casting it in iron, is compared with the heating curves of specimens exposed to the dilatometer furnace at temperatures of 2100 F (1149 C), curve *B*, and 1750 F (954 C), curve *C*. The specimen heating curves for the 2100 F (1149 C) furnace temperature, and the specimen cast in iron, are seen to follow closely together. The difference in temperature at a given time is small up to about 1200 F (649 C). At higher temperatures the temperature difference increases, and then decreases to a minimum about 10 min after pouring the casting or placing the sample in the furnace.

In any event, the two heating curves are sufficiently close to each other so that the two heating conditions should produce equivalent changes in high temperature properties. One may also observe in Fig. 4 that the temperature at the I.D. of the sample does not drop below the eutectic temperature of the iron until more than 10 min after pouring the iron. These experiments demonstrate that a dilatometer furnace temperature can be selected which will produce a specimen heating rate or temperature-time curve, similar to that caused by a molten iron casting.

Obviously, many combinations of casting size and shape and dilatometer furnace temperature could be studied in this way. Furthermore, the wall thickness of the hollow specimen could be reduced. However, these changes do not appear necessary, for reasons which will become evident.

### *Sand Heating Rates*

Other examples of heating rates of sand are available from the literature. A reference[19] provides heat-

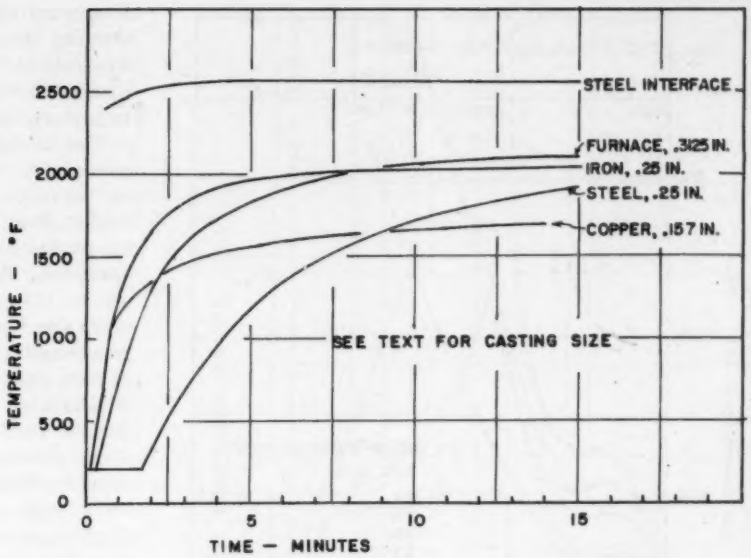



Fig. 5 — Heating curves of molding sand at the indicated distance from the mold surface when steel, gray iron and copper are cast against the sand (see text for details). Comparison is made with heating curve *F* of the hollow sample in the dilatometer at 2100 F (1149 C).

ing curves at the mold interface and 0.25 in. in from the interface when a 7 in. x 7 in. square section of steel bar is cast in green sand. These heating curves are marked *SI* (steel interface-mold surface) and S (Steel) 0.25 in from surface in Fig. 5. Another reference[20] shows heating curves for a point 0.25 in. from the mold surface where gray iron is cast in a 7 in. x 7 in. square bar; the curve is marked I (Iron) 0.25 in. in Fig. 5.

Another reference[21] provides a heating curve for a distance 0.157 in. from the mold surface of a copper casting. This curve is marked C (Copper) 0.157 in. on Fig. 5. For comparison, Fig. 5 also shows the heating curve *F* (Furnace), for the inside surface of the hollow specimen ($\frac{5}{16}$-in. wall thickness) placed in the dilatometer at 2100 F (1149 C). These curves all demonstrate finite heating rates which are of a similar order of magnitude. They cannot be expected to be identical, but the point is made that if it were necessary a furnce temperature and hollow specimen wall thickness could be chosen to simulate the heating curve of sand caused by casting metal against it. This is an unnecessary refinement in method as subsequent sections reveal.

The difficulty of measuring sand surface temperature, both in actual molds and on the test specimen, makes it impractical to consider heating curves of these surfaces. However, they must be also similar, since the heat will only be transferred in a given time interval if the temperature differences, surface to interior point, are equivalent in magnitude.

## DILATOMETER FURNACE TEMPERATURE

The effect of dilatometer furnace temperatures of 1200, 1500, 1750 and 2250 F (649, 815, 954 and 1232 C) on the hot compressive strength-time curve is shown in Fig. 6. The sand is a mixture of 6 per cent western bentonite-94 per cent silica sand of 68 AFS, 4 screen type with 2.6 per cent $H_2O$. At 1200 F (649 C), the strength rises from the green strength level to the dry compressive strength level, but no significant increase occurs from hot strength. At 1500 F (815 C), hot strength above the dry strength level begins after the sample has been in the furnace for about 4.5 min. At 1750 F (954 C), hot strength begins to develop after the sample is in the furnace about 3.5 min.

At the latter two temperatures, hot strength of the sand is still increasing after 10 min in the furnace. At 2250 F (1232 C), hot strength begins to develop after about 1.25 min in the furnace, reaches a peak after 2.5 min and then decreases. The decreasing strength is caused by fusion of the clay at the higher

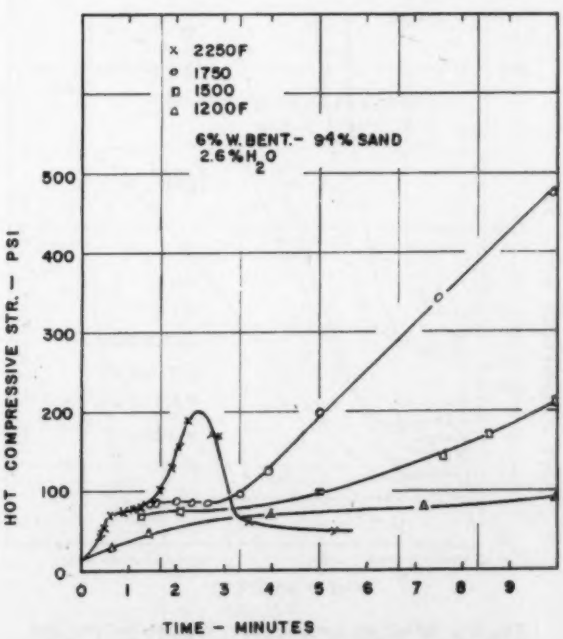



Fig. 6 — Effect of dilatometer furnace temperature on the hot compressive strength vs. time in the furnace curve for a 6 per cent western bentonite - 94 per cent 68 AFS silica sand, 2.6 per cent $H_2O$ mixture.

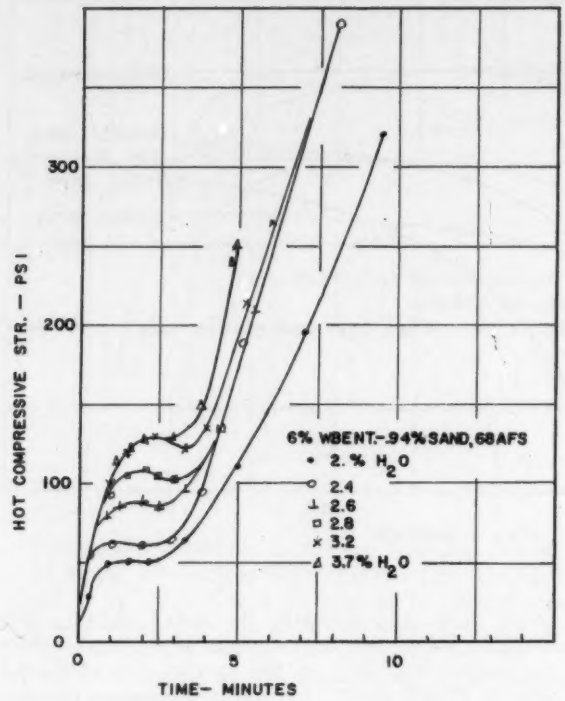



**Fig. 7 — Effect of moisture content on hot strength vs. time in the furnace curve of a 6 per cent western bentonite - 94 per cent sand mixture.**

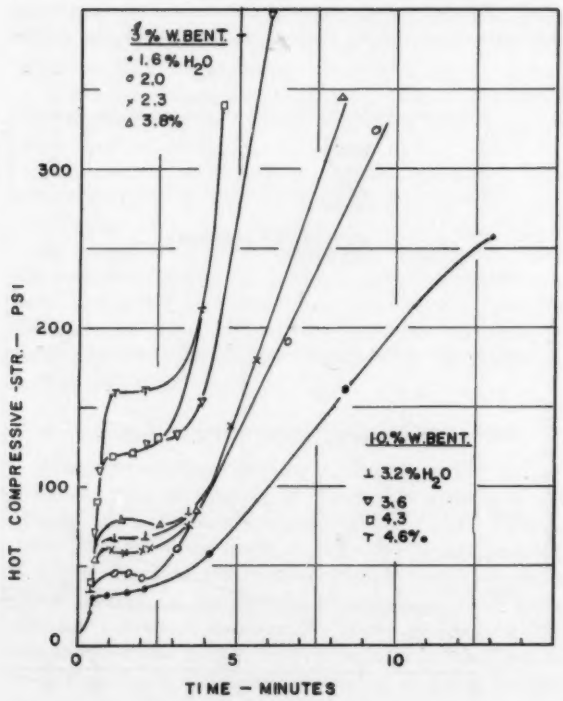



**Fig. 8 — Effect of varying clay content on the hot strength vs. time in the furnace curve. Mixtures containing 3 per cent western bentonite and 10 per cent western bentonite may be compared with 6 per cent western bentonite in Fig. 7.**

temperatures. Fusion occurs to the point where the samples no longer fracture but simply flow under load. Thus the strength-time, in the furnace curves in Fig. 6, follow relationships expected from equilibrium studies.[1]

The selection of a particular furnace temperature for heating the specimen is best made on the basis of what particular property change with time is under study, rather than on the basis of attempting to produce some particular rate of heating of the specimen. In the case of the higher temperatures, above 1750 F (954 C), a mechanical problem arises with the loading rate of the specimen. The changes in strength occur so rapidly during the first ½ min to 2 min that it becomes difficult to manipulate the dilatometer within the time required. For that reason, a furnace temperature of 1750 F (954 C) was most commonly used, except where special effects were studied at higher or lower temperatures.

Deformation of the sand could be considered here, too. However, since it follows expected behavior based on temperature, it will not be considered.

## MOISTURE AND CLAY CONTENT EFFECT

The effect of moisture variation on the hot strength — time curve for a 6 per cent western bentonite-94 per cent silica sand mixture is summarized in Fig. 7. Dry compressive strength level is, of course, expected to vary with moisture in the sand in the range indicated. Hot compressive strength after a given interval in the furnace is also higher at higher moisture contents. The effect of higher and lower clay contents, 10 and 3 per cent western bentonite, is compared with 6 per cent western bentonite in Figs. 7 and 8 again with moisture variable. The lower clay contents produce lower strength vs time curves and the higher clay content raises the curves as expected. Similar results were obtained for southern bentonite.

## CLAY TYPE EFFECT

Figure 9 compares hot strength-time curves for 6 per cent western bentonite, 6 per cent southern bentonite and 15 per cent fireclay bonded mixtures using a furnace temperature of 1750 F (954 C). The fireclay appears to behave differently than the bentonite bonded sand, since it shows no dry compressive strength level in the curve. However, this is to be expected on the basis of the equilibrium studies reported.[1]

## CORE SAND

Core sands were also studied using this method. However, the specimens were first baked and cooled to room temperature to develop the normal core oil bond. The data for one of the mixtures studies are shown plotted in Fig. 10. The mixture contained 1.75 per cent oil, 2.0 per cent cereal, 4 per cent water and 68 AFS 4 screen silica sand. As-baked, the room temperature compression strength was 1000 psi. Furnace temperatures of 1200 F (649 C) and 1750 F (954 C) were used. Hot deformation and compressive strength were measured. Figure 10 shows rapid decrease of strength to a less than 25 psi in less than 2 min in the furnace.

The core specimens also show a large increase in deformation at the time of the minimum strength. This is due to softening of the core oil bond and distillation of the volatile fraction during this time interval. Distillation may be observed, and gases coming from the sample burst into flame on coming into contact with air at the exit to the own atmosphere shroud after the core specimen is in the furnace for 17 sec at a furnace temperature of 1750 F (954 C). The minimum strength of 22 psi at 2 min is followed by a rise due to coking of the core oil bond when distillation ceases.

The end of distillation and the coking bond causes hot deformation to decrease from its maximum. With continued time in the furnace at 1200 F (649 C), the strength drops to zero psi due to oxidation of the carbon coke bond. However, at 1750 F (954 C) hot strength is maintained in this mixture from the ash of the oil and cereal and the approximately 0.5 per cent AFS clay content of the base sand. These results illustrate the importance of selecting proper furnace temperatures. At 1200 F (649 C), the coke bond is destroyed by oxidation but no high temperature fusion bond is developed. The same results are obtained at 1000 F (539 C), but the time is extended.

At 1750 F (954 C), a fusion bond develops from fusible inorganic materials in the sand. Lower percentages of oil and cereal lower the strength level of the coke bond as would be expected and also the fusion bond. The authors believe these results explain the behavior of cores and core sand molds in comparison with molding sands during casting operations. This is especially true if the hot strength-time curve of molding sands containing carbonaceous materials is compared with the curves for core sands.

## FOUNDRY MOLDING SAND

The hot strength-time curves for two foundry sands used for green sand molding are shown in Fig. 11. The analysis of these sands is given in the table. Many other sands were tested. The two listed in the table simply illustrate a lower and higher level of hot strength vs. time curve for a furnace temperature of 1750 F (954 C). It should be remembered that these curves are for specimens rammed to about 90-92 mold hardness. The principle points to be observed from these curves are that they behave the same as the new sand mixtures tested, and that the lower strength-time curve is not as high as that of some core sand mixtures once the dry strength level is reached.

Core sands are, of course, easier to ram to full density than are molding sands under foundry conditions. The lower curve in Fig. 11 shows a hot strength plateau for sand *B* at 34 psi. But this is at 92 mold hardness. Dry compressive strength in the table is shown to be 41 psi. This sand appears to have too low a strength level if it is to be used in molds at the 80 mold hardness level. It might be expected to cut or wash during pouring and to show dirt in the casting. The low strength level is due to the cellulose content of the sand.

This example shows how a sand may have a lower hot strength plateau (called hot-dry compressive strength) than might be expected from the commonly determined dry compressive strength test. This effect may be caused by excessive carbonaceous material in the sand for the percentage of clay present.

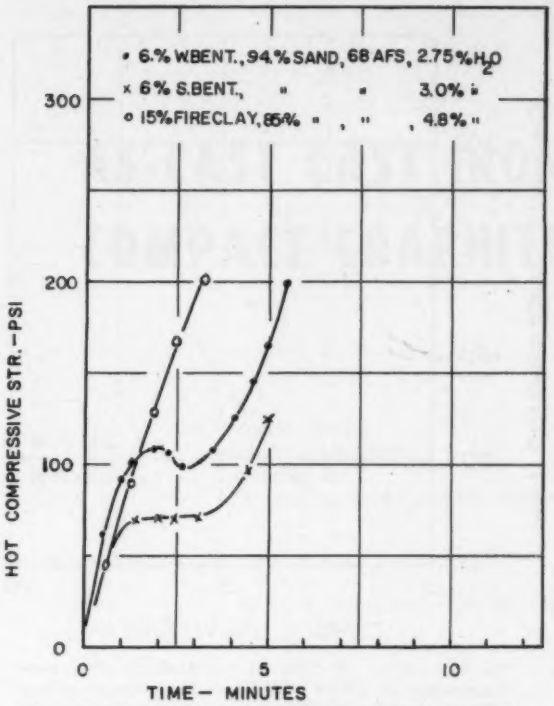



Fig. 9 — Effect of clay type on shape of hot strength vs. time in the furnace curve for sands containing 6 per cent western bentonite, 6 per cent southern bentonite and 15 per cent fireclay, respectively.

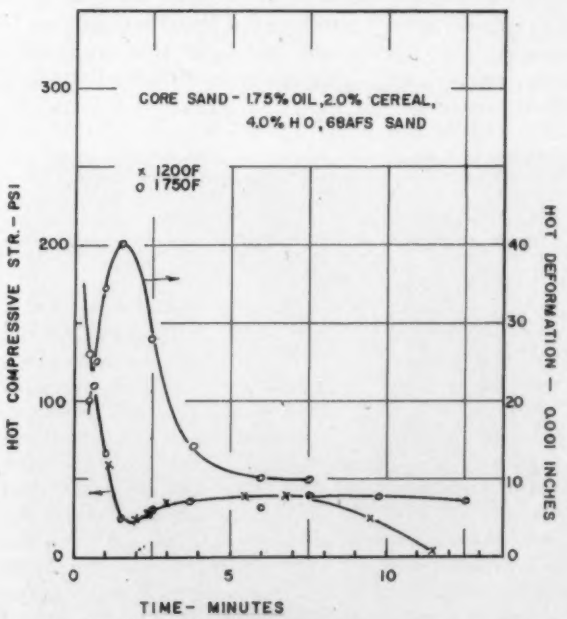



Fig. 10 — Effect of time of exposure to a furnace temperature of 1200 F (649 C) and 1750 F (954 C) on hot strength and deformation of hollow cores and specimens. Baked strength of the hollow cores at room temperature was 1000 psi.

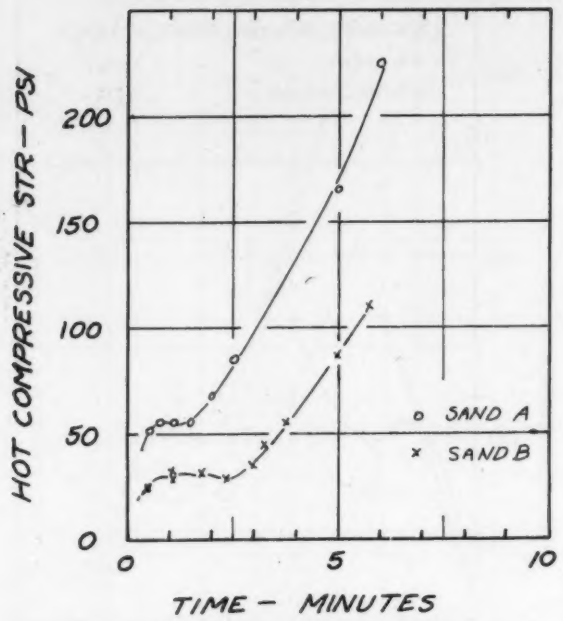



**Fig. 11 — Effect of time of exposure to a furnace temperature of 1750 F (954 C) on hot strength of two different foundry molding sands whose analyses are given in the table.**

**ANALYSIS OF TWO FOUNDRY SANDS USED FOR HOT STRENGTH-TIME TESTS**

| | Sand A | Sand B |
|---|---|---|
| AFS no. | 63 | 65 |
| AFS Clay, % | 10.6 | 14.6 |
| Adjusted Clay, % | 6.45 | 9.45 |
| Effective Clay, % | 6.0 | 6.0 |
| $H_2O$, % | 4.0 | 4.8 |
| GCS, psi | 14.0 | 13.8 |
| DCS, psi | 60.5 | 41.0 |
| 1750 F (954 C), volatile matter, % | 3.71 | 4.56 |
| Fixed Carbon, % | 4.39 | 4.18 |
| Total Carbon, % | 8.10 | 8.74 |
| Cellulose in Sand, % | 0.56 | 1.65 |
| Sea coal | balance | balance |
| Bond | W. Bent. | W. + S. Bent. |
| Metal Poured | Gray iron | Gray iron |

The three ram specimens used in this work had a 90-92 mold hardness. The effect of ramming on the hot strength-heating time curve of molding sands was studied. The results will not be considered in detail in this paper. It was found generally, however, that hot strength in specimens of 70 mold hardness is about 10 to 25 per cent of that at 90-92 mold hardness, being closer to 10 per cent when the sand was on the dry side and to 25 per cent when the sand was on the wet side. At 80 mold hardness, the hot strength was 25 to 50 per cent of the value at 90 to 92 mold hardness, moisture again being a major factor in the range.

## SUMMARY

A method of testing and a means of graphical presentation of the hot strength-time and temperature relationships in molding and core sands has been described. The hollow specimen used for testing has been demonstrated to have heating rates in the dilatometer, which are much like those encountered in the first 0.25 in. sand layer below the surface of molds cast with steel or gray iron. Molding sand and core sand mixtures have been shown to develop similar hot-dry compressive strength properties, and may therefore function equally well in producing certain castings. The testing method may be used to show the effects of carbonaceous additives, various types of bonding agents and for core and molding sands, although the main purpose of this paper has been to describe the testing procedure and a few of the useful principles uncovered.

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# AS-CAST CAST IRON COMPACT GRAPHITE

Progress Report
Malleable Division
Sponsored By
AFS Training and Research Institute

by R. W. Heine and T. W. Mueller

## ABSTRACT

**This report presents a brief summary of work at the University of Wisconsin, sponsored by the AFS Training and Research Institute under the direction of the AFS Malleable Division, with attempting to produce compact graphite in the as-cast structure of cast irons. The work was aimed at studying the effect of additions to the melt and variations in base composition of the iron.**

## PROCEDURE

Melting practices employing an induction furnace were used, as in previous work, in preparing 100 lb heats of iron.[1,2] After melting down, heating to 2800 F (1538 C), and holding, the iron was allowed to cool to 2750 F (1510 C) with power off. The addition was made by plunging the material wrapped in sheet steel into the molten iron. Pouring into molds was then done at 2700 F (1482 C). The 100 lb heats listed in Table 1 were made in this way. Nine lb heats were made in the same way in a small induction furnace. From these heats, 1⅞-in. D bars could be poured. Since these bars show the same fracture as the 2 x 2 x 8 in. bar, no change in interpretation is necessary. Heats made in the small induction furnace are listed in Table 2.

## FRACTURE

The terms gray, mottled and white fracture are well understood. However, in this work two additional fracture types were obtained that need to be explained. One of lesser importance is the white rim-gray center fracture, shown in Fig. 1. The other is a fracture which is named compact-graphite mottle fracture. The appearance is that of a white fracture, but close examination reveals many fine mottle spots of pin-point size. This fracture is shown in Fig. 2. Metallographic examination of the iron reveals that the mottle is in the form of compact graphite, such as shown in Figs. 3 and 4. The fracture is similar, or perhaps the same, as that of ductile iron in which the eutectic has failed to graphitize during freezing. This fracture will be given the symbol CW, compact graphite mottle-white fracture. The fracture results

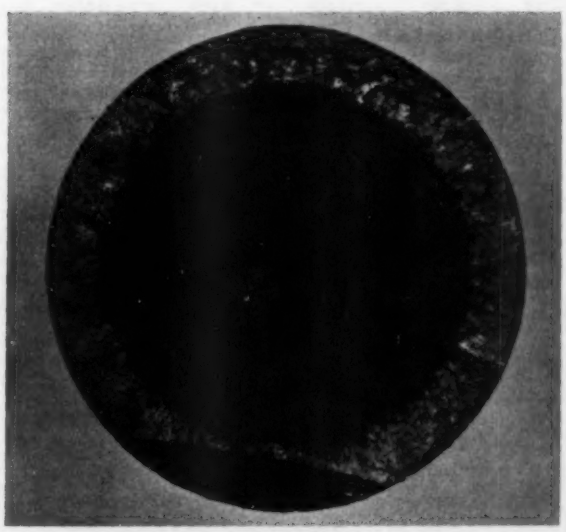


**Fig. 1 — Specimen M147. White iron, gray center fracture. 1⅞-in. bar.**

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**TABLE 1 — HEATS POURED FOR STUDY OF COMPACT GRAPHITE FORMATION (100 LB HEATS)**

| Heat No. | Analysis, % C | Si | S Orig. | S Final | Addition | Fracture 2 x 2 x 8 | Fracture 3 x 3 x 8 | Microstructure (Graphite Form) |
|---|---|---|---|---|---|---|---|---|
| M 9 | 2.54* | 2.02* | 0.050 | 0.050 | 0.1 Te | M | M | F, LC |
| M 10 | 2.50 | 2.02 | 0.050 | 0.050 | 0.01 Te | M | G | F, LC |
| M 11 | 2.50 | 2.00 | 0.050 | 0.050 | 0.2 Te | CW | CW | SC |
| M 15 | 2.32 | 1.05 | 0.051 | 0.035 | None | M | M | SC |
| M 18 | 2.32 | 1.05 | 0.051 | 0.040 | None | W | W | LC, SC |
| M 19 | 4.30 | 0.20 | 0.043 | 0.043 | None | IC | IC | SC |
| M 20 | 4.30 | 0.75 | 0.059 | 0.059 | None | G | G | F |
| M 21 | 4.30 | 0.31 | 0.046 | 0.046 | None | IC | IC | Not Observed |
| M 22 | 4.30 | 0.42 | 0.049 | 0.049 | None | IC | IC | Not Observed |
| M 23 | 2.32 | 1.05 | 0.051 | 0.035 | None | M | M | SC |
| M 24 | 2.38* | 0.98* | 0.051 | 0.025* | None | M | M | Not Observed |
| M 25 | 3.90 | 1.70 | 0.044 | 0.044 | 0.1 Te | M | M | F, LC |
| M 26 | 3.60 | 1.70 | 0.044 | 0.044 | 0.1 Te | M | M | LC, SC |
| M 27 | 3.90 | 1.70 | 0.044 | 0.044 | 0.01 Te | M | G | LC |
| M 28 | 3.70* | 1.70 | 0.044 | 0.030* | None | G | G | Type A Flake |
| M 29 | 3.90 | 1.70 | 0.044 | 0.025 | 0.01 Te | G | G | LC, F |
| M 30 | 3.94* | 1.70 | 0.044 | 0.020* | 0.02 Te | IC | G | S, SC |
| M 31 | 3.90 | 1.70 | 0.044 | 0.020 | 0.01 Te; 0.02 B | G | G | F |
| M 32 | 4.00 | 1.30 | 0.044 | 0.020 | 0.01 B | G | G | F |
| M 33 | 4.00 | 1.30 | 0.044 | 0.020 | 0.02 B | G | G | F |
| M 34 | 4.03 | 1.70 | 0.044 | 0.020 | 0.04 Te | IC | IC | S, SC, F |
| M 35 | 2.68 | 1.80 | 0.050 | 0.020 | 0.04 Te | M | M | S, SC |
| M 36 | 4.08 | 1.55 | 0.044 | 0.020 | 0.04 Te; 0.004 Mischmetal | M | IC | S, SC, F |
| M 37 | 2.68 | 1.85 | 0.050 | 0.020 | 0.04 Te; 0.02 B | M | M | SC |
| M 38 | 2.68 | 1.85 | 0.050 | 0.020 | 0.04 Te | M | M | SC |
| M 39 | 2.68 | 1.85 | 0.050 | 0.020 | 0.04 Te; 0.02 B | M | M | SC |
| M 40 | 2.68 | 1.85 | 0.050 | 0.018* | 0.04 Te; 0.02 B | M | M | LC |
| M 41 | 2.72* | 1.85* | 0.050 | 0.020* | 0.04 Te; 0.02 B | M | M | LC |
| M 42 | 2.64* | 1.83* | 0.050 | 0.020* | 0.02 Bi; 0.02 B | M | G | F, SC |
| M 43 | 2.68 | 1.70 | 0.050 | 0.015 | 0.02 B; 0.04 Bi | M | G | F, SC |
| M 44 | 2.68 | 1.70 | 0.050 | 0.015 | 0.02 B; 0.04 Te | W | W | |
| M 45 | 2.52 | 2.05 | 0.050 | 0.015 | 0.04 Bi; 0.02 B | M | G | F, LC |
| M 46 | 2.52 | 2.05 | 0.050 | 0.015 | 0.04 Te; 0.02 B | M | M | SC |
| M 47 | 2.52 | 2.05 | 0.050 | 0.015 | 0.04 Te | M | M | SC, LC |
| M 48 | 2.52 | 2.25 | 0.050 | 0.018* | 0.04 Te; 0.02 B | M | M | SC, F |
| M 49 | 2.52 | 2.25 | 0.050 | 0.014* | 0.04 Te | M | M | F, LC |
| M 52 | 2.50* | 2.35* | 0.050 | 0.018* | 0.03 Te; 0.68 Ce | CW | CW | SC, S |
| M 58 | 2.52 | 2.25 | 0.050 | 0.018* | 0.04 Te | M | M | SC, S |
| M 59 | 2.52 | 2.25 | 0.050 | 0.018 | 0.13 Ce | CW | CW | SC, F, S |
| M 60 | 2.52 | 2.25 | 0.050 | 0.018 | 0.26 Ce | CW | CW | SC, S |
| M 63 | 2.36* | 1.50* | 0.050 | 0.015* | 0.25 Ce | W | CW | SC, S |

IC — Inverse Chill
W — White
M — Mottled
G — Gray
CW — Compact Graphite Mottle, White Fracture
SC — Stubby Compact Graphite
LC — Loose Compact Graphite
S — Spherulitic Graphite
F — Flake Graphite
* — Commercial Laboratory Analysis

**TABLE 2 — HEATS POURED FOR STUDY OF COMPACT GRAPHITE FORMATION (9 LB HEATS)**

| Heat No. | Analysis, % C | Si | S Orig. | S Final | Addition | Fracture | Microstructure (Graphite Form) |
|---|---|---|---|---|---|---|---|
| M 70 | 2.70* | 1.17* | 0.071 | 0.036* | 0.15 Ce | G | F, SC, S |
| M 71 | 2.80* | 2.33* | 0.071 | 0.035* | 0.36 Ce | CW | SC, S |
| M 72 | 2.75 | 2.25 | 0.071 | 0.035 | 0.14 Ce; 0.1 Te | CW | SC, S |
| M 73 | 2.36* | 1.61* | 0.071 | | 0.21 Ce | G | F, SC |

*(continued on next page)*

TABLE 2 — HEATS POURED FOR STUDY OF COMPACT GRAPHITE FORMATION (9 LB HEATS) *(continued)*

| Heat No. | Analysis, % C | Si | S Orig. | S Final | Addition | Fracture | Microstructure (Graphite Form) |
|---|---|---|---|---|---|---|---|
| M 74 | 2.48* | 1.60* | 0.071 | | 0.40 Ce | CW | |
| M 75 | 2.45 | 1.55 | 0.071 | | 0.39 Ce (Late Si addition) | G | F, SC |
| M 76 | 2.45 | 1.55 | 0.071 | | 0.59 Ce | CW | |
| M 77 | 2.45 | 1.55 | 0.071 | | 0.84 Ce | CW | SC, S |
| M 78 | 2.43 | 2.25 | 0.071 | | 0.18 Ce | G | F |
| M 79 | 2.60* | 2.25 | 0.071 | | 0.44 Ce | CW | SC |
| M 80 | 2.54* | 2.25 | 0.071 | | 0.60 Ce | CW | SC |
| M 81 | 2.55 | 2.25 | 0.071 | | 0.78 Ce | CW | SC, S |
| M 82 | 2.50 | 1.50 | 0.071 | | 0.31 Ce | G | F, SC |
| M 83 | 2.50 | 2.25 | 0.071 | | 0.31 Ce | G | F, SC |
| M 84 | 2.50 | 2.25 | 0.071 | | 0.30 Mischmetal | G | F, SC |
| M 85 | 2.50 | 2.25 | 0.071 | | 0.48 Mischmetal | CW | SC |
| M 86 | 2.50 | 2.25 | 0.071 | | 0.31 Mischmetal | G | F, SC |
| M 87 | 2.50 | 2.25 | 0.071 | | 0.31 Mischmetal | G | F, SC |
| M 88 | 2.50 | 2.25 | 0.071 | | 0.99 Mischmetal | G | SC, F |
| M 89 | 2.50 | 2.25 | 0.071 | | 0.70 Mischmetal 0.04 Te | M | SC |
| M 90 | 2.58* | 2.25* | 0.071 | | 0.45 Mischmetal | CW | SC |
| M 91 | 2.50 | 2.25 | 0.071 | | 0.30 Mischmetal 0.01 Te | White rim | SC, F |
| M 92 | 2.50 | 2.25 | 0.071 | | 0.30 Mischmetal 0.05 Te | White rim | SC, F |
| M 93 | 2.50 | 2.25 | 0.071 | | 0.30 Mischmetal 0.1 Te | CW | SC, S |
| M 94 | 2.50 | 2.25 | 0.071 | | 0.50 Mischmetal 0.01 Te | CW | SC |
| M 95 | 2.50 | 2.25 | 0.071 | | 0.50 Mischmetal 0.05 Te | CW | SC |
| M 96 | 2.50 | 2.25 | 0.071 | | 0.50 Mischmetal 0.1 Te | CW | SC |
| M 97 | 2.50 | 2.25 | 0.071 | | 0.70 Mischmetal 0.01 Te | CW | SC |
| M 98 | 2.50 | 2.25 | 0.071 | | 0.68 Mischmetal 0.05 Te | CW | SC, S |
| M 99 | 2.50 | 2.25 | 0.071 | | 0.72 Mischmetal 0.1 Te | CW | SC |
| M 111 | 2.39 | 2.08 | 0.071 | | 0.7 Mischmetal 0.1 Te | W | |
| M 112 | 2.59 | 2.08 | 0.070 | | 0.7 Mischmetal 0.1 Te | CW | SC |
| M 114 | 2.58* | 2.05 | 0.070 | | 0.7 Mischmetal 0.1 Te | CW | SC |
| M 115 | 2.80* | 2.09* | 0.070 | | 0.7 Mischmetal 0.1 Te | CW | SC |
| M 116 | 2.99 | 2.08 | 0.070 | | 0.7 Mischmetal 0.1 Te | CW-G | SC, F |
| M 117 | 3.19 | 2.08 | 0.070 | | 0.7 Mischmetal 0.1 Te | G | SC, F |
| M 118 | 3.44 | 2.08 | 0.070 | | 0.7 Mischmetal 0.1 Te | G | F, SC |
| M 120 | 3.84 | 2.08 | 0.070 | | 0.7 Mischmetal 0.1 Te | G | F, SC |
| M 121 | 2.30 | 2.08 | 0.070 | | 0.7 Mischmetal 0.1 Te | CW | SC |
| M 122 | 2.08 | 2.08 | 0.070 | | 0.7 Mischmetal 0.1 Te | CW | SC |
| M 123 | 2.56* | 2.40* | 0.070 | | 0.7 Mischmetal 0.1 Te | CW | SC |
| M 124 | 2.38* | 2.57* | 0.070 | | 0.7 Mischmetal 0.1 Te | CW | SC |
| M 127 | 2.34* | 2.77* | 0.070 | | 0.7 Mischmetal 0.1 Te | CW | SC, F, S |
| M 128 | 2.50 | 4.00 | 0.070 | | 0.7 Mischmetal 0.1 Te | CW-G | SC, F |
| M 129 | 2.06* | 3.00 | 0.070 | | 0.7 Mischmetal 0.1 Te | CW-W | SC |
| M 130 | 2.08 | 4.00 | 0.070 | | 0.7 Mischmetal 0.1 Te | CW-G | SC, F, S |
| M 131 | 2.13 | 2.60 | 0.070 | | 0.7 Mischmetal 0.1 Te | CW-W | SC, S |
| M 132 | 2.25 | 3.00 | 0.070 | | 0.7 Mischmetal 0.1 Te | CW | SC, F, S |
| M 133 | 2.63 | 2.60 | 0.070 | | 0.7 Mischmetal 0.1 Te | CW-G | SC, F |
| M 134 | 2.50 | 3.00 | 0.070 | | 0.7 Mischmetal 0.1 Te | CW-G | SC, F |
| M 137 | 2.38* | 1.60 | 0.070 | | 0.7 Mischmetal 0.1 Te | CW | S, SC |
| M 138 | 3.20 | 1.50 | 0.070 | | 0.7 Mischmetal 0.1 Te | CW-W | SC |
| M 139 | 2.40 | 1.50* | 0.070 | | 0.7 Mischmetal 0.1 Te | CW-W | S, SC |
| M 140 | 2.80 | 1.50 | 0.070 | | 0.7 Mischmetal 0.1 Te | W | |
| M 141 | 2.50 | 1.75 | 0.070 | | 0.7 Mischmetal 0.1 Te | CW-W | SC |
| M 142 | 2.70 | 1.23 | 0.070 | | 0.7 Mischmetal 0.1 Te | W | |
| M 143 | 2.55 | 1.35 | 0.070 | | 0.7 Mischmetal 0.1 Te | CW-W | S, SC |
| M 144 | 2.85 | 1.20 | 0.070 | | 0.7 Mischmetal 0.1 Te | W | |
| M 145 | 2.70 | 1.35 | 0.070 | | 0.7 Mischmetal 0.1 Te | CW-W | S, SC |
| M 146 | 2.50 | 2.25 | 0.070 | | 0.7 Mischmetal 0.1 Te | CW | SC |
| M 147 | 2.50 | 2.25 | 0.070 | | 0.35 Mischmetal 0.05 Te | CW-G | SC, F |
| M 148 | 2.45 | 1.60 | 0.070 | | 0.35 Mischmetal 0.05 Te | CW-G | SC, S, F |
| M 149 | 2.45 | 1.60 | 0.070 | | 0.35 Mischmetal 0.1 Te | CW-G | SC, S, F |
| M 150 | 2.45 | 1.60 | 0.070 | | 0.7 Mischmetal 0.5 Te | CW-W | SC |
| M 151 | 2.45 | 1.60 | 0.070 | | 0.45 Mischmetal 0.08 Te | W | |
| M 152 | 2.45 | 1.60 | 0.070 | | 1.0 Mischmetal | W | |
| M 153 | 2.45 | 1.60 | 0.070 | | 0.45 Mischmetal 0.01 Te | CW-W | SC, S |

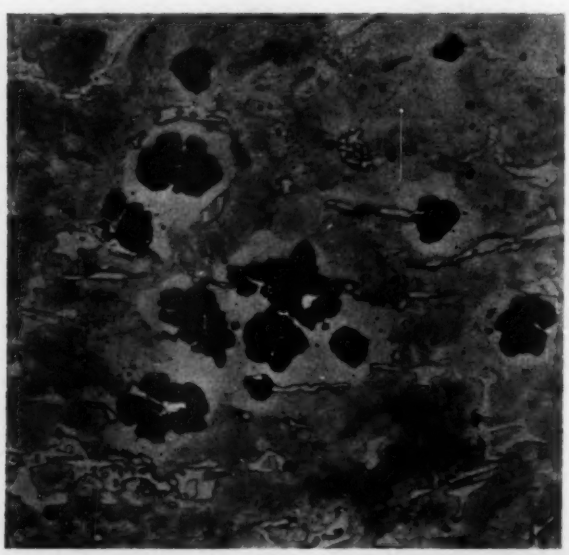
W — White
M — Mottled
G — Gray
CW — Compact Graphite Mottle, White Fracture
SC — Stubby Compact Graphite
S — Spherulitic Graphite
F — Flake Graphite
* — Commercial Laboratory Analysis



Fig. 2 — Specimen M52. Compact graphite mottle-white fracture. 3 x 3 x 8 in. casting. 8 ×.



Figs. 3 and 4 — Specimen M52 etched for 15 sec with 2 per cent nital. 3 x 3 x 8 in. casting. 100 ×.



for each heat are recorded in Tables 1 and 2. Cerium is shown to be most effective in producing white iron or compact graphite mottle.

## ADDITIONS

The additions used and the base iron compositions studied may be classified:

1. Desulfurized Heats
   a. 0.01 to 0.04 per cent Te additions to base compositions of 1.70 per cent Si and 3.90 per cent C, heats M25 to M31, and M34.
   b. Bismuth, same as (a), heats M32 and M33.
   c. 0.04 per cent Te and 0.02 per cent B additions to base compositions of 1.85 per cent Si and 2.68 per cent C, heats M37 to M41.
2. Cerium Added to Desulfurized Heats
   0.10 to 0.80 per cent Ce additions to base compositions of 2.50 per cent C and 1.50 to 2.25 per cent Si, heats M59, M60 and M63.
3. Cerium added to undesulfurized heats, 0.01 to 0.80 per cent, heats M70 - M83.
4. Mischmetal added to undesulfurized heats, 0.30 to 0.48 per cent added, heats M84 to M88.
5. Mischmetal and tellurium combined, added to undesulfurized iron, 0.30 to 0.72 per cent mischmetal and 0.01 to 0.10 per cent Te added, heats M91 to M99, M111, M112, M114-118, M120-124, M127-134, M137-153.

The effects of these additions on the fracture and type of graphite formed will be discussed.

## COMPACT GRAPHITE IN THE AS-CAST CONDITION

### *Desulfurization*

In order to produce compact graphite in the as-cast condition, it was thought that a low sulfur content might be helpful as in ductile iron. To achieve low sulfur irons basic melting practices were used. Basic slags composed of calcium carbide, calcium oxide and calcium fluoride were initially used and

were successful in reducing sulfur from 0.051 to 0.025 per cent. These slags were added to the melt at 2750 F (1510 C), held for 10 min and skimmed off.

Other slags used were composed of calcium oxide, magnesium oxide, sodium carbonate, portland cement, calcium carbide and silica. The basicity was generally kept at 2.3 and the weight about 4 per cent of the charge weight. The slags were added with the cold charge and kept on the molten bath until shortly before pouring. The sulfur was lowered from an original level of 0.051 to 0.015 - 0.025 per cent.

These desulfurizing treatments were applied to heats in both the hypoeutectic and hypereutectic composition ranges. In the hypereutectic region, tellurium additions were made to a base composition of around 3.90 per cent C and 1.70 per cent Si. Metallographic examination showed that in a normally gray freezing composition, tellurium suppressed proeutectic flake graphite precipitation. The graphite present was of lacy and compact type. The compact graphite was generally of stubby to slightly spherodized shape. While 0.01 per cent Te was sufficient to prevent proeutectic flake graphite, it was necessary to add 0.04 per cent Te to produce the best structure. Heat M34 showed the best compact graphite up to that time produced.

In some hypereutectic heats bismuth was added. Despite working at lower silicon levels (40 points lower) the bismuth was not able to prevent proeutectic flake graphite or Kish.

The base composition used for hypoeutectic heats was around 2.68 per cent C and 1.85 per cent Si. A standard addition of 0.04 per cent Te and 0.02 per cent B was made. Metallographic examination showed that heats not desulfurized contained lacy flake graphite. Upon desulfurizing, this lacy graphite disappeared. The other form of graphite noted was of a compact nature. This was of stubby to slightly spherodized shape. Heat M46 was the best structure produced.

### *Cerium Additions, Prior Desulfurization*

Since cerium is a known desulfurizer, it was decided to add this element as a late plunging addition. In heat M52, an addition of 0.03 per cent Te and 0.68 per cent Ce was made. The microstructure of this specimen contained well formed compact graphite, as shown in Figs. 3 and 4. This was by far the best structure produced. The casting fracture was also distinctively different from past heats in that it contained many mottle specks which were fine, almost pin-point in size. This was described earlier, and is called compact graphite mottle-white fracture, CW.

Additional heats were made using cerium additives but no tellurium. Heats M59, M60 and M63 show the effect of varying amounts of cerium additions. The change from gray fracture to compact mottle-white occurred at 0.25 to 0.30 per cent Ce added. In all cases the compact graphite was well formed but not as good as M52 where Te was used as well as Ce for addition. All heats where Ce alone was added showed remnants of flake graphite and had some stubby, in addition to the better formed compact graphite. Higher Ce additions produce better compact graphite at percentages above the minimum needed to produce a CW fracture.

### *Cerium Additions, No Prior Desulfurization*

Further work with cerium was continued. However, in this series of heats no desulfurizing was done other than by the cerium addition. Heats M73 through M83 show the effects of different per cent cerium additions at different silicon levels. The two silicon levels selected for the base composition were 1.55 and 2.25 per cent Si along with 2.50 per cent C. At each of these silicon levels a series of cerium additions (0.2, 0.4, 0.6 and 0.8 per cent) was made.

At any of these silicon levels the percentage addition required to change from flake graphite iron fractures to a compact graphite or a white iron structure was between 0.3 and 0.4 per cent Ce. Below 0.3 per cent Ce the desulfurizing and carbide stabilizing effect of the cerium was not sufficient to suppress flake graphite. Above 0.4 per cent Ce the flakes were suppressed and a CW structure resulted. The number of compact graphite particles formed was not appreciably affected by silicon percentage, but the size (and extent of graphitization) was greater at the higher silicon percentage.

### *Mischmetal Additions*

At this point it was decided to use mischmetal as a source of cerium. A commercial grade of mischmetal of analysis 47 per cent Ce and 95 per cent total rare earths was secured. Heats M84 and M85 compared the effect of mischmetal and comparable amounts of cerium. From these heats it was concluded that comparable amounts of cerium and mischmetal are equivalent in their effect on the microstructure of cast iron for the grade of mischmetal tested.

Another series of heats was run to determine the effect of the temperature of addition of mischmetal. Heat M86 had mischmetal added at 3000 F (1649 C) and heat M87 had it added at 2900 F (1593 C). Neither of these microstructures showed significant change over that when cerium was added at the standard temperature of 2750 F (1510 C).

In this latter work it was noticed that a rather sharp break existed between the formation of flake graphite and a white or CW structure as the cerium addition was increased. This break was narrowed down to about 0.4 per cent cerium or mischmetal in undesulfurized iron. Slightly below this level flake graphite existed, but also with abundant compact graphite. A mischmetal addition of slightly over 0.4 per cent eliminated these flakes, but also removed much of the desirable compact graphite.

### *Mischmetal Plus Tellurium Additions*

Up to this point the best microstructure had been achieved with a combined cerium and tellurium addition. To determine what effect tellurium had and also to determine an optimum addition, a series of heats was run (M91-M99). In this series the base com-

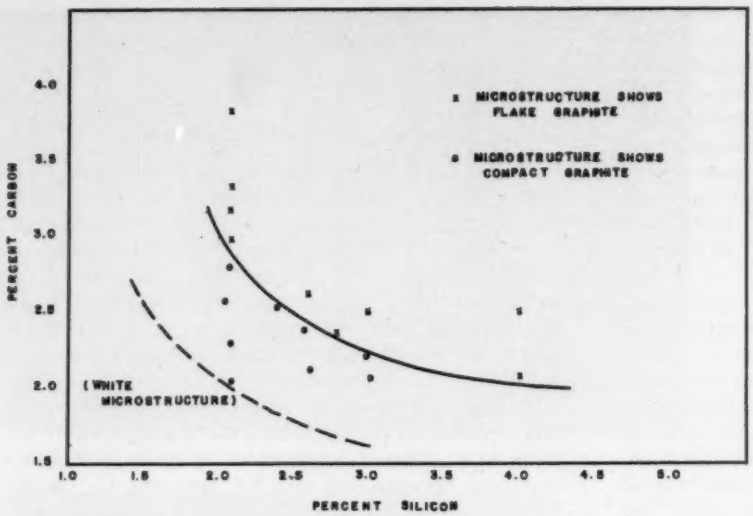



Fig. 5 — Composition range of compact graphite structures in as-cast irons as a function of per cent silicon and per cent carbon. A standard addition of 0.7 per cent mischmetal and 0.1 per cent Te was used.

TABLE 3 — BASE COMPOSITION EFFECT ON GRAPHITE STRUCTURE

| Heat No. | Analysis, % C | Si | C.E. | Fracture | Nodule No./mm² | Structure |
|---|---|---|---|---|---|---|
| M 114 | 2.58 | 2.05 | 3.26 | CW | 458 | SC |
| M 115 | 2.80 | 2.09 | 3.50 | CW | 342 | SC |
| M 116 | 2.99 | 2.08 | 3.68 | CW-G | 225 | F, SC |
| M 117 | 3.19 | 2.08 | 3.88 | G | | F, SC |
| M 118 | 3.44 | 2.08 | 4.13 | G | | F, SC |
| M 120 | 3.84 | 2.08 | 4.53 | G | | F, SC |
| M 121 | 2.30 | 2.08 | 2.99 | CW | 558 | SC |
| M 122 | 2.08 | 2.08 | 2.77 | CW | 440 | SC |
| M 123 | 2.56 | 2.40 | 3.36 | CW | | SC |
| M 124 | 2.38 | 2.57 | 3.24 | CW | 654 | SC |
| M 127 | 2.34 | 2.77 | 3.26 | CW | 700 | SC, S, F |
| M 128 | 2.50 | 4.00 | 3.83 | CW-G | | SC, F |
| M 129 | 2.08 | 3.00 | 3.08 | CW-W | 108 | SC |
| M 130 | 2.08 | 4.00 | 3.40 | CW-G | | SC, F |
| M 131 | 2.13 | 2.60 | 3.00 | CW-W | | SC |
| M 132 | 2.25 | 3.00 | 3.25 | CW | 543 | SC |
| M 133 | 2.63 | 2.60 | 3.50 | CW-G | | SC, F |
| M 134 | 2.50 | 3.00 | 3.50 | CW-G | | F, SC |

CW — Compact graphite mottle-white fracture.
W — White fracture.
G — Gray fracture.
F — Flake graphite.
S — Spheroidal graphite.
SC — Stubby or nonspheroidal graphite.

position was 2.50 per cent C and 2.25 per cent Si. At different levels of mischmetal addition (0.3, 0.5, 0.7 per cent) varying amounts of tellurium (0.01, 0.05, 0.1 per cent) were added.

It was found that with 0.3 per cent mischmetal addition 0.01 and 0.05 per cent tellurium additions were not sufficient to suppress flake graphite while 0.1 per cent Te was. M92 (0.05 per cent Te) showed the best compact graphite including some spheroids. A few surprising observations were: 1) M91 (0.01 per cent Te) and M92 showed a chilled rim 2) M92 showed heavier flakes than M91 3) M92 and M93 had quite a few sulfides present. The sulfides were not expected because of the desulfurizing effect of mischmetal, and are not present if tellurium is not used or the per cent added is low.

With the 0.5 per cent mischmetal addition all levels of tellurium addition suppressed flake graphite. M94 (0.01 per cent Te) displayed the best compact graphite. Again, it was surprisingly noted that many sulfide inclusions appeared in the higher tellurium addition heats, i.e., M95 (0.05 per cent Te) and M96 (0.1 per cent Te). Few if any spheroids were observed.

All the 0.7 per cent mischmetal heats, M97 (0.01 per cent Te), M98 (0.05 per cent Te) and M99 (0.1 per cent Te) produced the best microstructures up to this time. M99 had the best examples of compact graphite while M98 was second. Again, sulfides were noted in M98 and to lesser extents in M97 and M99.

Another series of heats was run at 1.50 to 1.60 per cent Si and 2.35 to 2.50 per cent C, in which the size of tellurium and mischmetal addition was varied. Results were similar to those of the higher silicon level. 0.7 per cent mischmetal and 0.1 per cent Te were found to give the best compact graphite structure. One interesting fact noted in this low silicon range was that a good compact graphite was observed at 2.40 per cent and 1.60 per cent Si, and an addition of 0.7 per cent mishmetal and 0.1 per cent Te. The graphite nodules were well formed, and some appeared to be good examples of spheroids. However, these nodules were small.

### *Carbon and Silicon Effect*

To determine the influence of base composition on graphite structure, heats were made with a standard addition of 0.7 per cent mischmetal and 0.1 per cent Te. These heats are listed in Table 3, and plotted in Fig. 5. In cases where it was possible to make nodule counts these are listed. However, such counts should be considered with reservations as the error involved in this low nodule counting area is considerable.

In these heats the variables were the per cent carbon and the per cent silicon and the carbon equiva-

lent. In heats M112, M114, M115, M116, M117, M118, M121 and M122 the per cent silicon was held constant at about 2.08 per cent, while the per cent carbon was varied from 2.08 to 3.84 per cent. This gave a carbon equivalent (C.E.) range from 2.77 to 4.53. A result of this series was that the nodule number increased to a maximum as the per cent carbon was increased from 2.08 to 2.30 per cent. Increasing the per cent carbon above this point resulted in lower nodule counts.

The microstructures showed less compact graphite and a decrease in the quality of the graphite as the per cent carbon was raised. In a reciprocal manner the amount of flake graphite increased. These developments proceded in a progressive manner up to a C.E. of 4.3 or the eutectic composition. A heat M112 of hypereutectic C.E. showed flakes and compact graphite. It was not possible to take a nodule count, but it did appear there were slightly more nodules in this heat than in M118, a comparable heat just below the eutectic C.E. of 4.3.

In another series of heats both the per cent silicon and the per cent carbon were varied at a constant C.E. The range for C.E. was 3.00 to 3.83. Heats covered were M123, M124 and M127 through M134. It was noted that in comparison to heat M114, heats M124, M127 and M132 showed that at a particular C.E. a higher proportion of silicon resulted in an increased nodule number. Nodule counts made for M124, M127 and M132 showed that the nodule number reaches a maximum and then decreases as the per cent silicon is increased at a constant C.E.

Again, caution should be observed because of the inherent error involved in making these counts at low nodule numbers. As C.E. was increased the graphite was observed to become less compact. To summarize, with additives of 0.70 per cent Ce and 0.10 per cent Te to undesulfurized iron, the range of 2.35 to 2.50 per cent C and 1.60-2.25 per cent Si produced the maximum number and size of compact graphite nodules, as-cast.

## SUMMARY

Tellurium has a major effect of suppressing flake graphite nucleation in the normal eutectic temperature range. It has a minor effect of permitting a limited nucleation and growth of compact graphite. These effects are dependent on the carbon and silicon content and amount of tellurium addition. At carbon and silicon percentages just over the mottle limit for a 0.01 per cent Te addition, the graphite is lacy and the fracture mottled. Raising the Te addition to 0.04 per cent permits compact graphite to predominate. The fracture then changes from mottled to a compact graphite-white fracture.

Raising the carbon and silicon percentages causes a return of the lacy graphite, and requires an increase in Te addition to eliminate this flake graphite. There appears therefore to be a limiting number of compact graphite nodule that can be developed using Te alone. The maximum number of compact graphite nodules is 150/mm$^3$ developed in heat M11. Tellurium was effective in both hypoeutectic and hypereutectic compositions.

Bismuth appeared incapable of preventing flake graphite nucleation in a mottled fracture in the hypereutectic composition range. No work was done with hypoeutectic heats.

Cerium was able to prevent flake graphite from forming at 1.50 per cent to 2.25 per cent Si and 2.50 per cent C. It also served to nucleate compact graphite and spheroids. The addition required depended on the percentage sulfur in the iron. In undesulfurized iron at 0.07 per cent S, an addition of 0.40 per cent Ce is the minimum required. The fracture changes from gray to compact mottle-white fracture with no transition to the conventional mottle fracture. These observations also apply to the use of mischmetal to replace cerium.

When Te and cerium are combined, a lower percentage of cerium can be used to obtain compact graphite. For example heat M93 had 0.30 per cent cerium added and would have a gray fracture, but by the addition 0.10 per cent Te was caused to have a compact mottle-white fracture.

Based on the additions and base compositions studied, a combination of cerium or mischmetal and tellurium has produced the best as-cast compact graphite. This is exemplified by heats M52, M99 and M139, and Figs. 3 and 4. Base composition has considerable effect on as-cast compact graphite. In the range of 2.35 to 2.50 per cent C and 1.60 to 2.25 per cent Si, the maximum amount of compact graphite was noted.

It should be recognized that this is a progress report of a limited amount of research aimed at producing compact graphite in cast irons of a chemical composition in the malleable iron range. There is no intention of advocating a certain addition or process, nor of implying that this can be accomplished on commercial castings. However, it is hoped that research will be continued in this direction.

## ACKNOWLEDGMENT

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# COST CONTROL THROUGH PROCESS QUALITY CONTROL

by C. E. Haney


## ABSTRACT

**The author's company has established a process quality control program for cost reduction and quality improvement. This article discusses the need for process control, the approach to identification of significant process cost and quality factors, application of process controls and organization of the new department.**


## INTRODUCTION

It is tough to make a profit in the steel foundry business in the current competitive market. Prices are low, material and labor costs are continually rising and production quality requirements are getting more rigid all the time. Under these conditions, management looks one remedy squarely in the eye – cost reduction.

There are, of course, a number of ways to reduce costs. Some common examples are work simplification programs, intelligent and alert purchasing practices, up-to-date plant layouts, a streamlined organization and other efforts designed to promote efficiency. Good management must recognize the importance of all such programs in order to insure growth in spite of adverse current conditions.

A newer approach to cost reduction which can be extremely important, particularly in the jobbing foundry where true repetition is rare, is process quality control.

Quality control usually implies inspection and control of an end product or component. Process quality control provides inspection and control of processes, and considers the end product, the individual casting, as it is affected by processes in its manufacture. Processes which are capable of producing the product to a satisfactory quality level must be chosen and kept in control.

The following discussion outlines the author's company's efforts in the field of control of process. The idea is not original – other foundries have set up parts of the program in the past. The methods of implementing and integrating process quality control in this operation may present some new twists for consideration.

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## NEED FOR PROCESS QUALITY CONTROL

### *The Trial Casting Problem*

An illustration of the need for control of process is brought out by experiences with trial castings. When a customer's drawing arrives at the plant and a pattern is made, the first step, as in most foundries, is to make a trial before embarking on the production run. The trial is fully evaluated to be sure that it meets the customer's specification before releasing the pattern for production. A fairly common incident has been production of a suitable trial casting which passes all evaluation resulting in release of the pattern for production, and subsequent production of a group of castings which must be scrapped or heavily repaired before shipment.

A recent specific example was a group of washer castings for a missile launching installation. These were low alloy weighing about 100 lb, and had neck-down cores between the riser and the casting. The trials were excellent, there was no evidence of any defect at the riser contact or under the neck-down core and the castings passed x-ray and magnetic particle specifications without repair. The pattern was released for production, and the first 40 castings arrived in the cleaning room with serious sagging difficulties in the neck-down core.

The resulting depression required weld buildup and grinding and increased the cost of production. A few degrees increase in pouring temperature had caused the defect, and correction of this detail of process allowed further production without trouble. This example illustrates the way controlled process could have reduced costs by providing consistant quality.

### *Inconsistent Scrap Rates*

From time to time high production items of low scrap history suddenly spring forth as scrap problems due to lack of process quality control. For example, a small casting which was made on a production

basis for years with well below one per cent scrap average suddenly turned up as a 13 per cent scrap item. The reason, broken cores; the remedy, correction of a defective core setting process. This resulted in the casting, which triggered the investigation, dropping back to its usual scrap rate. More important, scrap rates on other jobs, which were being adversely affected by the faulty process, were improved.

### *Inconsistent Repair Requirements*

In spite of a lot of experience making certain production castings, wide variations in cleaning room repair time required to prepare the casting for shipment were found. One of the best examples was a large stainless casting which had cleaning and repair times as low as 48 hr, and as high as 250 hr, within a group of about 50 castings. The correction of the condition causing this wide variation was modification of the gating system. Turbulent metal delivery had resulted in inconsistent surface on the casting requiring much grinding and chipping.

By the time full process quality control treatment had been applied, the average time was reduced from 118 hr on the first two-thirds of the order to 55 hr on the final third. In addition, the customer has reported a 60 per cent improvement in casting quality as indicated by dye penetrant examination following final machining. Movies were used to investigate this problem, and changes made after studying the pictures.

It is recognized that cost reductions of this magnitude are not usually possible; this particular casting required nearly perfect x-ray and dye penetrant inspection and an excellent surface. However, lesser but still significant cost reductions are possible on almost any type of casting.

## IDENTIFICATION OF SIGNIFICANT PROCESS COST AND QUALITY FACTORS

### *"Pilot Heat" System*

As a result of observing the difficulty of reproducing results, much attention was given to the quality control program. Quality control principles were already being applied to the obvious cost and quality factors, and an extensive casting inspection program was in operation. Some of the quality control efforts were already directed toward process—sampling of raw materials, daily sand testing, melt analysis using a direct reading spectrograph and control of pouring temperature with an immersion thermocouple.

In addition, many practices had been established and adhered to which apparently affected consistency of product—careful control of charge materials, purchase of refractories from reputable suppliers, development of standard ladle preparation techniques, mold handling and drying practices, sand purchasing and process sample testing, and so on.

In addition, the laboratory provides all types of tests for verifying the end product quality, and the inspection force is equipped from *A* to *Z* for discovering flaws, including a 24 million volt betatron. As a matter of fact, from time to time it has been facetiously suggested by the company that the biggest product consistency problem is being so good at finding out what is wrong with the castings. If the castings could be given the quick once over of 15 years ago, the company would go merrily on its way, confident of excellent continuing quality.

Inconsistencies in scrap rate were noticed, and repair requirements continued in spite of application of the quality controls. The obvious conclusion that at least some of the factors which had an effect on quality were not being controlled, and the company set about the job of identifying these factors. At this point the "pilot heat" procedure was originated. This consisted of assigning two top men, the Asst. Works Mgr. and the Manager of the Metallurgical Department, to the job of observing every detail of heats scheduling castings showing wide quality variations.

Their approach was to gather all the information they could on the details of processes in practice. Nothing was excluded as being too unimportant for consideration as a possible factor. Samples of the type of information gathered are shown in Figs. 1 to 4.

An example of the type of information collected and considered during this pilot heat period of the development of process quality control was handling of backing sand at one molding station. One molder would shovel spilled sand from the floor into the mold as backing sand without adding any water to it. The molder on the next machine liked to sprinkle down the floor around his machine, and when he shoveled this sand into the mold as backing it was considerably wetter. True, it was never established definitely that this had an effect on quality, but it is a good illustration of the type of detail which is normally completely unnoticed that was considered as a possible factor in process quality control during the pilot heat period.

During this pilot heat development period, it was noticed that casting quality jumped up when a heat was labeled a "pilot heat"; no bad castings were produced during these heats. Since this was not limited to one pattern or another, it was concluded that the effect of added attention on process was responsible.

### *Brainstorming the Problem*

When the pilot heat procedure was well established, a brainstorming session was arranged. Here an attempt was made to get every idea supervisors and pilot heat observers could suggest on possible quality factors in the processes. This was approached by taking the top ten scrap causes and discussing the possible reasons for their occurrence. A sample idea list which resulted is shown in Fig. 5.

Notice that a number of ideas were listed which have limited bearing on the scrap cause. It is felt that the atmosphere of free discussion with no idea being rejected was most beneficial.

### *Process Quality Control Check Lists*

The ideas gathered at the brainstorming session were discussed and either retained or discarded, leading into the development of the process quality control manual. A page from this manual is shown in Fig. 6.

PILOT HEAT

MOLD DATA

HEAT NO. _____ Date _____

Time molded _____ By _____
Mold hardness _____ C _____ D
Time wash applied _____ Lbs. wash _____ Method _____
Wash viscosity _____
Time mold completed _____ Time oven started _____
_____ hrs. to _____ . _____ hrs. at _____
Time out of oven _____ Time cores in mold _____
Time heater on _____ Time heater off _____
Time closed _____ Core temperature _____
Temperature measured where _____
Temperature measured how _____
Time heat tapped _____ Time mold poured _____
Pour time, seconds _____
Ladling detail remarks _____
_____
_____
Pouring temperature _____ How measured _____

**Fig. 1 — Sample of mold data gathered. See also Fig. 2.**

PILOT HEATS

MOLD DATA (2)

SAND DATA
Facing Sand

| | | Permeability | | Compression | | | | |
|---|---|---|---|---|---|---|---|---|
| Type | Moisture | Green | Dry | Green | Dry | Flowability | Deformation | Density |
| | | | | | | | | |
| | | | | | | | | |

Backing Sand

| | | Permeability | | Compression | | | | |
|---|---|---|---|---|---|---|---|---|
| Type | Moisture | Green | Dry | Green | Dry | Flowability | Deformation | Density |
| | | | | | | | | |

REMARKS
Molding _____
_____
_____
Mold washing _____
_____
_____
Mold drying _____
_____
Mold handling _____
_____

**Fig. 2 — Sample of mold data gathered, in addition to Fig. 1.**

As is evident, this is an edited list of the ideas from the brainstorming session in the form of a checklist for the process quality control inspector to use in performing his job. This manual is used by the process inspector in investigation of deficiencies, either by location or by job. Notice that "process" and "pattern" factors are included in this checklist. As is immediately obvious, a defect under investigation may be caused by a process deficiency (misrun, for example, by low pouring temperature) or a problem with an individual pattern (misrun, for example, by a poorly designed gate).

## APPLICATION OF PROCESS CONTROLS

### *Casting Planning Stage*

Choosing processes capable of producing the casting required is the responsibility of the job methods

PILOT HEAT

CORE DATA

Heat No. _____ & _____ Date _____

| Core Hardness | | Hardness Green | | Scratch | |
|---|---|---|---|---|---|
| Core Number | _____ | C | D | C | D |
| Core Number | _____ | C | D | C | D |

Core Maker _____ Core Sand _____

| Type & Batch | Moisture | Permeability Green | Permeability Dry | Flowability | Deformation | Density |
|---|---|---|---|---|---|---|
| | | | | | | |
| | | | | | | |

Time core completed _____ Time start dry _____
Time first coat of wash _____ Method _____
_____ hours to _____ _____ hours at _____
Time out of oven _____ Time drag core washed _____
Time pasted _____ Time cope core washed _____
Time in oven for wash dry _____ _____ min. at _____
Date wash batch prepared _____ Viscosity _____
Wash application method _____
Core temperature when washed _____ Cope _____ Drag _____
How measured? _____

**Fig. 3 — Sample of core data gathered. See also Fig. 4.**

PILOT HEATS

CORE DATA (2)

Wash applied by _____ Core pasted by _____
Core to hot room _____ (Date) _____ (time)
Core out of hot room _____ (Date) _____ (time)
Core in mold _____ (time)

REMARKS
Core making _____
_____
_____
Core washing _____
_____
_____
_____
Core handling _____
_____
_____
_____

**Fig. 4 — Sample of core data gathered, in addition to Fig. 3.**

group. This group functions on the staff of the foundry superintendent, and its recommendations on risering, gating, pattern equipment and special foundry techniques must be accepted by the superintendent or one of his assistants, and the molding foreman before the trial is rammed up. Permanent records are initiated on each new job, including detailed rigging sketches. When trial castings have been made and evaluated, the results are added to the information in the files. Any changes required are noted, and subsequent trial castings results are recorded until a satisfactory part is produced and the pattern is released for production.

The purpose of making such a point of recording every step of the trial casting work is twofold. First, it is useful in troubleshooting should a problem develop at some later time. Second, successful techniques used on one casting can later be applied to similar castings. The files are arranged by customer name, with cross references locating the casting folder by pattern number and casting description.

If, during this planning stage, the job under consideration appears likely to be particularly troublesome, or if it promises to be an important production item, it is designated a *P* order; a "Process Quality Control" order. This prefix on the order number says "look out for this: it is tough and important," and alerts everyone to give special care to the job.

If some specific problem is foreseen, special precautions will be taken as appropriate. Examples are special handling instructions, trial evaluation beyond the specification, sending the trial to the customer for approval or any number of similar special steps designed to ensure a satisfactory quality level.

If the problem cannot be predicted before the trial is made, but special control is necessary, the process quality control inspector will be asked to pilot the heat in which the trial is poured.

The job methods system of casting planning has contributed materially to scrap reduction and improved casting quality with lowered cleaning and repair costs. It is tied in directly with the process quality control program; the files of both are consolidated where individual jobs are reported on, and process quality control inspection is often called upon for investigation of a pattern problem by the Job Methods Department.

### *Process Controls*

When processes have demonstrated their capacity to produce a satisfactory trial casting, responsibility for control of the processes is assumed by the supervisory group. Process quality control is applied as staff assistance where required.

The processes included in the current program are molding, coremaking and pouring. Conspicuous by their absence are melting, sand preparation, cleaning and heat treating. Melting is controlled by the metallurgical laboratory staff, and statistical quality control and inspection of the process, as well as record keeping, are the responsibility of this group.

Sand preparation is controlled by the sand lab, which is a part of job methods. Sand sampling and charting of properties found in the samples are long

MISRUN

PROCESS

A. Wrong pouring temperature (cold metal or too hot).
B. Heavy Oxide (Hi alloy-Mn).
C. Improper pouring technique.
   1. Ladle too high.
   2. Slow pouring.
   3. Interrupted pouring.
   4. Mold improperly weighted.
   5. Number of stops.
D. Lip designed wrong. Wide stream.
E. Sticky stopper.
F. Stopper size. Pouring too slow.
G. Ladle too full.
H. Cold teapot spout.
I. Poor shank fit.
J. Unskilled steady man.
K. Mold media.
L. Mold temperature.
M. Location of mold on rack.
N. Mold position. Uphill pouring.
O. Floor preparation. Slow heat pour-off.
P. End core size and cup size. (c)
Q. Metal in ladle as related to size castings. (Carrying metal back and forth in pouring ladle.)
R. Spinning speed.

PATTERN

A. Gate too small.
B. Improper ingate contact.
C. Improper flows.
D. Rigging (Williams vs. Open Riser)
E. Gate location.
F. Chilling - location, size.
G. Size - shape pouring cup.
H. Pounds castings per box.
I. Casting position.
J. Casting location as related to gate.
K. Casting design - Metal section.
L. Core venting.
M. Low Cope - insufficient head.

**Fig. 5 — Sample idea list resulting from brainstorming session.**

standing procedures, as are periodic inspections of sand handling and mixing equipment. A newer tool which permits "phantom" process inspection is the use of a recording wattmeter on the sand mill to observe and record the mixing procedure.

MISRUN CHECKLIST

Inspector:_____ Date:_____ Location:_____
Pattern or Patterns Inspected_____ Heat No._____
Process
Pouring Temperature_____
Pouring Technique Comments_____
_____
Ladle Height O. K.?_____ Stops per Ladle_____
Pouring Time_____
Ladle Comments_____
_____
Spout O. K._____ Filled Properly_____
Stopper O. K._____
Mold Placement, Position_____
Pattern
Gating Comments_____
_____
Size_____ Contact Size_____
Location_____ Sprue size & type_____
Cope Height_____
Rigging Comments_____
_____
Riser Placement_____
Chill Placement_____
Recommendations
_____

**Fig. 6 — Sample page from process quality control manual. Compare with checklist in Fig. 5.**

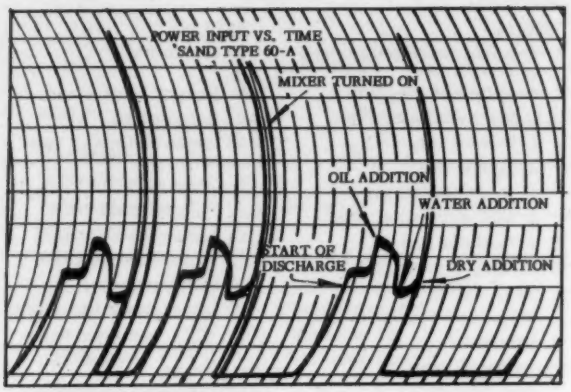



Fig. 7 — Tape on which wattmeter has plotted power input vs. time for batches of the same type of sand.

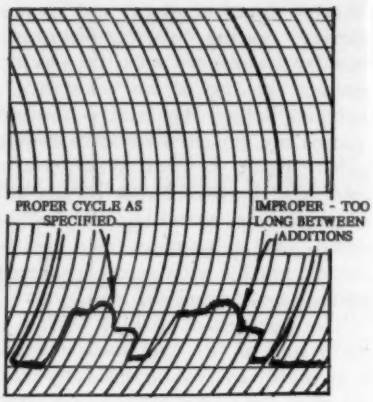



Fig. 8 — Wattmeter tape for power input vs. time for sand type (61D) showing proper and improper mixing of batches of sand.

Figure 7 shows a tape on which this device has plotted power input versus time for a group of batches of the same type of sand. These sand batches have been mixed properly and consistently. Notice the similarity in curve shape. The chart reads from right to left and each vertical graph line represents 2 min in time. The curve fluctuations are marked to show the operation details.

In Fig. 8, curves showing improper and proper mixing of batches of another type of sand are shown. This particular cycle calls for mixing 3 min after both water and oil additions. In the first cycle, 4 min went by before the subsequent addition. The second curve describes a properly mixed bath. The sand laboratory has excellent control of sand preparation processes, and the application of the new process quality control program to other, more fertile fields was considered more important.

A number of controls have long been applied to heat treating processes. Periodic oven surveys and oven temperature-time charts are used, as is inspection of the process by members of the metallurgical laboratory and operating supervision.

The cleaning process in all of its complexity is not included in the current process quality control program. While it is true that cleaning processes greatly influence the final quality of the product, the inherent casting quality as it is shaken out is a much greater factor in final cost, and thus efforts have been concentrated on processes which can improve the quality of the casting as it arrives in the cleaning room.

The areas of concentration; molding, coremaking and pouring, are handled in one of three ways by the process quality control inspector—by spot checks, trouble shooting and pilot heat procedures.

### *Spot Checks*

As an example, during a typical spot check of a molding location the inspector shows up unannounced and looks over the entire process. The pattern is inspected for general condition and suitability, as are all other pieces of equipment in use. Flasks (watch those pins and bushings—go, no-go gage them), jackets for snaps (clean and square), pattern plates (how is the match? section a mold), core fit (who knocked the corner off that print?), pattern wear (check the castings for dimensions if you are suspicious), squeeze pressure (how is mold hardness?) and so on down the list of equipment in use.

The inspector then goes over the process as affected by the molder. Proper facing (check mold hardness, facing depth, placement of special sands in fillets), ramming OK (watch for ram-off in the sectioned mold), gates cut and matched properly (uniform coat on entire mold), mold closing (cleanliness pays) and on through his checklist of molding processes.

The inspector's observations are reported to the superintendent who takes whatever corrective action is required through the foreman in charge of the molding area.

### *Trouble Shooting*

If the scrap report shows that one particular pattern is giving trouble, or if cleaning room supervision reports an excessive repair or processing problem, the process quality control inspector may be called upon for determination of the cause of the defective castings. For example, Fig. 9 shows an excerpt from a scrap report which indicates that pattern T-196 has recently become a problem from the standpoint of misrun.

This report is made up by the tabulator from punched cards; the illustration is in different form because of size restrictions, but the column headings are self-explanatory and are as they appear on the report. The scrap versus production figures are checked against current target scrap level for this defect, shown in Fig. 10. These targets were arbitrarily chosen for 1960 based on desired improvements on losses for 1959. One more criterion is used which is completely arbitrary; this is the $100 minimum loss figure.

If less than $100 has been lost in scrap on an item, it is not pursued until the higher loss items have been worked on. This is done to avoid emphasis on low dollar producing items and low production jobs. It is easy to understand that higher percentage losses can be expected during early stages of production;

| Location | Pattern Number | Cause | Current Month # Prod | Current Month # Scrap | Current Month $ Prod | Current Month $ Scrap | Year-to-Date # Prod | Year-to-Date # Scrap | Year-to-Date $ Prod | Year-to-Date $ Scrap | % of Scrap Year to Date | % of Scrap Current |
|---|---|---|---|---|---|---|---|---|---|---|---|---|
| 2 | T196 | | 50 | | 1,171 | | 326 | | 8,845 | | | |
| 2 | T196 | 2 (Slag Inclusion) | | | | | | 2 | | 58 | | |
| 2 | T196 | 10 (Misrun) | | 8 | | 238 | | 8 | | 238 | | |
| 2 | T196 | 13 (Crack) | | | | | | 1 | | 22 | | |
| 2 | T196 | 40 (Deformed or broken) | | | | | | 1 | | 28 | | |
| 2 | T196 | 57 (Low Physicals) | | | | | | 1 | | 28 | | |
| 2 | T196 | 59 (Quench Cracks) | | | | | | 1 | | 30 | | |
| 2 | T196 | | 50 | 8 | 1,171 | 238 | 326 | 14 | 8,845 | 404 | 19.0 | 4.5 |

Fig. 9 — Excerpt from a scrap report which indicates that pattern T-196 has recently become a problem from the standpoint of misruns.

this $100 cutting line lets the job "settle down" after the trial stage before a lot of effort is devoted to correction of scrap difficulties.

### *Control Limits*

Figure 11 shows the control limits which are used in deciding whether or not to work on a scrap problem exceeding the $100 minimum, relating its importance in the overall scrap loss to numbers produced. This chart has been adapted from one designed by Kenneth M. Smith in a series of recent articles on the "average job" quality control system. As you can see, for a given target scrap level, a higher loss is tolerated in low production items than in high production.

Figure 12 shows the flag card used to notify pattern storage that action is to be taken on this pattern the next time an order is entered. Notice that action course no. 4 has been circled by the scrap report evaluator. This means that the process quality control inspector will be asked to investigate this job. Several other alternates are provided which can be used for different problems; for example, if shrinkage were the problem, this flag would serve as a reminder for job methods to review the risering, and action alternative no. 2 would be circled on the card.

The flag card is returned to the job method files when the next order is entered, and is matched with the duplicate card which is placed in the file folder with any supporting information. When the process quality control inspector is called upon, such as he is in the sample misrun problem, he will attack the correction by the use of his checklist which has been developed for the defect. His observations and resulting changes are recorded in memo form and retained in the job methods file after corrections are complete.

### *Pilot Heat Procedures*

Still another way the process control inspector functions is in the continuation of the pilot heat procedure. The approach to "piloting" a job has been covered in an earlier section of this discussion; it differs from spot checking and troubleshooting in that the problem to be corrected is usually nonspecific.

For example, there would be no need to pilot the sample pattern number T-196, which is a known misrun offender; a checklist has been developed for that purpose. If, however, on some other pattern the Cost Department reported wide fluctuations in cleaning costs from casting to casting, and further checking showed the defects requiring correction to be of numerous types, asking the process quality control inspector to run a pilot heat would be considered. He would then report his observations in much the same way as the original "piloters" did—no holds barred on ideas, and no detail too trivial for consideration.

Assuming the piloted heat produces satisfactory

| | Defect No. | 1959 % of Gross | Target 1960 (Percent based on Gross Production) |
|---|---|---|---|
| Misrun | 10 | .47 | .23 |
| Cracks | 13 | .32 | .16 |
| Slag Inclusion | 2 | .22 | .11 |
| Quench Cracks | 59 | .16 | .08 |
| Oxide Metal | 36 | .16 | .08 |
| Deformed or Broken | 40 | .16 | .08 |
| Pockmarks | 26 | .15 | .07 |
| Shrink | 14 | .13 | .06 |
| Runout | 11 | .12 | .06 |
| Shift | 6 | .12 | .06 |
| Core Broken | 21 | .09 | .04 |
| Poured Short | 12 | .08 | .04 |
| Core Scabs | 56 | .07 | .03 |
| Mold Scabs | 55 | .05 | .02 |

Fig. 10 — Current scrap level control targets. These targets were arbitrarily chosen for 1960 based on desired improvement on losses for 1959.

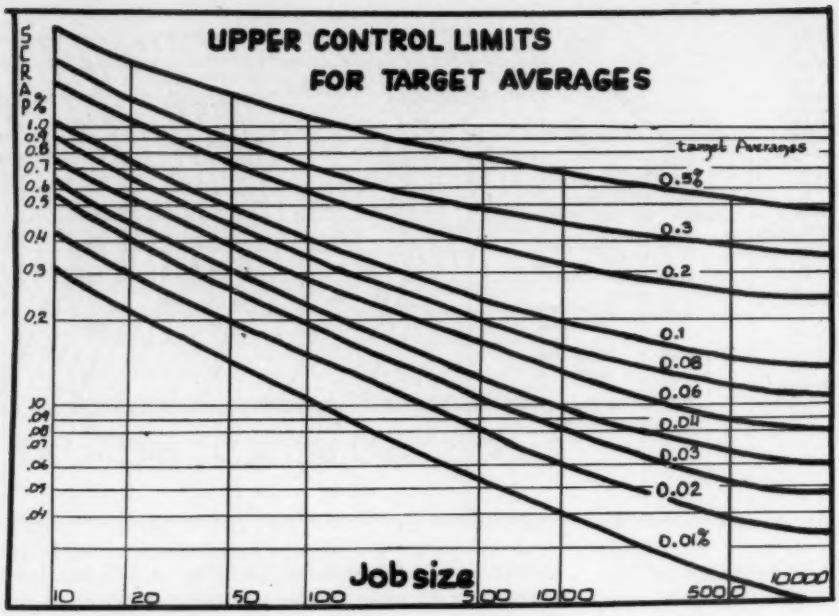



**Fig. 11 — Control limits which are used in deciding whether or not to work on a scrap problem exceeding the $100 minimum, relating its importance in the overall scrap loss in numbers produced.**

castings, as is often the case in practice, the process details observed are adopted as standard for making that particular casting. Special instructions are written for further production. These special instructions are the most important result of a pilot heat; a base for future comparison is established which allows detection of the subtle variations and degenerations which creep into most repetitive nonmechanized operations.

Because of the time required to run them, pilot heats are relatively rarely run, and are limited to high production, repetitive items.

### *Inspection Controls*

The Casting Inspection Department is responsible for verifying final quality of the product, and also for assisting supervision by rejection of bad castings at various stages of production.

Final inspection to be sure the casting meets the specification is most certainly an important function, but the Casting Inspection Department provides a number of services which tie in closely with process quality control—reporting of scrap for tabulation by the Accounting Department, notification of defects needing correction on nonscrap castings, and most important, suggesting means for correction of defects observed. The chief casting inspector sits in on planning sessions in both lab and foundry, and his observations and ideas for corrective action are valuable. Because of the information which he has on defect recurrences, he can often sense trends which show the need for process control work. For example, assume that process quality control has corrected a recurring defect, and that production is satisfactory for a while, but the earlier problem begins to show up again. The inspector is in the best position to wave the red flag and call for renewed corrective effort, and can often suggest the solution through his experience in observing.

In the area of cost control and reduction, the Inspection Department has a vital function in processing castings in the cleaning room—the question of "Does it meet the specification in all respects and is it good enough?" is paramount. Equally important economically is the question "Has too much work been done and is it too good?"

In a shop which produces a wide variety of castings from a quality standpoint, making the casting to the required quality level is vital. It is easy to visualize a grinder who has spent half his day polishing liquid oxygen service valve bodies doing too much work on a ball mill liner. The Inspection Department helps to prevent this by watching for castings which are dressed up beyond necessity.

TO: PATTERN STORAGE When reordered send this sheet to Job Methods File

Main_____

Centrif._____ _____ (Date)

Please flag pattern No _____, last produced in_____ (Location)

This pattern shows_____scrap of_____produced for scrap cause #_____

(_____).

The next time we make this casting we will proceed as follows: (Check appropriate item)

1. Consult Planners
   a. Rigging
   b. Casting configuration
2. Job Methods review of rigging
3. Job Methods review of casting configuration
4. Notify Process Control Inspector
5. Send note to Floor Foreman
6. Other

**Fig. 12 — Flag card used to notify pattern storage that action is to be taken the next time an order is entered. The circled no. 4 means that the process quality control inspector will be asked to investigate the job.**

## ORGANIZATION AND RESULTS

### *Organization*

Figure 13 shows the way the Process Quality Control Department fits into the foundry organization.

Notice that reporting is to the foundry superin-

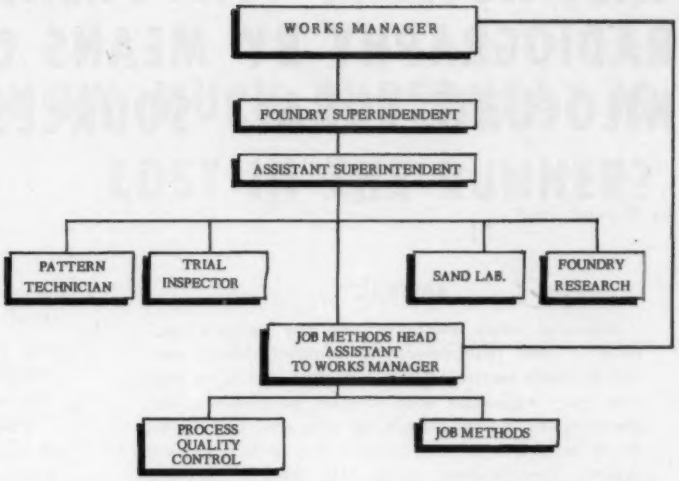



Fig. 13 — The way process quality control fits into the foundry organization.

tendent, through one of his assistants. This firmly establishes the service nature of the function—process quality control effort is thus applied where the superintendent feels it is most needed.

The line organization is separate from this chart. The assistant superintendent shown here also has line foremen reporting to him, but this chart shows only his staff capacity.

Centralization of planning and process control staff under this one assistant superintendent is also important. By this arrangement, he is in the position to coordinate the activities of the entire staff to produce the most effective service to line supervision.

### *Staffing*

Development of procedures for process quality control was a joint effort on the part of the Works Manager, the Foundry Superintendent and the Job Methods Head.

Actual inspection has thus far been performed by one man, who is an experienced and respected foundryman. An injury left him with a disabled arm which incapacitated him for his former leadman's job. His long experience in various jobs throughout our plant, coupled with an especially alert and questioning mind, makes him an ideal process quality control inspector.

The experience of this individual has been an important factor in the apparent initial success of this new approach. He is respected by the men in the shop who know that he is capable of doing any job which he is inspecting. His performance so far indicates that future additional inspectors should be carefully chosen from shop personnel.

### *Results*

It is difficult to estimate the part which process quality control has had in the general improvement of manufacturing costs which have been experienced in the last year. The program is new and small, and several other changes have been concurrent. Line supervisors have become more concerned with scrap losses, with increasing emphasis on their responsibility for control of process with staff assistance available as needed.

Figure 14 shows the recent downward trends in scrap losses and manhours per ton, which have both been influenced by this new program. These are the two yardsticks, other than profit, that are known to use in evaluating effectiveness of cost control through process quality control.

## ACKNOWLEDGMENT

This paper has been prepared with the cooperation and assistance of Mr. William A. Meyer, Assistant to Vice-Pres. and Works Mgr. of ESCO Corp. Mr. Meyer is in charge of the job methods and process quality control work.

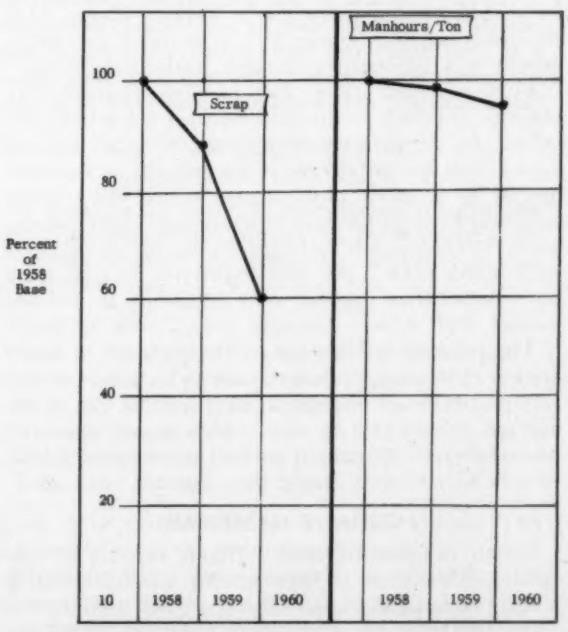



Fig. 14 — Recent downtrends in scrap losses and manhours per ton, which have been influenced by this program.

# CAST NUCLEAR COMPONENTS RADIOGRAPHY BY MEANS OF KILOCURIE COBALT SOURCES

by Richard Lent


## ABSTRACT

**Kilocurie cobalt sources enable the author's company to meet requirements of stringent quality control on heavy section castings without resorting to high cost x-ray equipment with difficult maintenance. Exposure times of from 1.2 min for 2 in. steel to 175 min at 36 in. source to film distance for 10 in. steel are typical. This program meets MIL Spec 271A with qualification from the Bureau of Ships for the range of 1½ to 10 in. of steel.**


## HEAVY SECTION RADIOGRAPHY

Radiography of heavy sections is by no means a new field, as it had been accomplished originally by pioneers in the field using radium sources. The results obtained by this approach to radiography was a far cry from present day radiography requirements, whereby sensitivities of 2 per cent or better are now routinely met.

With the advent of the betatron and linac, heavy section radiography moved forward to a more precise standard. Meanwhile radiography using isotopes had suffered, for to radiograph a heavy section by means of the isotope required a physically large source to accomplish the radiography in a reasonable time. The geometry conditions of such a large physical source in a radiographic setup was a limiting factor. If a physically smaller source had been used the radiography was defeated by excessive scatter conditions.

At the present time, due to high neutron flux reactors, available radioactive isotopes are greatly enriched. In the parlance of the field they are said to have a high specific activity. This now allows a greater output in a small physical size which fully meets radiography requirements of geometry. These sources allow heavy section radiographs to 12 in. of steel with better than 2 per cent sensitivity, and in some cases better than one per cent sensitivity in sections greater than 2 in.

The predominate isotope to be discussed in heavy section radiography is cobalt −60. The nature of this isotope lies in an emanation of a gamma ray of the average energy of 1.25 mev, which to all intents is mono-energetic. We might say that this behaves similar to a heavily filtered 2 mev x-ray beam.

## COBALT CAMERAS

Cobalt cameras are now available capable of containing 3000 curies of high specific activity enabling a beam of radiation approximating 2700 R.H.M. The focal spot available is one cm. The work at the author's company's foundries is accomplished with this focal spot size. The camera is versatile in its mounting, allowing all of the necessary adjustments to accomplish the right direction of the beam to produce the needed radiograph.

The cobalt camera is remote controlled by means of a console incorporating automatic exposure timing features, and three separate safety controls to insure that the source is properly safed. These safety controls assure that the source will be returned to the safe position should unauthorized entry be made into the exposure area, or should there be any loss of control from the console.

At the company's foundries, radiography of stainless steel castings for nuclear valve parts is performed by 1000 curies of cobalt giving a beam of approximately 1100 R.H.M. Some typical exposure times with this cobalt source using sensitive film to a density of 2.0 at a 60 in. source to film distance are:

1) 2 in. steel – 1.2 min.
2) 5½-in. steel–8 min.
3) 7 in. steel – 22 min.
4) 10 in. steel – 175 min at a 36 in. source to film distance.

The radiography program meets MIL Spec 271A with qualification from the Bureau of Ships for the range of 1½ to 10 in. of steel.

## CLASS ONE RADIOGRAPHY

Experience at the author's company has been an achievement of Class 1 radiography of steel sections above 3 in., due to the high output of the cobalt source. It might be noted that if the present source were replaced with the high specific activity cobalt now available exposure times could be cut by a factor of 2.5. The technique requires 0.010 in. lead screens front and back with 0.030 in. front filter and the cassette backed up by ½-in. of lead in cases where the scatter is great.

While the film interpretation is more difficult with the lessened contract of cobalt, this is more than offset by the wide latitude gained in the use of cobalt as a radiographic source. This wide latitude allows interpretation of wide variations in thickness, as is encountered in the typical foundry product without the loss of sensitivity in these varying thicknesses.

In summary, the findings at the author's company enable the company to state that kilocurie cobalt source material have definitely met the requirements of the stringent quality control on heavy section castings without resorting to a high cost x-ray installation whose maintenance about a foundry is difficult.

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# HOW MUCH SUPERHEAT IS LOST IN THE RUNNER?

by J. W. Hlinka, V. Paschkis and F. S. Puhr


## ABSTRACT

A runner can be thought of as a thick walled channel through which a hot liquid flows. There is a formula that allows estimation of, under certain conditions, how much heat is lost by the metal as it passes through the runner, and what the temperature of the metal will be as it exits. In addition, the formula discloses how the various runner design characteristics influence the loss. Providing all other design characteristics are held fixed, the loss in the runner depends on:

1. The length of the runner. Increasing the length of the runner by some factor will increase the temperature loss by the same factor.
2. The perimeter of the runner. Increasing the perimeter of the runner by some factor increases the temperature loss by the same factor (providing the flow rate stays the same).
3. The flow rate (lb/min of metal flowing in the runner). Increasing the flow rate by a factor decreases the temperature loss by the same factor.
4. The critical design characteristic. A most convenient form of the above items results if we multiply the length (item 1) by the perimeter (item 2) to obtain the surface area of the runner. Dividing the surface area by the flow rate (item 3) leads to a critical design characteristic of surface area to flow rate. This is the most important factor which governs the temperature loss for a given metal and mold material. Increasing the critical characteristic by a factor increases the temperature loss by the same factor.
5. Comparing two metals, the one having a specific heat larger than the other. The metal with the larger specific heat will have a smaller temperature loss. The specific heat is the most important thermal property of the metal.
6. The temperature loss depends on the square root of the product of conductivity, density and specific heat of the mold material.
7. Nearly the full effect of an increase in superheat is realized in the mold.
8. The temperature of the metal leaving the runner increases during the time the mold fills. In other words, the temperature loss decreases with time. More precisely, the temperature loss decreases inversely with the square root of time.
9. While it is highly probable that some freezing takes place along the walls of most runners, this factor is disregarded.


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## INTRODUCTION

A large body of research in foundries has been devoted to solidification problems, including both experimental and mathematical studies. Generally, such work dealt only with the freezing in the mold cavity, and even the filling time of the cavity was usually disregarded. The present study deals with the temperature loss in runners, and should lead to a better understanding of the design requirements of runners, sprues, etc., in short, any opening through which hot metal is flowing.

In this study some considerations are presented which were made in anticipation of a more thorough examination, together with preliminary estimates, using a simple formula, that can be found in books on heat transfer[1] and here in the Appendix. This formula predicts the temperature loss (heat loss) in a runner and indicates which factors dominate the loss and in what way; therefore it has practical value.

## FACTORS CONTROLLING TEMPERATURE LOSS

As usual in an estimate, refinements are dropped and only items which appear to be important are considered. Factors that affect the temperature loss in a runner are:

1. Kind of metal being poured.
2. Kind of mold material surrounding the runner.
3. Runner shape, perimeter and length.
4. Flow rate of the liquid metal.
5. Pouring temperature.
6. Mold temperature before pouring.
7. Time after the runner is filled.

Figure 1 is a schematic of a runner system. The numbers relate to the seven factors listed.

Regarding item 7, the runner is at first cold, and the superheat lost in the runner is large. Later on, when the runner walls are warmed, the superheat lost will be smaller. Therefore, the temperature of the metal leaving the runner increases with time. This

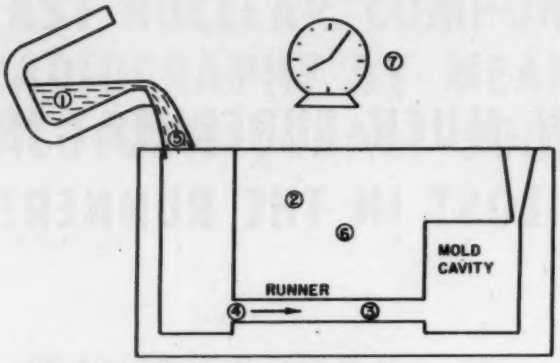



Fig. 1 — Simplified gate-mold system.

is important because to show the temperature of the metal entering the mold, the time this temperature exists must be shown. In this report, time is always measured from the start of filling and not the start of pouring, i.e., time zero, when the counting starts, is the time when the metal first leaves the runner and enters the mold cavity.

Of course, there are other factors which will affect the temperature loss, such as the skimming gate, which will influence not only the flow rate, but the kind of flow, the heat transfer coefficient developed at the interface between the metal and runner, the fact that the thermal properties of both runner and metal change with temperature, and, of greatest importance, the freezing and remelting which can take place. These and other factors were neglected in this preliminary study.

## APPROXIMATIONS TO FIT FORMULA

A mathematical statement of the problem is given in the Appendix for those who prefer equations, but here the above list of items and the assumptions which were made are surveyed so that the formula can be applied to a runner.

**Kind of metal being poured.** The way the problem is stated in the Appendix, regarding how heat is transferred in the metal, the specific heat of the liquid metal is the only physical property which appears. Because of the turbulence in flowing metal, differences in the thermal conductivity of different metals are ignored. Freezing and remelting are disregarded, if any takes place, and thus the latent heat is also ignored. The density enters into the flow rate.

**Kind of mold material surrounding the runner.** It is assumed that the thermal properties of the runner material are not affected by changing temperatures, but rather stay constant during the filling time.

**Runner shape, perimeter and length.** The shape (i.e., trapezoidal, square, etc.) of the cross-section of the runner is disregarded, and only the product of the perimeter and the length is pertinent. This product, perimeter multiplied by length, is the surface area of the runner in contact with the liquid metal.

**Flow rate of the liquid metal (lb/min).** It is assumed that the metal flows through the runner at a constant flow rate. This is not true, because the pressure head in the sprue varies, at least when it is being filled. Also, a back pressure develops as the mold fills, which creates a varying velocity in the runner and thus a varying flow rate. However, an average flow rate is easily determined. Take the total weight of the metal that leaves the runner and divide this weight by the estimated filling time.

**Pouring temperature.** The temperature of the metal entering the runner is the same as the pouring temperature, and it does not vary during the filling time.

**Mold temperature before pouring.** The formula requires that, before pouring, the temperature of the mold material surrounding the runner is the same throughout. But this temperature need not be the shop temperature, since the formula applies to heated molds.

## DETERMINING RUNNER TEMPERATURE LOSS

If the preceeding conditions are accepted, then the formula that may be applied to determine the temperature loss in a runner is:

$$\text{Temp. Loss} = \text{Constant} \times \frac{\text{Surface Area}}{\text{Flow Rate}} \times \frac{(\text{Pouring Temp.} - \text{Mold Temp.})}{\sqrt{\text{Time}}} \qquad (1)$$

It is convenient to express the temperature loss in percentage form. While usually the per cent temperature loss uses the pouring temperature as a reference or base point, the difference between the pouring temperature and the mold temperature will be used here. This is an improvement, because for metals having low melting temperatures the effect of the mold temperature cannot be ignored. Also, using the difference as a base makes it possible to determine the effect of heated molds. Thus:

$$\text{Per cent Temp. Loss} = \frac{\text{Temp. Loss}}{(\text{Pouring Temp.} - \text{Mold Temp.})} \times 100$$

$$= \text{Constant} \times \frac{\text{Surface Area}}{\text{Flow Rate}} \times \frac{100}{\sqrt{\text{Time}}} \qquad (2)$$

Where:

Temperature Loss is defined as the difference between the pouring temperature and the temperature of the metal leaving the runner (in degrees Fahr.).

The bracket term Pouring Temp. – Mold Temp. means the pouring temperature minus the mold temperature before pouring. For instance, if we pour at 1300 F into a mold at 100 F, the Pouring Temp. – Mold Temp. term in the formula would have a value of 1200 (use degrees Fahr. only).

The Per cent Temperature Loss, as defined by equation two, is the temperature loss divided by the Pouring Temp. – Mold Temp. and multiplied by 100. A per cent temperature loss in the runner of 5 per cent means 0.05 times the difference between pouring and mold temperatures. As an example, if the mold is at a shop temperature of 80 F,

and the pouring temperature is 1280 F, then the temperature loss in the runner would be 0.05 times 1200, or 60 F.

The Constant in the equations depends on the specific heat of the metal being poured and the runner material, and is given in Fig. 2 for sand runners only.

The Surface Area of the runner is the perimeter multiplied by the length (use in.).

The Flow Rate is the weight of the casting, including risers, etc., but not the gate assembly, divided by the estimated filling time (use lb for weight and min for time).

Time in the equation is defined as the time after filling begins (expressed in min).

This equation, discussed in detail in the Appendix, shows how the temperature loss of the liquid metal depends on the various factors. To what extent this simple equation will check with practice can be ascertained only when experimental data become available. The authors, however, have confidence that within practical limits, the results are useful.

There is a restriction which must be made when the equation is used. The equation is valid for per cent temperature losses not larger than 25 per cent and not smaller than 0.1 per cent. This is a wide range, since it is unlikely that 25 per cent of the pouring temperature would be lost in the runner, and losses of less than 0.1 per cent are not significant. Furthermore, the formula applies only when the temperature loss is less than the amount of superheat.

## REMARKS

A conclusion which can be drawn from the formula is that when a given metal is being cast, the pouring temperature does not greatly affect the temperature loss in a runner. That this conclusion is correct can be reasoned from both heat transfer considerations and from mathematics. When two bodies, having different temperatures, are brought into direct contact, such as exists between the liquid metal and the mold, the amount of heat lost from the hot body to the cold body depends on the temperature difference between the bodies.

Therefore, if as an example, the pouring and mold temperatures are increased by 10 per cent, both the heat loss and the temperature loss in the runner will be increased by 10 per cent. Now, a 10 per cent increase in temperature may not seem like a large number, but 10 per cent of the pouring temperature when aluminum is being cast represents an additional superheat of about 120 F, which is quite a bit. Also, since all temperatures are increased by 10 per cent, the temperature loss will be increased by 10 per cent. Therefore in the example, the increase in the temperature loss would be 10 per cent of 120 F, or 12 F.

This means that of the 120 F additional superheat, 108 F finds its way into the mold cavity. For steel with a pouring temperature of 2800 F, a 10 per cent increase in superheat would be 280 F, of which 28 F would be an added loss in the runner, while 252 F will find their way into the mold cavity. The conclusion is that almost the entire increase in superheat finds its way into the mold cavity. The statement that temperature losses are proportional to the temperature difference between the metal and mold is not limited to the runner, but holds also for the gate assembly, except in regions where heat is lost by radiation and before freezing starts.

The equation arrives at the same conclusion, since it states that the temperature loss in the runner is directly proportional to the temperature difference between the pouring temperature and the mold temperature before pouring. Then the same argument holds. If this difference is increased by a certain percentage, then the temperature loss in the runner will at all times be higher by the same percentage, provided all the other factors remain unchanged.

In practice, the small effect of varying the superheat on the temperature loss would be unobservable, because the pouring temperatures could be varied

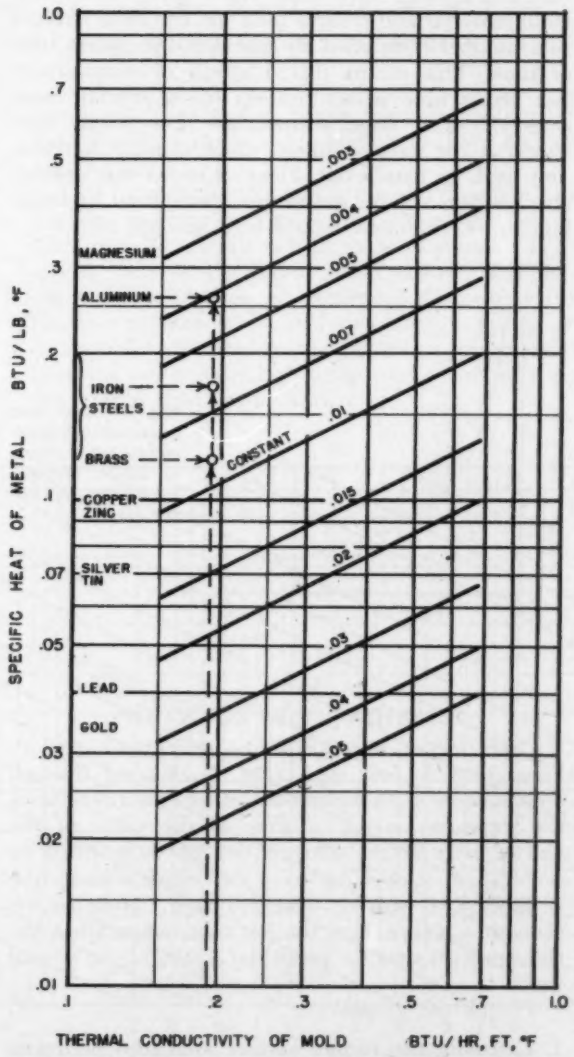



Fig. 2 — Constant for common metals flowing in sand runners to be inserted into either equation (1) or (2).

only within a fairly narrow range, unless special equipment were built.

*Critical Ratio*

It is quite interesting to note in the equation that for a given metal and mold material the per cent temperature loss, at a given time, depends only on the surface area of the runner divided by the flow rate. Since this combination seems to be the most important factor in runner design, the ratio of surface area to flow rate will hereafter be called the critical ratio.

The formula states that the per cent temperature loss has a one-to-one relationship with the critical ratio, which means that if the ratio is multiplied by a factor, the per cent temperature loss will change by the same factor. For example, if the ratio is doubled, the loss would be doubled. It is useful to know that the temperature loss is directly proportional to the surface area of the runner, and inversely proportional to the flow rate.

Another point of interest is the way the temperature loss changes with time. For a given metal and mold material and critical ratio the equation reduces to a number multiplied by one over the square root of time. This means that a graph of temperature loss versus time in all runners has a similar characteristic of a large temperature loss (large heat losses) at the start of filling, which rapidly decreases and tends to flatten out. Figure 3 shows this general characteristic. If this graph were plotted on logarithmic paper, the curve would be a straight line.

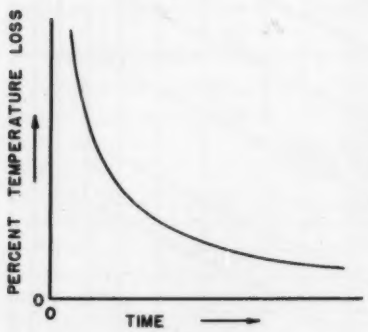



**Fig. 3 — Per cent temperature loss for metals has a general shape, decreasing during the filling time.**

## GRAPHING THE EQUATION

Three metals were graphed — aluminum (Fig. 4), brass (Fig. 5) and steel (Fig. 6), flowing through sand runners. Each figure may be used for any alloy of the respective metal, as long as the value of the specific heat of the alloy is one of those listed in Table 2. If it does not have this value, a correction is applied, as will be discussed later. These figures show at a glance, how the per cent temperature loss changes with time for particular values of the critical ratio.

To use the graphs:

1. Calculate the runner surface area (the perimeter multiplied by the length).
2. Calculate the flow rate (weight of casting, including risers, divided by the estimated filling time).
3. Divide the surface area by the flow rate (this number is the critical ratio).
4. Refer to the figure corresponding to the metal being cast, and locate the line corresponding to the critical ratio (this line is the percentage temperature loss versus time for this runner).
5. To convert the percentage into degrees temperature loss, multiply the per cent temperature loss by the difference between the pouring temperature and the mold temperature, and divide by 100.

An example using aluminum (Fig. 4) is: Take a runner having a perimeter of 4 in. and a length of 10 in. The surface area is then 4 times 10, or 40 sq. in. The weight of the casting, including risers, is 40 lb, and the expected filling time is 4 min. This results in an average flow rate of 10 lb/min. Then 40 sq in. divided by 10 lb/min yields a critical ratio of 4. Refer to Fig. 4 for aluminum flowing through a sand runner. The dashed line shows the per cent temperature loss during the time of filling.

If in this example, the pouring temperature is 1400 F and the mold temperature before pouring is 100 F, the percentage temperature loss is multiplied by 1300 and divided by 100 to obtain the temperature loss in degrees Fahr. The factor 100 is used to reduce the percentage to a fraction. These results are shown in Table 1 for several values of time as taken from Fig. 4.

**TABLE 1**

| Time, min | Temp. Loss, % | Temp. Loss, F, |
|---|---|---|
| 0.1 | 5.0 | 65 |
| 0.2 | 3.6 | 47 |
| 0.5 | 2.3 | 30 |
| 1.0 | 1.6 | 21 |
| 2.0 | 1.2 | 16 |
| 4.0 | 0.8 | 10 |

In Figs. 4, 5 and 6, the physical properties of metal and sand listed in Table 2 were used. However, the data from these figures can be corrected to account for different kinds of sand. Figure 7 gives a correction factor, based on the thermal conductivity of the sand being used. To obtain the temperature loss for a particular sand, multiply the temperature loss, as given in Figs. 4, 5 and 6 by the correction factor,

**TABLE 2**

| Sand Properties (from R. W. Ruddle[2]) | |
|---|---|
| Thermal conductivity, Btu/hr, ft, F | 0.191 |
| Density, lb/ft³, F | 91.0 |
| Specific heat, Btu/lb, F | 0.254 |

| Metal | Specific Heat, Btu/lb, F | Constant, see Fig. 2 |
|---|---|---|
| Aluminum | 0.26 | 0.00406 |
| Brass | 0.12 | 0.00880 |
| Steel | 0.17 | 0.00620 |

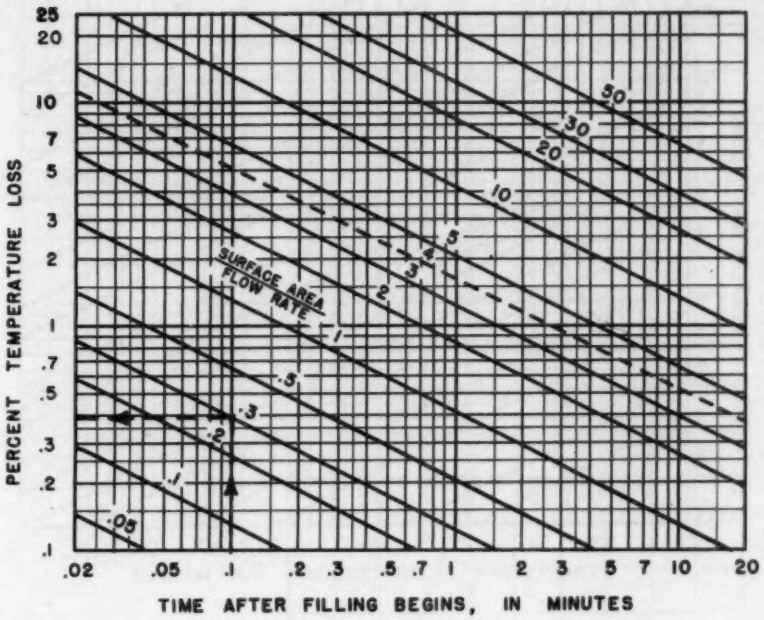



Fig. 4 — Per cent temperature loss vs. time for fixed values (marked on the curves) of the critical ratio of surface area (sq in.) to flow rate (lb/min). This figure applies to aluminum and its alloys which have a specific heat of 0.260 in sand runners (conductivity 0.191).

as given in Fig. 7, corresponding to the conductivity of the sand being used.

Another way to account for a different sand is to determine the property constant from Fig. 2 and use equation (1) or (2) directly.

## EXTENDING GRAPHS TO OTHER METALS

Comparison of Figs. 4, 5 and 6 shows that the per cent temperature loss is inversely proportional to the specific heat of the metal. In other words, a metal having a specific heat two times larger than another metal will have half as much per cent temperature loss, providing the critical ratio and runner material are the same.

This means that the results for aluminum in sand runners, as given by Figs. 4 and 7, may be used as a base for other metals flowing in sand runners, if the specific heat of the liquid metal is known. This is the major advantage of expressing the temperature loss as a percentage, because then steel, copper, lead, etc., which are poured at vastly different temperatures, have the same percentage temperature loss scale and new graphs need not be drawn.

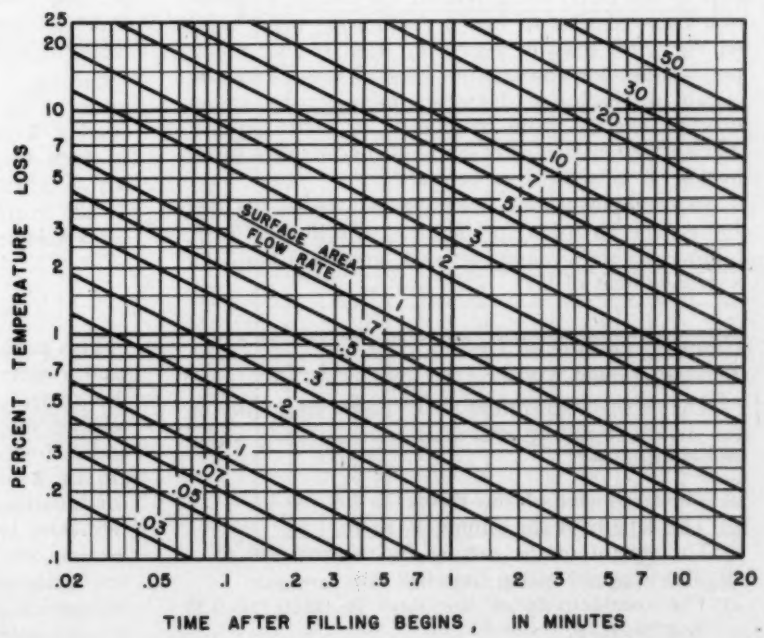



Fig. 5 — Per cent temperature loss vs. time for fixed values (marked on the curves) of the critical ratio of surface area (sq in.) to flow rate (lb/min). This figure applies to brasses which have a specific heat of 0.12 in sand runners (conductivity 0.191).

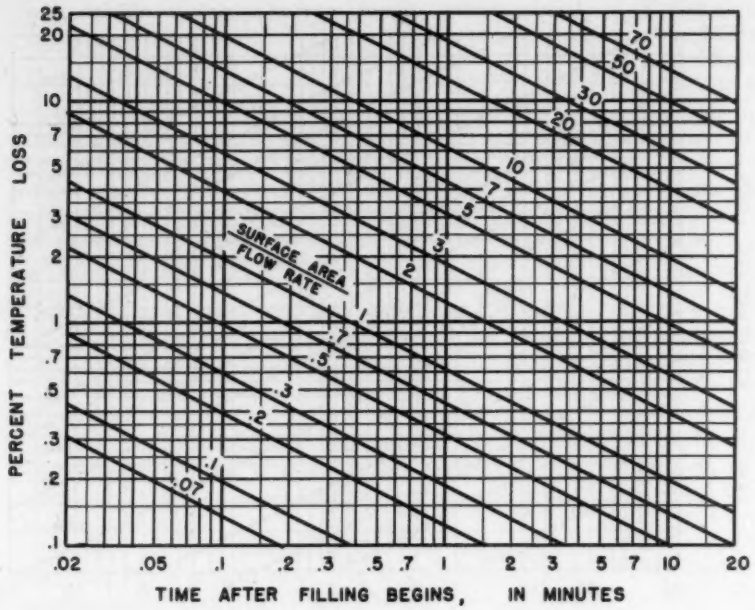



**Fig. 6 — Per cent temperature loss vs. time for fixed values (marked on the curves) of the critical ratio of surface area (sq in.) to flow rate (lb/min). This figure applies to steels which have a specific heat of 0.17 in sand runners (conductivity 0.191).**

For example, to obtain Fig. 5 for brass by correcting the data of Fig. 4 for aluminum, multiply the aluminum results by the specific heat of aluminum (0.26) and divide by the specific heat of brass (0.12). This gives a correction factor of 2.16. Note that all of the per cent temperature loss data in Fig. 5 for brass are 2.16 times higher than that for aluminum (Fig. 4) for the same value of the critical ratio.

Figure 8 is a plot of correction factors which may be used to determine the temperature loss for most metals in sand runners. To obtain the temperature loss in a given metal, multiply the temperature loss, as given by Fig. 4, for aluminum by the correction factor given in Fig. 8, corresponding to the specific heat of the metal being poured. Of course, a correction curve similar to Fig. 8 could have been based on Fig. 5 for brass or Fig. 6 for steel. In Fig. 8 ball park values for the specific heat of some common metals are indicated. However, more precise values may be selected from handbooks or other sources. The correction factor for the sand conductivity, given by Fig. 7, is applied as before.

An example of how these figures are to be used for a steel, having a specific heat of 0.15 Btu/lb, F, follows. Figure 6 cannot be used, since it is based on a specific heat of 0.17.

### *Example: Pouring of Steel Through a Sand Runner*

What is the temperature of the liquid steel entering the mold cavity 0.10 min after start of filling under the conditions:

a) The perimeter of the runner is 2.5 in.
b) The length of the runner is 30 in.
c) The weight of the casting and riser is 500 lb.
d) The expected filling time is 2 min.
e) The conductivity of the sand is taken as 0.35 Btu/hr, ft, F.
f) The specific heat of the steel is taken as 0.15 Btu/lb, F.
g) The pouring temperature is 3100 F, and the mold temperature is 100 F.

From (a) and (b) multiply the perimeter by the length to obtain the surface area. This gives a surface area of 75 sq in. From (c) and (d) obtain the flow rate, which is the weight divided by the filling time. This gives a flow rate of 250 lb/min. Divide the surface area from (1) by the flow rate from (2) and obtain a critical ratio of 0.30. On Fig. 4 for aluminum, look up, for the time of 0.1 min and a critical ratio of 0.30, the per cent temperature loss, which is 0.39 per cent. When multiplied by, 3100 F minus 100 F (from g), and divided by 100, this yields an uncorrected temperature loss of 11.7 F.

On Fig. 7, look up, for the sand conductivity as given by (e), the first correction factor, which is 1.41. On Fig. 8, look up, for the specific heat of the steel as given by (f), the second correction factor, which is 1.75. Multiply the 11.7 F temperature loss from (4) by the correction factors 1.41 (5) and 1.75 (6), to get a temperature loss for steel of 28.9 F.

## CONCLUSION

This presentation was limited to metals flowing in sand runners. However, the formula used to draw the charts can be applied, with some reservations, to other molds, and a similar chart can be drawn for each. However, to avoid the drawing of additional graphs, a formula is provided which includes cross combinations of all common metals with common refractory mold materials, superheat, preheat, points in time, etc. However, since symbols and dimensionless ratios are involved in the discussion of the development of the formula, it is included in the Appendix only.

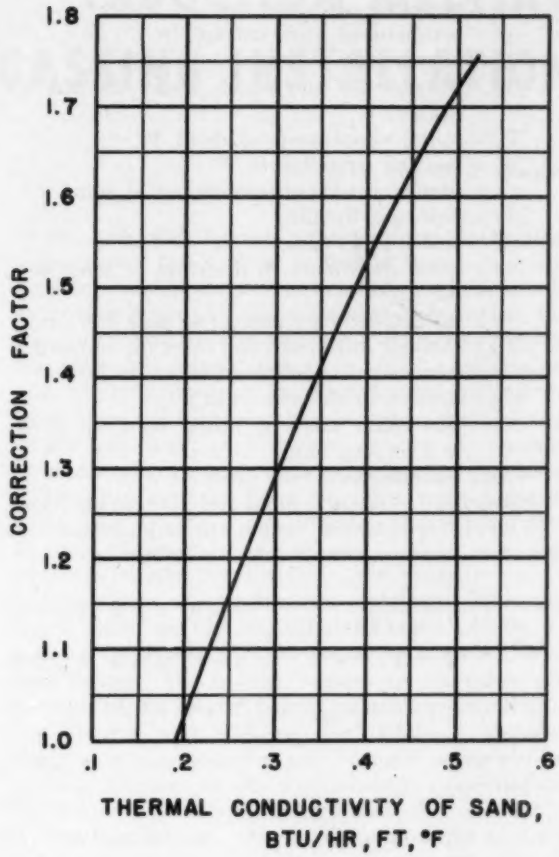



Fig. 7 — Correction factor for sands having thermal conductivities other than 0.191. Apply these corrections to Figs. 4, 5 and 6.

It is interesting to note that the formula predicts some freezing will usually take place in all runners. At the start of filling, the formula says that the temperature of the metal entering the cavity is far below the melting point, but this is impossible, because there would be plugging. The formula, then, is not applicable at the start of filling of the mold. It indicates that a frozen layer forms along the walls, which is later remelted by the upstream metal.

If the upstream metal is not hot enough to remelt this layer, the frozen thickness will usually grow and plugging takes place. Since it takes time for this layer to grow, it may be that plugging will not take place until after the cavity is partially filled. This process is quite complicated[3] and is, of course, unaccounted for in this paper. However it should be kept in mind, since the freezing and remelting will become an important factor in answering the question of whether the runner will plug.

## REFERENCES

1. H. S. Carslaw and J. C. Jaeger, *Conduction in Heat Solids*, second edition, Oxford at the Clarendon Press, p. 396 (1959).
2. R. W. Ruddle, *The Solidification of Castings*, second edition, The Institute of Metals, p. 380 (sand C at 677 C) (1957).
3. "Analysis of Fluidity of Metals," by J. Niesse, M. Flemings, H. F. Taylor. Report published as part of Doctoral Requirements, Massachusetts Institute of Technology.

## APPENDIX

$A_c$ = cross-sectional area, in.$^2$ (sq. in).
$A_s$ = surface sectional area, in.$^2$ (sq. in).
c = specific heat, Btu/lb, F.
k = thermal conductivity, Btu/min, in., F.

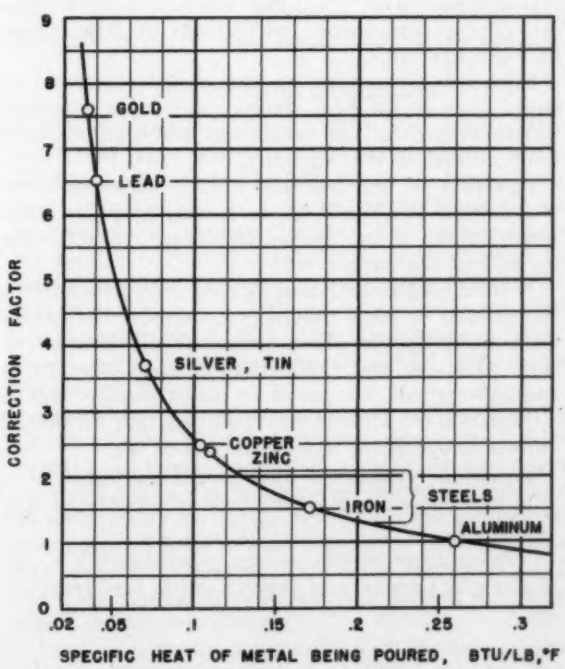



Fig. 8 — Correction factor for any metal or alloy having a specific heat other than 0.260. Apply these corrections to Fig. 4 only.

$L$ = length of runner, in.
$p$ = perimeter of runner, in.
$p'$ = perimeter of plane surface, in.
$T$ = temperature, F.
$T_e$ = temperature of liquid metal at exit of runner, F.
$T_m$ = initial temperature of mold, F.
$T_p$ = pouring temperature, F.
$t$ = time after start of pouring (see $\tau$), min.
$W$ = flow rate, lb/min.
$w$ = velocity, in./min.
$x$ = space coordinate in direction of mass flow, in.
$y$ = space coordinate normal to mass flow, in.
$\alpha$ = thermal diffusivity, in.$^2$/min (sq in./min).
$\theta$ = dimensionless temperature loss ratio.
$\rho$ = density, lb/in.$^3$ (lb/cu in.).
$\tau$ = time after start of filling of mold cavity $= (t - L/w)$, min.
$\phi$ = dimensionless time ratio.

Subscripts 1 – refers to mold and 2 – refers to metal.

The system equations which approximate the temperature loss in a runner may be written as

$$\frac{\delta^2 T_1}{\delta y^2} = \frac{1}{\alpha}\frac{\delta T_1}{\delta t} \qquad y>0 \quad x>0 \quad t>0 \quad (1)$$

$$T_1 = T_2 \qquad y=0 \quad x>0 \quad t>0 \quad (2)$$

$$p' k_1 \frac{\delta T_1}{\delta y} = \rho_2 c_2 A_c \frac{\delta T_2}{\delta t} + W c_2 \frac{\delta T_2}{\delta x} \qquad y=0 \quad x>0 \quad t>0 \quad (3)$$

$$T_1 = T_m \qquad y>0 \quad x>0 \quad t=0 \quad (4)$$

$$T_2 = T_p \qquad x=0 \quad t>0 \quad (5)$$

These equations describe a physical system wherein a hot fluid flows past the surface of a semi-infinite region bounded by $y = 0$ and $x = 0$. The coefficients are all constant, being independent of time, space and temperature.

Equation (1) applies to the mold material and stipulates that the heat conduction in this region is uni-directional in the y-direction and that no heat flows normal to that direction.

Equation (2) assumes that heat is perfectly transferred from the liquid to the runner wall. In other words, there is no contact resistance between the metal and the runner wall.

It is implied in equation (3) that the turbulence in the fluid is so large that no temperature differences exist in the fluid in planes normal to the direction of flow. Also, the heat conduction in the x-direction is negligibly small, compared to the transport term.

Equation (4) imposes the condition that the initial runner temperature is uniform and at $T_m$.

Equation (5) specifies that at the entrance (at $x = 0$), the temperature of the fluid is maintained constant at $T_p$.

The solution to this system, equations (1) to (5), is given by Carslaw and Jaeger[1] and is here rewritten as:

$$T_1 = (T_p - T_m)\,\mathrm{erfc}\left\{\frac{k_1 p' x + y W c_2}{2 W c_2 [\alpha_1 (t - x/w)]^{1/2}}\right\} + T_m \qquad (6)$$

$$T_2 = (T_p - T_m)\,\mathrm{erfc}\left\{\frac{p' x}{2 W c_2}\left[\frac{(k\rho c)_1}{t - x/w}\right]^{1/2}\right\} + T_m \qquad (7)$$

$(t - L/w)$ is replaced by $\tau$, since the prime concern is the temperature loss in the liquid between $x = 0$ and $x = L$, and equation (7) becomes, at $x = L$:

$$\frac{T_p - T_e}{T_p - T_m} = \theta = \mathrm{erf}\left\{\frac{p' L}{2 W c_2}\left[\frac{(k\rho c)_1}{\tau}\right]^{1/2}\right\} \qquad (8)$$

The assumption is now made that equation (8) applies to a closed channel, such as the runner, simply by replacing the perimeter $p'$ by $p$. Now, since $p\,L$ is the surface area of the channel, equation (8) may be written as:

$$\theta = \mathrm{erf}\left\{\frac{A_s}{2 W c_2}\left[\frac{(k\rho c)_1}{\tau}\right]^{1/2}\right\} \qquad (9)$$

For brevity, let:

$$\phi = \frac{A_s}{2 W c_2}\left[\frac{(k\rho c)_1}{\tau}\right]^{1/2} \qquad (10)$$

then:

$$\theta = \mathrm{erf}\,\phi = f(\phi) \qquad (11)$$

which clearly shows that the dimensionless temperature loss is a function of one dimensionless time variable only. Therefore, the dimensionless temperature loss $\theta$ can be expressed graphically as a single curve, using $\phi$ as a dimensionless time variable.

Practical considerations lead to limits for $\theta$. If the value of $\theta$ exceeded 0.25, unrealistically high losses of superheat would be implied. Values of $\theta$ below 0.001 would mean, in case of aluminum, a temperature loss in the runner of about one F, and in case of steel of about 3 F. Temperature losses below these values are practically insignificant. Within these limits one can write, with excellent approximation:

$$\theta = 1.12\,\phi \qquad 0.001 \le \theta \le 0.25 \qquad (12)$$

Then equation (9) may be written:

$$\theta = K\left[\frac{A_s}{W}\right]\frac{1}{\sqrt{\tau}} \qquad 0.001 \le \theta \le 0.25 \qquad (13)$$

where the property constant $K$ is given by:

$$K = 0.56\,\frac{\sqrt{(k\rho c)_1}}{c_2} \qquad (14)$$

Figure 2 is a plot of equation (14), using a density of 91 lb/cu ft, and a specific heat of 0.254 Btu/lb, F.

Of course, any numerical values for the parameters may be inserted into equation (13), corresponding to any combination of mold and metal, runner geometry, etc. This equation should apply well for all common metals in sand runners, but for metal-walled runners, the effect of the boundary conductance as well as freezing and remelting may perhaps not be negligible.

If in equation (13) the mass rate of flow $W$ is replaced by the product of velocity $w$, the cross-sectional area $A_c$ and the density $\rho$, one finds that for any given metal and mold, the temperature loss at all times $\tau$ is proportional to the ratio of surface to cross-sectional area, and inversely proportional to the linear velocity.

# INDUCTION MELTING AND DEGASSING INSTALLATION

by O. E. Erickson


## ABSTRACT

**The details of the induction vacuum equipment at the author's company are described, and the benefits received in the details of making a heat are given.**


## INTRODUCTION

The facilities that were available at the author's company prior to the addition of the induction melting and degassing furnaces were two 8000 lb and one 3000 lb acid-lined arc furnaces. These were utilized for production of stainless steel, carbon steel and other low alloys. This foundry, like many others, is faced with the problems of processing a high proportion of revert materials, and also of providing a means of processing small heats of special analyses.

These two requirements led to an extensive investigation of the commercial melting equipment available, and to the eventual installation of the induction melting and degassing equipment. With this installation the company has been able to realize these advantages:

1) For over one year stainless steel has been produced with a gas threshhold level below that required to produce castings free from pinhole porosity when cast in green sand molds. This has been effected without inordinate control over mold moisture content, pouring temperature or the use of special deoxidizing materials such as selenium.
2) A second major advantage that came with the use of the induction furnace has been the elimination of the need for wash heats to clean up furnaces to prevent contamination of subsequent heats. With the induction furnaces, various furnace shells are utilized for each major category of melt chemistry produced.
3) A third advantage results from the ability to produce heats consisting of 100 per cent revert.

The accumulative affects of these three advantages have resulted in a reduction of heats out of specification and losses attributed to high gas content, and have provided final products more acceptable to the ultimate user.

This induction melting vacuum degassing installation consists of a 300 kw induction melting unit with two one ton furnaces arranged and modified so that a vacuum degassing chamber may be placed over either furnace. The induction unit, rated 300 kw, is the recently introduced static frequency converter which, by means of a certain type of transformer design, converts 3-phase, 60-cycle power to 180 cycle single phase of about 750 volts.

The converter consists of a cubicle containing a power unit with 60-cycle capacitors for maintaining a line power factor of about 0.90 or better. A second cubicle contains the furnace control, and consists of the usual capacitors for resonating the coreless furnaces and auxiliary components. Two meters are used for observing applied power. A set of push buttons and rotary switch are provided for controlling the power into the furnace. A two-position selector switch in the base of the control panel permits the operator to select one of two furnace positions for operation.

Two furnace stands are provided, each adequate to support a furnace case of sufficient size for a 3000 lb melt. However, at present the furnaces have coils of size to handle normal 2000 to 2400 lb melts.

These furnaces are mounted on special fabricated bases, which are dish shaped and flanged to form the base of the closed vacuum degassing chamber when used as such. The furnace stations are located in pits in front of the control panel to facilitate handling of the charges and the vacuum degassing chamber (Fig. 1). This equipment can be used as normal air melting furnaces at any time without involving the use of any vacuum equipment, if desired.

The vacuum degassing equipment is essentially a large steel chamber or bell jar lowered over the air melting furnace and sealed to the specially fabricated base plate (Figs. 2 and 3). Proper manifolding, valves and a pump, permit application of vacuum to either furnace station as may be desired by the operator.

## DEGASSING CHAMBER

The vacuum degassing chamber is a plain carbon steel tank approximately 3/8-in. in thickness with a double walled water cooled dome. A large O-ring seal is maintained in a groove in the flange on the bottom of the degassing chamber. This insures a good vacuum seal when the chamber is lowered on to the base plate. This degassing chamber is approximately 8 ft in diameter, 8 ft high and is moved by means of an overhead crane. The base also contains sealed power ports for entrance of the power leads to the vacuum ports.

When the vacuum equipment is not used, a protective guard is placed over the flange of the base to prevent any damage while air melting. Pneumatically operated valves are included to enable the operator to select either station as desired for the application of vacuum. A filter or trap is included in the vacuum line in order to prevent metallic or refractory particles from impinging on the moving parts of the pump which would cause damage.

Vacuum is obtained by use of an 850 cfm mechanical pump driven by a 40 hp motor. Theoretically, with a clean dry system, this pump would be capable of

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reducing the pressure to 200-500 microns in about 10 to 15 min.

For indications of vacuum conditions in the processing chamber, or for independent performance checking of the mechanical pump, a two-station automatic gage is included. The stations are located so that chamber pressure and pump performance may be monitored. This gage has an indicating range from 0 to 100 mm mercury.

The furnaces are lined with a high MgO contact refractory and the material is rammed dry using a transite form. The first heat is held 20 min at 3150 F (1732 C) to sinter the lining. It is then cooled back to 2900 F (1593 C), vacuum boiled and used as a production heat. The furnaces average 50 heats per lining when melting stainless at 3150 F (1732 C).



Fig. 1 — The furnace stations are located in pits in front of the control panel to facilitate handling of the charges and the vacuum degassing chamber.

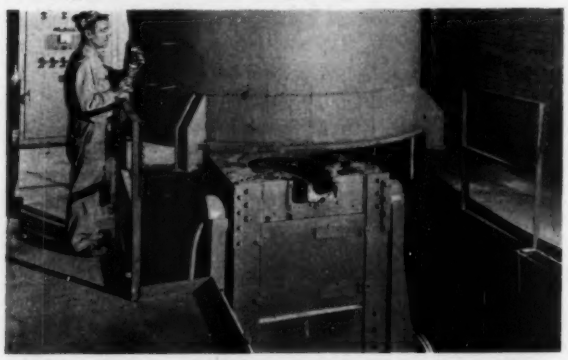


Fig. 2 — The vacuum degassing equipment, essentially a large steel chamber, is lowered over the air melting furnace.

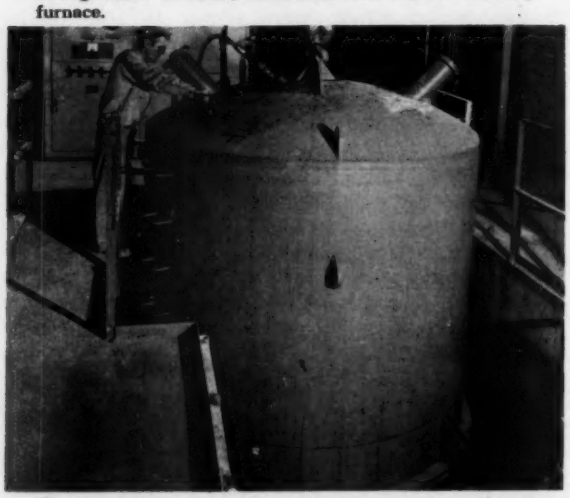


Fig. 3 — When the vacuum degassing equipment has been lowered into place, it is sealed to the specially fabricated base plate.

The making of a heat consists of melting, adjusting the carbon to allow for the removal of 0.06 to 0.12 per cent carbon, heating to 2950 F (1621 C), vacuum degassing, heating to 3150 F (1732 C), adjusting the silicon and manganese and tapping into a teapot ladle.

The charge may consist of 100 per cent revert material such as the 11.5 to 14.00 per cent chrome grade known as A.C.I. CA 15. After a meltdown an analysis is taken, and carbon added to give a bath with approximately 0.20 per cent carbon.

When a temperature of 2950 F (1621 C) is reached, the bell is lowered over the furnace and the vacuum pump started. The air admittance valve is fully open, and the operator controls the height of the boil by opening or closing this valve. The time consumed will be from five to 12 min, depending on the melt in carbon and the carbon required by the specification.

After this period the vacuum pump is shut down, the air admittance valve opened, and when atmospheric pressure is reached the bell removed. A sample is then taken to determine the carbon and chromium, and the necessary additions of these elements are made plus 0.50 per cent silicon and 0.50 per cent manganese. The melt is superheated to 3150 F (1732 C), 0.10 per cent aluminum is added and the furnace tapped.

The next heat is made faster by pouring 200 to 300 lb of metal back into the furnace to serve as a molten starter. The procedure for the austenitic and other grades of stainless is exactly the same except no aluminum is used.

## CONCLUSION

No figures are available regarding the exact amount of hydrogen, nitrogen or oxygen removed by this process. Test bar results indicate that the tensile and ductility are no better than a well made arc furnace heat. However, the average ductility is better, indicating that it is perhaps easier to make a top quality stainless heat by vacuum degassing than in the arc furnace.

The equipment can also be used for melting under an inert atmosphere such as argon. Consideration is also being given to pouring in an argon envelope and using argon for purging the mold.

Heats are also being made by dead melting in the arc furnace, pouring into the induction and vacuum boiling. Metal such as the CA 15 (11 to 13 chrome grade) dead melted under the arc would always be gassy, probably from hydrogen pickup and nitrogen fixation under the arc. After vacuum boiling the metal is never gassy.

Over 1,000 heats have been made using the vacuum to lower the pressure sufficiently to cause a carbon-oxygen reaction in the presence of high chromium. No castings have shown evidence of pinhole porosity or gassiness. The average physical properties are raised and the efficacious use of revert is accomplished.

This tool has become an aid in efficient production and quality improvement.

# POROSITY IN ZINC DIE CASTINGS

by G. P. Lewis, D. A. Craw and R. C. Bell


## ABSTRACT

**Laboratory methods have been used to study the nature of porosity in commercial zinc die castings. The techniques employed involve primarily density and gas content determinations and gas analysis. By relating the results of these measurements to conditions in the die cavity at the time of solidification it is possible to obtain information on the sources of porosity in the casting tested.**

**The approach described provides a means of differentiating in a quantitative way between the effects of shrinkage and entrapped gas. It also provides information on the origin of the gas and its final distribution in the casting.**

**The results were obtained using these techniques on a number of commercial die castings are described. Included are tests designed to examine the effects of lubrication and injection pressure on the porosity of the casting.**


## INTRODUCTION

The subject of casting porosity is of vital interest to every die caster. Because of the nature of the casting process, pressure die castings contain a relatively high degree of porosity as compared to products of most other casting and fabrication techniques. With proper die design and casting conditions the porosity level is controlled and the porosity is essentially interior; in such cases it is usually of no consequence or concern. If, however, because of faulty or unsuitable die design or improper operating conditions the level of porosity is exceedingly high or the pores extend to the surface layers, then failure or deterioration of the casting in service can result.

Gross porosity, for example, can lead to mechanical failure of the casting under stress. In some instances a degree of pressure tightness is required in the finished part and if the level of porosity is excessive the casting may not meet this requirement. The most frequent problem caused by porosity, however, is concerned with the occurrence of pores in the surface layers of castings which are to be buffed and plated. In such cases the porosity can lead to defects in the electroplated coatings and consequent inferior corrosion resistance in the finished part. Thus, in his efforts to produce consistently high quality die castings, the die caster is constantly concerned with porosity.

Barton,[1] Ruttewit[2] and other workers have considered the way in which die design and operating variables affect the occurrence of porosity from different sources on the basis of practical die casting experience and the existing theories of metal flow during the injection and filling of die cavities. Sachs[3] approached the problem using the laws of mechanics, hydro and thermodynamics, and developed a mathematical expression for the amount of trapped air in a die casting based on the characteristics of the casting machine and the cavity geometry.

In the present work an attempt has been made to determine the sources of porosity in commecial die castings by making laboratory measurements on the castings themselves. The techniques developed have been used to study the porosity in a number of different castings produced in commercial die casting plants using a variety of casting conditions.

## LABORATORY METHODS FOR POROSITY STUDY

A measurement of the level of porosity in a die casting can be readily made by a simple density determination. By comparing the density of the casting or a section of the casting with the theoretical density of the alloy the actual volume of voids can be calculated. The amount of gas in the voids can be determined by melting the casting under vacuum and measuring the gas released. Any gas dissolved in the alloy and released during vacuum melting will be included in this measurement, however, as will be shown later, in the case of zinc alloy castings this contribution is negligible. Information on the origin of the gas can be obtained by analysis.

The data obtained from the above measurements can be used to deduce the conditions existing in the die cavity at the time the alloy solidified. The volume this gas would occupy under the conditions of temperature and pressure in the die cavity just prior to solidification is compared with the volume of voids existing just after solidification.

The latter figure is calculated from density considerations making allowance for the contraction of the casting during cooling to room temperature. The difference between the volume of gas and the volume of voids represents solidification shrinkage. Since some of the solidification shrinkage is taken up by flow of metal into the die during solidification, the difference between the volume of voids and gas has been described as unfed solidification shrinkage.

A vital factor in these calculations and one that is difficult to estimate is the pressure existing on the molten alloy at the time of solidification. In recent work carried out in Japan,[4] the variations in metal pressure during the filling of the die have been measured. It has been found that pronounced variations in pressure occur each time the momentum of the injected metal changes, as for example when the cavity or one section of it becomes filled. The completion of the cavity filling results in a pronounced pressure peak whose magnitude is dependent on the cavity design and the applied pressure.

It has been postulated[1] and confirmed,[4] that asso-

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ciated with this pressure peak is an accession of heat to the extent that the solidified skin of the casting may be remelted. The pressure falls away rapidly from the peak value and reaches a steady value apparently corresponding to the applied pressure at the plunger. This is the value that is desired for the calculation of gas volume, since it describes the condition of the gas just prior to solidification of the bulk of the injected alloy. This figure can be deduced from a knowledge of the pressure applied at the plunger or by measuring the pressure of gas in voids located so that no unfed shrinkage would exist, as for example in a blind runner.

## EXPERIMENTAL PROCEDURE

The casting to be examined was first cut into convenient-sized sections for density and gas determinations. These samples generally weighed between 50 and 200 grams. Certain sections were taken for metallographic examination.

### *Density Measurement*

The density of the casting sections was determined by the buoyancy method in which the loss in weight of the sample when suspended in distilled water was used to calculate its volume. Corrections were made to reduce weighings to vacuo, to allow for the volume of water displaced by the suspension wire and to adjust for the absolute density of water at the temperature of measurement.

### *Gas Content and Composition Determination*

The gas content was determined by a vacuum fusion technique. The volume of gas released was measured by observing the pressure of this gas in a system of known volume at a measured temperature.

The gas composition was determined by mass spectrometric analysis of small samples sent to the National Research Council of Canada. These samples, ranging in volume from 0.05 to 2 cc (STP), were contained in 50 cc sealed flasks equipped with a break-seal.

### *Calculations*

The density at 20 C (68 F) of die casting alloy containing 4.0 per cent aluminum has been reported as 6.70[5] and 6.6[6] g/cc. Cast samples which were essentially void-free, and whose compositions were determined, were prepared in the laboratory using carefully controlled freezing conditions. The highest density obtained for the 4.0 per cent alloy approached 6.70 g/cc, and this figure was taken as the true density.

The coefficient of linear expansion for the alloy has been reported as $37.1 \times 10^{-6}$ per degree centrigade for the range 20 C (68 F) to the melting point. If the measured density of the sample at 20 C (68 F) is *d*, then the volume of voids, $V_v$, in cc/100 g at the solidus temperature, 380 C (716 F), is

$$100\left(\frac{1}{d}-\frac{1}{6.70}\right)[1+(3\times 37.1\times 10^{-6}\times 360)]$$

or

$$V_v=\frac{104.01}{d}-15.52$$

In calculating the volume of gas present at the solidification temperature $V_g$, it is assumed that the oxygen content of the trapped air was present at the time of solidification. However, since no molecular oxygen was found in the gas recovered by vacuum fusion, the appropriate correction to the measured gas has to be made.

If the measured gas content is called S cc/100 g at 0 C and 760 mm pressure (STP), and the measured or estimated nitrogen content is C volume per cent, then the volume of trapped gas at STP is

$$S+\left(\frac{21}{78}\right)\left(\frac{C}{100}\right)S=S(1+0.00269\,C)$$

The analyses of a number of samples of gas collected from castings showed that the percentage nitrogen varied from 73.5 to 88.8 and averaged about 85. A value for C of 85 was used in the calculations, and it will be found that variations from this value within the range quoted introduce only a small error (< 3 per cent) in the calculated gas volume.

It is assumed that, at the time of solidification, the temperature of the trapped gas in the cavity will approach that of the alloy.[5] Then, converting the volume of gas to conditions just prior to solidification, 380 C (716 F), and pressure P (psi) gives

$$V_g=S[1+(0.00269)(85)]\left(\frac{14.7}{P}\right)\left(\frac{653}{273}\right)$$

$$V_g=43.2\,\frac{S}{P}\,\text{cc}/100\,\text{g}$$

There is a possibility that the oxygen in the trapped air would be completely reacted at the time of solidification, in view of the turbulence and intimate contact between the gas and metal. If this is so, then no correction for oxygen should be made. This would not affect the values of $V_g$ calculated from pressures obtained from the blind runner measurements described later. However, the values of $V_g$ calculated on the basis of pressure measured at the plunger would be 19 per cent lower if the correction for oxygen is not employed.

The difference between the volume of gas just prior to solidification $V_g$, and the volume of voids after solidification $V_v$, is the unfed solidification shrinkage, USS.

$$\text{i.e. } V_v-V_g=\text{USS}$$

The density of liquid alloy at 380 C (816 F) is reported[5] as 6.25 g/cc. The density of solid alloy at the solidus temperature calculated from a density of 20 C (68 F) of 6.70 g/cc and a linear coefficient of expansion of $37.1 \times 10^{-6}$ is 6.44 g/cc.

The theoretical solidification shrinkage for the alloy is therefore 0.477 cc/100 g or 2.98 volume per cent.

## POROSITY NATURE AND DISTRIBUTION IN AN AUTOMOTIVE WINDOW FRAME CASTING

Detailed investigations were carried out on the automotive window frame casting, shown in Fig. 1. The casting consists of a C-shaped frame with a relatively thick rectangular-shaped base. It is fed through two gates—a long thin gate along the side of the frame

about 8½-in. in length by 24-43 mils in thickness and a short thicker gate at the base measuring about 2 in. in length by 50-70 mils in thickness. The weight of the trimmed casting is about 1600 g (3.5 lb).

A total of 19 castings produced in three different dies were examined. All three dies were of essentially the same design, though there were some differences in gate thicknesses. The castings were cut into sections, as shown in Fig. 2, and density and gas determinations made. Section 8M was used for metallographic examinations. Complete results for one casting from each series of samples examined are given in Tables 1 and 2. The gas content of the various sections is plotted in Fig. 3.

Analysis of the castings revealed that the aluminum content was in every case sufficiently close to 4.0 per cent that the effect of variations in the theoretical density of the alloys could be neglected.

To use the density and gas figures to describe conditions in the die at the time of freezing it is necessary to determine the pressure which existed in the cavity after filling but before freezing. This pressure dictates the total volume of the gas bubbles before general freezing commences, and would correspond to the steady pressure reached after the final pressure peak in the experiments reported by Barton and Sakui.[4] If the hydraulic pressure after filling is known then from the ratio of the area of the shot cylinder to that of the metal cylinder, the magnitude of the applied pressure at this point should be calculable.

Failing this information, the pressure could be determined from a section of the casting which was molten after the completion of die filling, and which froze directionally in such a way that virtually no unfed shrinkage resulted. Under these conditions, the volume of the voids after freezing would be equal to the volume of gas before freezing (i.e., USS = 0, therefore: $V_g = V_v$), and the pressure can be calculated using the relationship between $V_g$ and P developed above. In the casting under consideration, it was assumed that the blind runner, marked in Fig. 2, fulfilled these requirements.

A further requirement met by the blind runner was the presence of an appreciable amount of gas so that the effects of any errors in measurement are minor. Densities and gas contents for the blind runners were determined and the pressures calculated. The results are shown in Table 3.

**TABLE 1 — DENSITY OF DIE CASTING SECTIONS**

| Die | A | A | O | B |
|---|---|---|---|---|
| Casting No. | 300b | 305c | 300f | 326a |
| Weighted Mean | 6.45 | 6.42 | 6.50 | 6.54 |
| Position | Density, g/cc | | | |
| 1 | 6.65 | 6.61 | 6.65 | 6.63 |
| 2 | 6.63 | 6.59 | 6.63 | 6.61 |
| 3 | 6.62 | 6.55 | 6.61 | 6.61 |
| 4 | 6.45 | 6.42 | 6.47 | 6.49 |
| 5 | 6.32 | 6.24 | 6.38 | 6.41 |
| 6 | 6.42 | 6.32 | 6.42 | 6.44 |
| 7 | 6.40 | 6.33 | 6.51 | 6.53 |
| 8 | 6.24 | 6.21 | 6.43 | 6.47 |
| 9 | 6.29 | 6.37 | 6.35 | 6.54 |
| 10 | 6.44 | 6.50 | 6.51 | 6.64 |
| 11 | 6.53 | 6.54 | 6.58 | 6.66 |
| 12 | 6.53 | 6.53 | 6.54 | 6.61 |

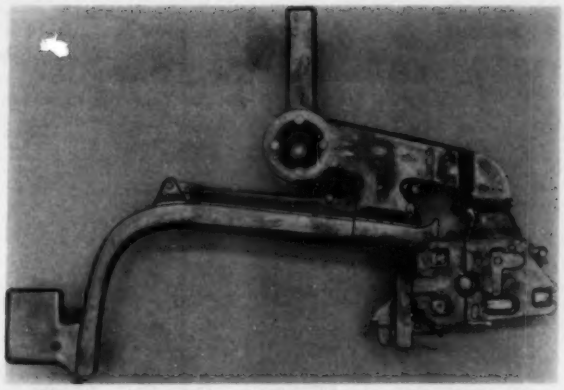


**Fig. 1 — Automotive window frame casting used for porosity investigations.**

**TABLE 2 — GAS CONTENT OF DIE CASTING SECTIONS**

| Die | A | A | O | B |
|---|---|---|---|---|
| Casting No. | 300b | 305c | 300f | 326a |
| Weighted Mean | 7.51 | 7.78 | 6.05 | 8.49 |
| Position | Gas Content, S, cc/100g (STP) | | | |
| 1 | 2.68 | 3.31 | 2.56 | 6.41 |
| 2 | 4.25 | 4.97 | 4.43 | 7.47 |
| 3 | 3.97 | 4.79 | 3.85 | 6.45 |
| 4 | 6.00 | 6.24 | 5.48 | 8.85 |
| 5 | 8.04 | 9.94 | 6.94 | 12.86 |
| 6 | 8.98 | 10.49 | 7.77 | 11.89 |
| 7 | 10.21 | 10.37 | 7.25 | 8.93 |
| 8 | 13.45 | 12.65 | 8.97 | 10.15 |
| 9 | 11.45 | 8.63 | 8.73 | 7.92 |
| 10 | 7.84 | n/a* | 4.98 | 3.56 |
| 11 | 7.44 | 5.75 | 5.44 | 4.81 |
| 12 | 5.84 | 7.17 | 6.16 | 8.51 |

*n/a indicates not available.

**TABLE 3 — MEASUREMENTS OF DENSITY AND GAS CONTENTS OF BLIND RUNNERS AND CALCULATED METAL PRESSURE**

| Die | Casting No. | d, g/cc | S, cc/100g (STP) | $V_v$ ($V_g$), cc/100g | Pressure, psi |
|---|---|---|---|---|---|
| A | 300b | 6.57 | 12.30 | 0.31 | 1714 |
| A | 305c | 6.52 | 11.82 | 0.43 | 1187 |
| O | 300f | 6.53 | 15.60 | 0.41 | 1644 |
| B | 326a | 6.51 | 25.05 | 0.46 | 2353 |

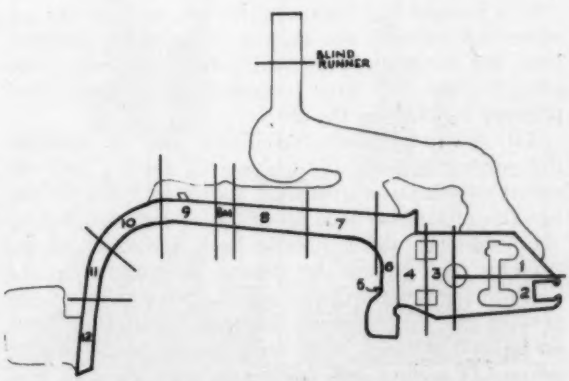



**Fig. 2 — Location of sections used for density and gas determinations.**

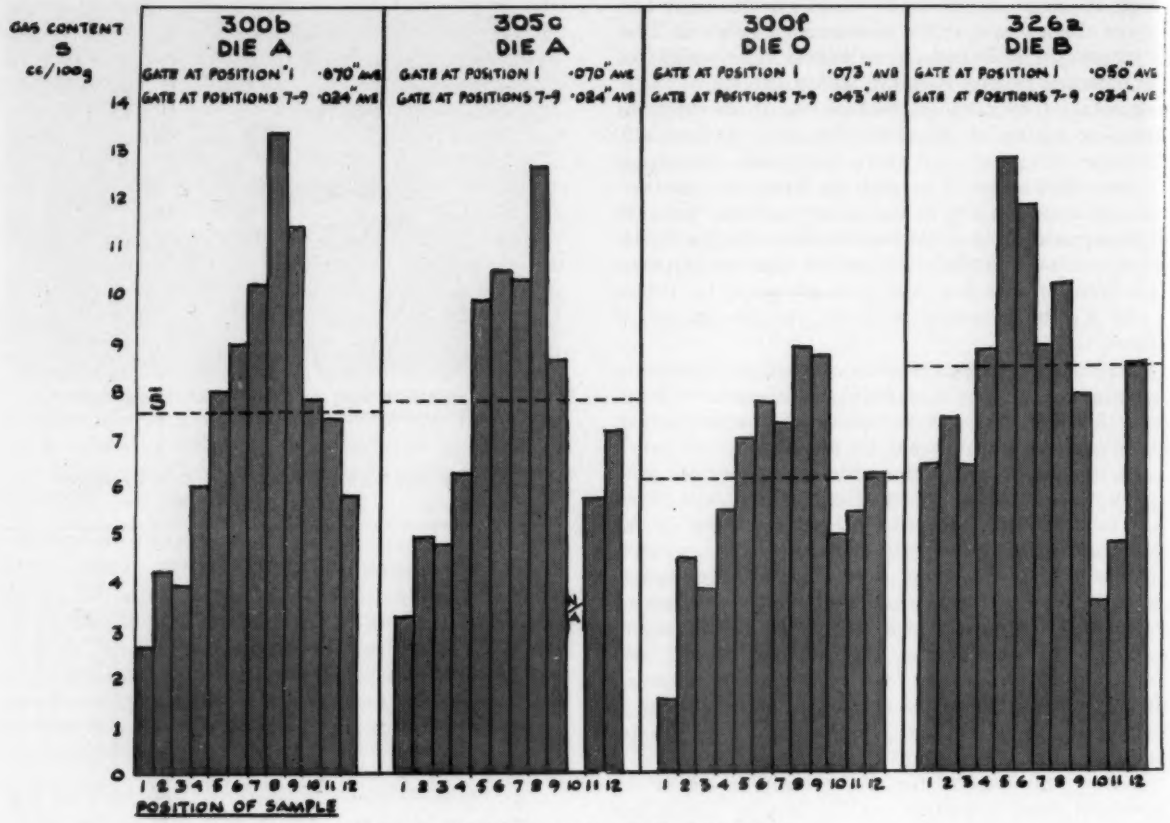



Fig. 3 — Variation of gas content with position.

### *Metal Pressure Variations*

Wide variations in metal pressure for the different castings are indicated by these calculations. It should be pointed out that the castings were produced on different machines except for the two series from die *A* and in this case there was a lapse of several months between the series. In a given series produced over a short period, the indicated blind runner pressures were quite consistent, varying less than 10 per cent from the average. Consideration of the results of measurements made on the castings themselves reveals that pressure variations of the order indicated in Table 3 did exist and confirms, at least partially, the validity of the assumptions.

It is known that operating factors, such as the accumulator pressure, the amount of gas in the accumulator, the condition of the hydraulic system and of the plunger rings, can have a marked effect on the final pressure applied to the metal.

The above pressures have been used to calculate the volume of voids $V_v$, volume of gas $V_g$, and the unfed solidification shrinkage, shown in Table 4. The functions have also been plotted as a histogram (Fig. 4).

Considerable differences in both densities and gas contents exist among the castings produced from the different dies. It will be noted, however, that the general pattern of density and gas variations is similar for all castings, with the sections in the middle portion (Positions 4-9) being generally lower in density and higher in gas content than the remainder of the casting (Positions 1-3 and 10-12).

The gas content of the various sections of the castings provides an indication of the flow pattern of the metal in filling the cavity, and of the manner in which the air present in the cavity is moved by the incoming metal. Because of the location of the two gates, it would be expected that the gas content would be high at a point between them where the two inflowing streams of metal meet. This has in fact occurred (Fig. 3). It is interesting to note further that in the case of the castings produced in dies *A* and *O* the highest gas content occurs at position 8, while in the casting produced in die *B* it occurs at position 5.

As noted in Fig. 3, it was found that the average thickness of the gate at position one in die *B* was 50 mils, whereas in dies *A* and *O* the average thickness of this gate was about 70 mils. The gate at positions 7-9 on die *B* was intermediate in thickness between that on dies *A* and *O*. It would appear that, within limits, the size of the gate at position one would have the greater effect in determining the meeting point of the streams from the two gates, and that the variation noted is the reason for the difference in gas distribution.

### *Shrinkage vs. Porosity*

An outstanding point revealed by the calculated values for $V_v$, $V_g$ and USS (Table 4 and Fig. 4) is the large contribution of shrinkage to the porosity level. For every casting the unfed shrinkage makes up over 50 per cent of the volume of voids when averaged

TABLE 4 — CALCULATED VALUES FOR $V_v$, $V_g$, AND USS

| Die | Casting No. | Position | d, g/cc | S, cc/100g (STP) | $V_v$, cc/100g | $V_g$, cc/100g | $\frac{V_g}{V_v}$, % | USS, cc/100g | USS, % of Theoretical |
|---|---|---|---|---|---|---|---|---|---|
| A | 300b | 1 | 6.65 | 2.68 | 0.12 | 0.07 | 58.3 | 0.05 | 10.5 |
| | | 2 | 6.63 | 4.25 | 0.17 | 0.11 | 64.7 | 0.06 | 12.6 |
| | | 3 | 6.62 | 3.97 | 0.19 | 0.10 | 52.6 | 0.09 | 18.7 |
| | | 4 | 6.45 | 6.00 | 0.60 | 0.15 | 25.0 | 0.45 | 94.3 |
| | | 5 | 6.32 | 8.04 | 0.94 | 0.20 | 21.3 | 0.74 | 155.1 |
| | | 6 | 6.42 | 8.98 | 0.68 | 0.23 | 33.8 | 0.45 | 94.3 |
| | | 7 | 6.40 | 10.21 | 0.73 | 0.26 | 35.6 | 0.47 | 98.5 |
| | | 8 | 6.24 | 13.45 | 1.15 | 0.34 | 29.6 | 0.81 | 169.8 |
| | | 9 | 6.29 | 11.45 | 1.02 | 0.29 | 28.4 | 0.73 | 153.0 |
| | | 10 | 6.44 | 7.84 | 0.63 | 0.20 | 31.7 | 0.43 | 90.1 |
| | | 11 | 6.53 | 7.44 | 0.41 | 0.19 | 46.3 | 0.22 | 46.1 |
| | | 12 | 6.53 | 5.84 | 0.41 | 0.15 | 36.6 | 0.26 | 54.5 |
| Weighted Mean | | | 6.47 | 7.52 | 0.56 | 0.19 | 33.9 | 0.37 | 77.6 |
| A | 305c | 1 | 6.61 | 3.31 | 0.22 | 0.12 | 54.5 | 0.10 | 21.0 |
| | | 2 | 6.59 | 4.97 | 0.26 | 0.18 | 69.2 | 0.08 | 16.8 |
| | | 3 | 6.55 | 4.79 | 0.36 | 0.17 | 47.2 | 0.19 | 39.8 |
| | | 4 | 6.42 | 6.24 | 0.68 | 0.23 | 33.8 | 0.45 | 94.3 |
| | | 5 | 6.24 | 9.94 | 1.15 | 0.36 | 31.3 | 0.79 | 165.6 |
| | | 6 | 6.32 | 10.49 | 0.94 | 0.38 | 40.4 | 0.56 | 117.4 |
| | | 7 | 6.33 | 10.37 | 0.91 | 0.38 | 41.8 | 0.53 | 111.1 |
| | | 8 | 6.21 | 12.65 | 1.23 | 0.46 | 37.4 | 0.77 | 161.4 |
| | | 9 | 6.37 | 8.63 | 0.81 | 0.31 | 38.3 | 0.50 | 104.8 |
| | | 10 | 6.50 | n/a* | 0.48 | — | — | — | — |
| | | 11 | 6.54 | 5.75 | 0.38 | 0.21 | 55.3 | 0.17 | 35.6 |
| | | 12 | 6.53 | 7.17 | 0.41 | 0.26 | 63.4 | 0.15 | 31.4 |
| Weighted Mean | | | 6.42 | 7.78 | 0.68 | 0.28 | 41.2 | 0.40 | 83.9 |

*(continued in next column)*

TABLE 4 — CALCULATED VALUES FOR $V_v$, $V_g$, AND USS *(continued)*

| Die | Casting No. | Position | d, g/cc | S, cc/100g (STP) | $V_v$, cc/100g | $V_g$, cc/100g | $\frac{V_g}{V_v}$, % | USS, cc/100g | USS, % of Theoretical |
|---|---|---|---|---|---|---|---|---|---|
| O | 300f | 1 | 6.65 | 2.56 | 0.12 | 0.07 | 58.3 | 0.05 | 10.5 |
| | | 2 | 6.63 | 4.43 | 0.17 | 0.12 | 70.6 | 0.05 | 10.5 |
| | | 3 | 6.61 | 3.85 | 0.22 | 0.10 | 45.5 | 0.12 | 25.2 |
| | | 4 | 6.47 | 5.48 | 0.56 | 0.14 | 25.0 | 0.42 | 88.0 |
| | | 5 | 6.38 | 6.94 | 0.78 | 0.18 | 23.1 | 0.60 | 125.8 |
| | | 6 | 6.42 | 7.77 | 0.68 | 0.20 | 29.4 | 0.48 | 100.6 |
| | | 7 | 6.51 | 7.25 | 0.46 | 0.19 | 41.3 | 0.27 | 56.6 |
| | | 8 | 6.43 | 8.97 | 0.66 | 0.24 | 36.4 | 0.42 | 88.0 |
| | | 9 | 6.35 | 8.73 | 0.86 | 0.23 | 26.7 | 0.63 | 132.1 |
| | | 10 | 6.51 | 4.98 | 0.46 | 0.13 | 28.3 | 0.33 | 69.2 |
| | | 11 | 6.58 | 5.44 | 0.29 | 0.14 | 48.3 | 0.15 | 31.4 |
| | | 12 | 6.54 | 6.16 | 0.38 | 0.16 | 42.1 | 0.22 | 46.1 |
| Weighted Mean | | | 6.50 | 6.09 | 0.48 | 0.16 | 33.3 | 0.32 | 67.1 |
| B | 326a | 1 | 6.63 | 6.41 | 0.17 | 0.12 | 70.6 | 0.05 | 10.5 |
| | | 2 | 6.61 | 7.47 | 0.22 | 0.14 | 63.6 | 0.08 | 16.8 |
| | | 3 | 6.61 | 6.45 | 0.22 | 0.12 | 54.5 | 0.10 | 21.0 |
| | | 4 | 6.49 | 8.85 | 0.51 | 0.16 | 31.4 | 0.35 | 73.4 |
| | | 5 | 6.41 | 12.86 | 0.71 | 0.24 | 33.8 | 0.47 | 98.5 |
| | | 6 | 6.44 | 11.89 | 0.63 | 0.22 | 34.9 | 0.41 | 86.0 |
| | | 7 | 6.53 | 8.93 | 0.41 | 0.16 | 39.0 | 0.25 | 52.4 |
| | | 8 | 6.47 | 10.15 | 0.56 | 0.19 | 33.9 | 0.37 | 77.6 |
| | | 9 | 6.54 | 7.92 | 0.38 | 0.15 | 39.5 | 0.23 | 48.2 |
| | | 10 | 6.64 | 3.56 | 0.14 | 0.07 | 50.0 | 0.07 | 14.7 |
| | | 11 | 6.66 | 4.81 | 0.10 | 0.09 | 90.0 | 0.01 | 2.1 |
| | | 12 | 6.61 | 8.51 | 0.22 | 0.16 | 72.7 | 0.06 | 12.6 |
| Weighted Mean | | | 6.54 | 8.50 | 0.38 | 0.16 | 42.1 | 0.22 | 46.1 |

*n/a indicates not available

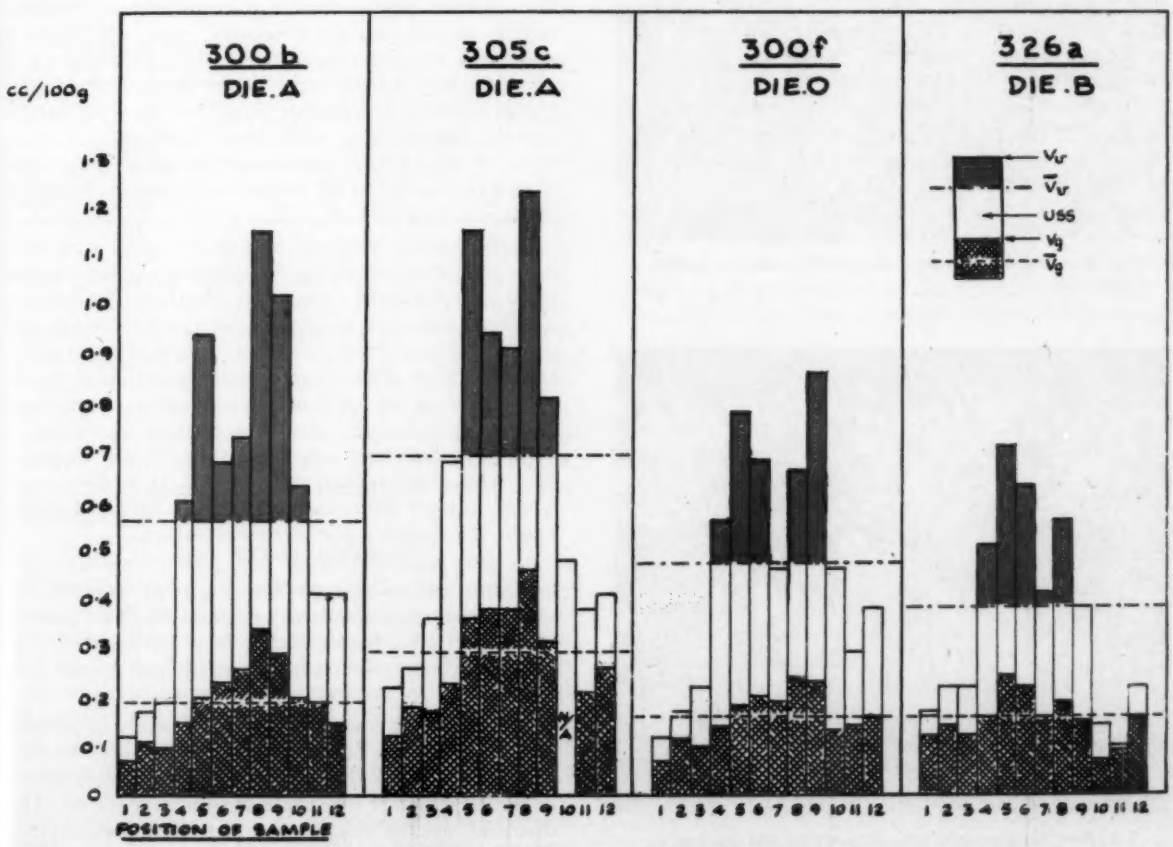



Fig. 4 — Variation of functions $V_v$, $V_g$ and USS with position.

over all sections. In sections of high porosity (density between 6.2 and 6.3) shrinkage contributes between 60 and 70 per cent of the total volume of the voids. In these cases the calculated shrinkage is considerably greater than the theoretical solidification shrinkage for the alloy.

This is interpreted as indicating that these regions froze late in the solidification cycle, and that molten metal moved from them to adjacent regions to compensate for shrinkage. The generally high level of shrinkage found in castings 300b, 305c and 300f indicates that the gates froze quite early in the cycle preventing access of molten metal under pressure during most of the freezing period.

The two castings from die *A* show similar gas contents, indicating similar flow patterns in the die as might be expected. There was, however, an appreciable difference in applied pressure, as shown in Table 3, and this would result in a greater volume of gas ($V_g$) in 305c at the time of freezing. The extent of unfed shrinkage was apparently the same in the two castings; thus the difference in gas volume has given rise to a slightly lower density in casting 305c.

The casting from die *O* (300f) shows an improvement from the point of view of overall porosity on these from die *A*. The indicated pressure on the melt after filling was similar to that of casting 300b; however, the gas content was lower, and therefore the gas volume was less. The amount of unfed shrinkage was also lower. It will be noted that the greatest difference in gas content occurred in sections 7-12 of the castings, particularly near the gate. As noted in Fig. 3, the gate thickness was appreciably greater in the case of die *O*, and this may well have led to reduced turbulence and less entrapment of gas in this region.

Casting 326a, which was produced in die *B*, showed a still lower porosity level than the other castings (Fig. 4). While the overall gas content was a little higher than that of the other castings (Fig. 3), its calculated volume was somewhat lower because of the higher indicated pressure. The unfed solidification shrinkage was appreciably lower, and these two factors together accounted for the decreased porosity level in the solidified casting.

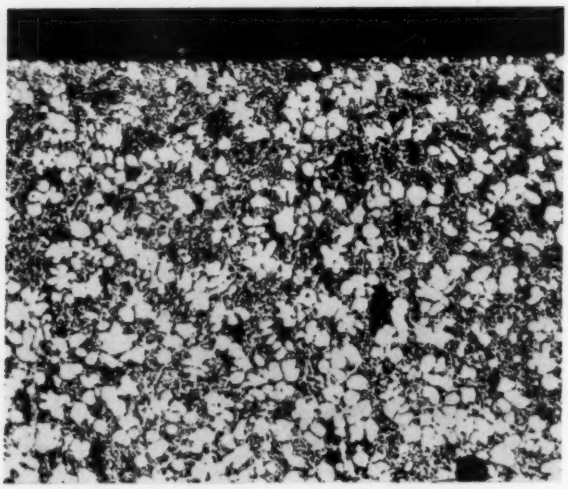


**Fig. 5 — Surface structure at position 8M on casting 300b. 250 ×.**



**Fig. 6 — Surface structure at position 8M on casting 305c. 250 ×.**

### *Die Thermal Conditions*

The porosity and other characteristics of the castings are also affected by the thermal conditions existing in the die during filling and solidification. Some information of a qualitative nature on these conditions can be obtained from a metallographic examination of the casting structures. Since the surface structures vary in nature from place to place on a given casting, it is necessary when comparing a number of castings to examine structures at identical locations. Figures 5 to 8 illustrate comparative structures on the castings under consideration. The fields shown are located at the ejector die surface at identical points on section 8 M.

Castings 300b, 305c and 300f show similar structures indicating that the thermal conditions, at least at the location considered, were also similar. The surface structure observed in casting 326a was however distinctly different from that found in the other castings. Over much of the surface, which was adjacent to the ejector die at the position examined, a pronounced growth of columnar dendrites existed, as shown in Fig. 8. It has been observed that, in die castings, this type of structure often occurs at surface areas where a high heat extraction rate during freezing might be expected, as at a convex surface.

In the case of casting 326a then, this structure may represent more rapid freezing rates, possibly due to lower die temperatures or more effective cooling, as compared to those existing in the other dies. The difference in thermal conditions indicated may have contributed to the lower porosity levels.

An important consideration in the study of porosity is its distribution between the surface layers and the interior of the casting, since this may determine the extent to which it will cause detrimental effects. This distribution was readily obtained from examination of the metallographic sections. In castings 300b, 305c

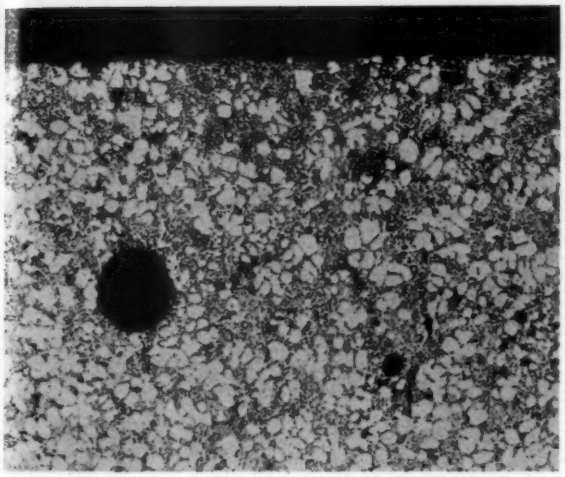


Fig. 8 — Surface structure at position 8M on casting 326a. 250 ×.

Fig. 7 — Surface structure at position 8M on casting 300d. 250 ×.

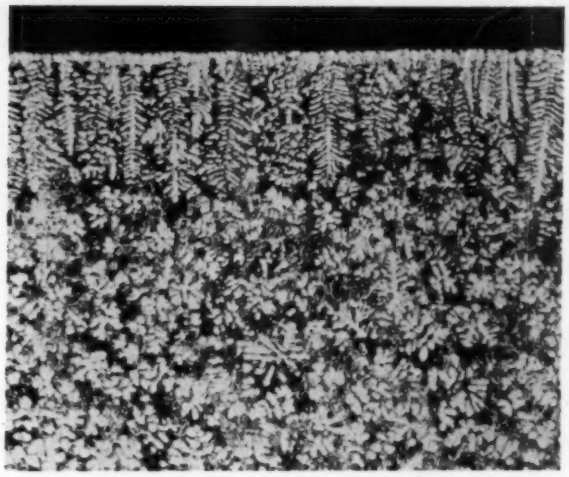


TABLE 5 — GAS ANALYSIS

| | | | | |
|---|---|---|---|---|
| Die | A | A | O | B |
| Position | 9 | 8 | 8 | 8 |
| Sample | 300a | 305c | 300d | 326a |
| Gas Content for sample, cc/100g (STP) | 10.41 | 12.65 | 9.13 | 10.15 |
| Gas | Analysis, in volume % | | | |
| $H_2$ | 13.8 | 12.3 | 12.9 | 25.0 |
| $N_2$ | 84.6 | 86.0 | 86.0 | 73.5 |
| Ar | 0.96 | 1.03 | 1.00 | 0.86 |
| $CO_2$ | 0.18 | 0.40 | — | 0.15 |
| $CH_4$ | 0.37 | 0.25 | 0.26 | 0.40 |

and 300f pores extended close to the surface layers (within 5 mils of the surface), while in casting 326a the surface layers were relatively free from porosity. This suggested that surface porosity was not directly related to the general porosity level but rather to thermal conditions existing in the cavity. This is supported by the observation that low densities result mainly from high levels of unfed shrinkage, which would be expected to affect mainly the center of the casting.

## GAS COMPOSITION

To investigate the source of the gas present in commercial die castings, detailed analyses of gas samples taken from a number of die castings and samples of die casting alloy were carried out. The analysis was done on a mass spectrometer by the National Research Council of Canada. Typical results from the die casting samples are given in Table 5. The gas samples were taken from sections of the castings, and the gas content of the section used in each case is tabulated. The analyses of gas samples from castings 300b and 300f were not available, however, castings 300a and 300d quoted in the table were produced at the same time as 300b and 300f, respectively.

The gas sample obtained by vacuum fusion will include, along with the gas released from the pores, gas which was in solution in the alloy. To check the extent of the contribution of dissolved gas, several samples of pore-free cast alloy were melted under vacuum and the amount of gas evolved was measured and analysed. Of the order of 0.4 cc of gas/100 g of alloy (STP) was obtained from these samples, and it analysed approximately 40 per cent hydrogen and 25 per cent nitrogen by volume. The other major constituent of the gas was carbon dioxide, amounting to 30 per cent. Thus, the amount of gas which could originate from solution in the alloy is quite small in comparison with the total gas recovered from the die casting samples.

It will be seen from Table 5 that the principal constituents in the gas taken from the castings are nitrogen and hydrogen. It should be noted however that since carbon monoxide and nitrogen both have a mass of 28, they are not readily distinguishable on the mass spectrometer and are reported together as nitrogen. Since the solubility of nitrogen in the alloy is low, it can be assumed that practically all of the nitrogen present originated from trapped air in the casting. The nitrogen-argon ratio is somewhat greater than the normal 83 : 1 for air and probably the excess nitrogen reported (up to 0.5 cc/100 g) is actually carbon monoxide, possibly originating from lubricant.

The results show a hydrogen content of from 1.2 to 2.5 cc/100 g. According to the findings quoted above, about 0.2 cc/100 g could come from the alloy. It would appear that the remainder must have come from the reduction of water vapor in the entrapped air, or from the breakdown of lubricant in the

TABLE 6 — LUBRICATION EFFECT ON DENSITY AND GAS CONTENT

| Die | Casting No. | Lubrication Conditions | Mean Density, Pos. 7-12, g/cc | Density, Pos. 8 g/cc | Gas Content (S), Pos. 8, cc/100g (STP) | Metal Pressure (As indicated by blind runner, psi | $V_g$, cc/100g |
|---|---|---|---|---|---|---|---|
| A | 305a | No lubricant | 6.40 | 6.26 | 11.45 | 1281 | 0.39 |
| A | 305b | No lubricant | 6.39 | 6.28 | 11.53 | 1228 | 0.41 |
| A | 305c | Normal spray Water emulsion lubricant (X) | 6.39 | 6.21 | 12.65 | 1187 | 0.46 |
| A | 305d | Excessive spray Lubricant (X) | 6.38 | 6.14 | 15.45 | 1221 | 0.55 |
| A | 305e | Excessive spray Lubricant (X) | 6.40 | 6.25 | 13.77 | 1277 | 0.47 |
| A | 305f | Normal spray Organic solvent lubricant (Y) | 6.38 | 6.21 | 13.05 | 1152 | 0.49 |
| A | 305g | Excessive spray Lubricant (Y) | 6.36 | 6.22 | 13.53 | 1230 | 0.48 |
| A | 305h | Normal spray Organic solvent lubricant (Z) | 6.41 | 6.30 | 10.73 | 1295 | 0.36 |
| A | 305i | Excessive spray Lubricant (Z) | 6.39 | 6.22 | 12.81 | 1287 | 0.43 |
| A | 305j | Excessive spray Lubricant (Z) | 6.38 | 6.22 | 14.57 | 1235 | 0.51 |

die. Calculations show that if the entrapped air were warm and of high humidity, the hydrogen which would be present as water vapor could account for practically all of the rest of the hydrogen found. On the other hand, if the lubricant used were a water-base type this could well be another source of water vapor.

In the event that vaporization was not complete when the die closed it could also lead to increased gas content in the casting. Similarly, it might be expected that organic-based lubricants could contribute hydrogen and lead to increased gas content as well, if their end points were too high.[7] To investigate further the effects of lubrication on porosity and gas content, a separate series of tests was carried out.

### *Lubricant Tests*

The lubricant tests were done on the window frame casting previously described. Differing types and amounts of lubricant were applied to the die, as outlined in Table 6. This table also shows the mean density of the six positions on the frame portion of the casting (position 7-12) and the gas content of position 8 on the castings. The castings were all produced on the same die over a short period of time with other casting conditions kept as constant as possible.

The composition of the gas released from position 8 in each case is shown in Table 7. Taking as a base level the hydrogen content for the nonlubricated castings (1.2 cc/100 g), the increase in hydrogen content for each of the lubricated castings is also shown.

TABLE 7 — LUBRICANT EFFECT ON GAS COMPOSITION

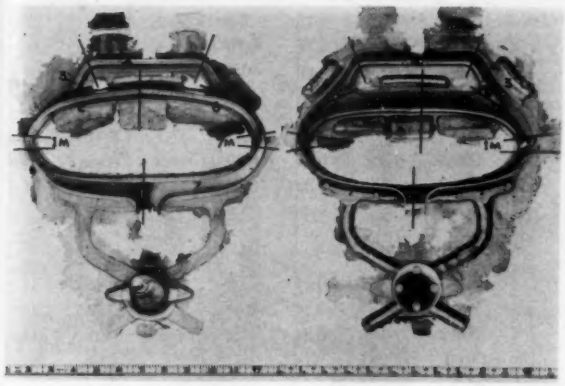
| Casting No., Pos. 8 | Lubricant | Gas Composition, Volume % $H_2$ | $N_2$ | Ar | $CO_2$ | $CH_4$ | Hydrogen Content, cc/100g (STP) | Increase in Hydrogen Content, cc/100g (STP) |
|---|---|---|---|---|---|---|---|---|
| 305a | None | 10.0 | 88.8 | 1.05 | — | 0.15 | 1.15 | −0.05 |
| 305b | None | 10.8 | 87.9 | 1.03 | 0.10 | 0.14 | 1.25 | +0.05 |
| 305c | Type X | 12.3 | 86.0 | 1.03 | 0.40 | 0.25 | 1.56 | 0.36 |
| 305d | Excess Type X | 15.2 | 83.6 | 0.94 | — | 0.30 | 2.35 | 1.15 |
| 305e | Excess Type X | 19.1 | 79.5 | 0.90 | — | 0.52 | 2.63 | 1.43 |
| 305f | Type Y | 12.7 | 86.0 | 1.02 | 0.1 | 0.18 | 1.66 | 0.46 |
| 305g | Excess Type Y | 19.9 | 78.7 | 0.84 | 0.1 | 0.47 | 2.69 | 1.49 |
| 305h | Type Z | 15.4 | 83.3 | 1.08 | — | 0.22 | 1.65 | 0.45 |
| 305i | Excess Type Z | 15.2 | 83.6 | 0.90 | — | 0.26 | 1.95 | 0.75 |
| 305j | Excess Type Z | 21.8 | 76.8 | 0.84 | — | 0.56 | 3.18 | 1.98 |

The results of this test show, first of all, that the use of lubricants even in excessive amounts had little or no effect on the mean density of the casting. The gas content of position 8 did however increase with the use of normal amounts of lubricant, and an even greater increase resulted from the use of excessive amounts. Gas contents for the other sections of the castings are not available, but it would appear that while an increase was noted in position 8 the change over the whole casting was not sufficient to effect the mean density appreciably. The density of position 8 varied widely but seemed to bear little relationship to the lubricant conditions. This was apparently because of variations in the distribution of unfed shrinkage in the different castings.

Considering the gas composition, a definite increase in the hydrogen content (and also in the methane content) resulted from the use of lubricants. In the case of position 8 this was sufficient to account for an increase in the hydrogen content of up to 1.5 to 2 cc/100 g in instances of excessive amounts of lubricant spray. With normal amounts of spray increases in hydrogen content of the order of 0.4 cc/100 g were found.

Summarizing the findings on gas porosity, it has been found that the gas present in commercial die castings consists principally of nitrogen and hydrogen with minor amounts of argon, carbon dioxide, methane and carbon monoxide also present. The nitrogen has originated from trapped air in the castings; the oxygen associated with the nitrogen has been lost by reaction either during casting or during the melting

Fig. 9 — Automotive ornament casting used for investigation of effects of variations in injection pressure on porosity.



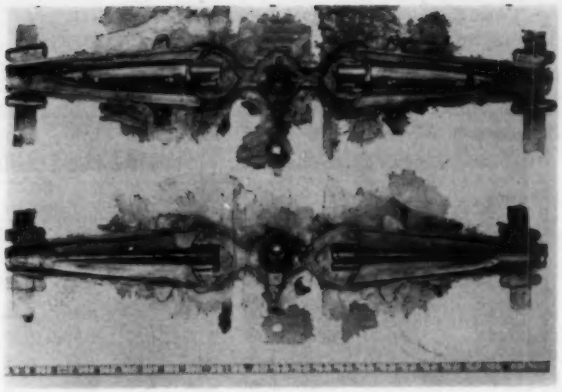


Fig. 10 — Automotive taillight casting used for investigation of effects of variations of injection pressure on porosity. Locations of sections used for density and gas determinations are marked.

procedure for gas recovery. The hydrogen content can be readily accounted for as originating from a combination of sources—dissolved gas in the alloy, water vapor in the trapped air and decomposition of lubricants. The use of lubricant even in excessive quantities had little or no effect on the overall density of the casting, though it did increase the gas content appreciably at some locations.

## INJECTION PRESSURE EFFECT ON POROSITY IN AN AUTOMOTIVE TAILLIGHT CASTING

Two further series of tests were carried out to study the effect of injection pressure on the porosity and gas content of die castings. The tests were performed on the automotive ornament casting, shown in Fig. 9, and on the automotive taillight, shown in Fig. 10. Since the results obtained for the two castings were generally similar, only the findings for the taillight casting will be described.

The taillight casting was fed by two gates each of which was 4 in. in length and averaged about 35 mils in thickness. The weight of the trimmed casting was about 500 g (1.1 lb).

The test castings were produced in a single run using seven different pressures. The hydraulic pressure was varied in 100 psi steps from 1200 to 600 psi, corresponding to metal pressures of 2140 to 1070 psi. Six castings were produced at each pressure. After every third shot the die was lubricated. As far as possible, all conditions were kept constant during the test; at the lowest pressure (1070 psi) however, some difficulty was experienced in locking the machine and the casting rate was necessarily cut from three to two/min. Die temperatures were 221-238 C (430-460 F) for the ejection die and 238-249 C (460-480 F) for the cover die. Metal temperature was 418-421 C (785-790 F).

Two castings from each pressure, one before and one after lubrication, were chosen for testing. The castings were cut into nine sections, as shown in Fig. 10. Density determinations were carried out on all sections. However, since the casting was symmetrical in shape, and since density determinations indicated that the two sides were similar, gas determinations were limited to sections 4 to 7 inclusive. These gas determinations were performed on castings from three pressures—2140, 1600 and 1070 psi. For the remainder of the pressure levels, determinations were made on section 7 only.

### *Density and Gas Determinations*

All of the density and gas determinations are shown in Tables 8 and 9 and Fig. 11. These figures have been used to calculate the functions $V_v$, $V_g$ and USS, which are shown in Table 10 and are also plotted on the histogram (Fig. 12). For these calculations the metal pressure figure determined from the hydraulic pressure measured during casting has been used for the pressure on the liquid alloy in the die cavity before freezing.

While it is recognized that this calculated pressure may be subject to some error due to operational factors, the fact that all of the castings were produced on the same machine over a short period of time should make the results comparative. It should be pointed out also that, as mentioned previously, if the oxygen in the entrapped gas is all reacted prior to the complete filling of the cavity, the calculated values of $V_g$ should be about 19 per cent lower than shown.

In general, the castings were found to be of high density and low gas content compared to other commercial castings which have been examined. The amount of unfed shrinkage was also quite low, amounting to less than 40 per cent of the theoretical solidification shrinkage for the alloy.

TABLE 8 — INJECTION PRESSURE EFFECT ON DENSITY OF TAILLIGHT MOUNT CASTINGS

| Position | Density, g/cc | | | | | | |
|---|---|---|---|---|---|---|---|
| **Nonlubricated Castings** | | | | | | | |
| Casting No. | 1200-3 | 1100-5 | 1000-5 | 900-5 | 800-5 | 700-5 | 600-2 |
| Pressure, psi | 2140 | 1960 | 1780 | 1600 | 1420 | 1250 | 1070 |
| Weighted Mean | 6.59 | 6.59 | 6.58 | 6.56 | 6.57 | 6.56 | 6.56 |
| 1 | 6.61 | 6.61 | 6.59 | 6.58 | 6.60 | 6.57 | 6.58 |
| 1M | 6.48 | 6.47 | 6.45 | 6.44 | 6.44 | 6.41 | 6.43 |
| 2 | 6.60 | 6.59 | 6.58 | 6.57 | 6.58 | 6.56 | 6.57 |
| 3 | 6.61 | 6.60 | 6.59 | 6.57 | 6.60 | 6.56 | 6.56 |
| 4 | 6.59 | 6.59 | 6.59 | 6.55 | 6.57 | 6.56 | 6.55 |
| 5 | 6.60 | 6.58 | 6.58 | 6.56 | 6.55 | 6.57 | 6.56 |
| 6 | 6.59 | 6.58 | 6.58 | 6.56 | 6.56 | 6.56 | 6.54 |
| 7M | 6.49 | 6.45 | 6.47 | 6.43 | 6.43 | 6.45 | 6.40 |
| 7 | 6.62 | 6.61 | 6.61 | 6.61 | 6.60 | 6.61 | 6.60 |
| **Lubricated Castings** | | | | | | | |
| Casting No. | 1200-4 | 1100-6 | 1000-6 | 900-6 | 800-6 | 700-6 | 600-3 |
| Weighted Mean | 6.59 | 6.57 | 6.57 | 6.57 | 6.57 | 6.56 | 6.56 |
| Pressure, psi | 2140 | 1960 | 1780 | 1600 | 1420 | 1250 | 1070 |
| 1 | 6.61 | 6.58 | 6.59 | 6.57 | 6.60 | 6.59 | 6.58 |
| 1M | 6.45 | 6.45 | 6.46 | 6.46 | 6.41 | 6.41 | 6.40 |
| 2 | 6.60 | 6.58 | 6.58 | 6.59 | 6.58 | 6.57 | 6.57 |
| 3 | 6.59 | 6.60 | 6.60 | 6.59 | 6.59 | 6.57 | 6.57 |
| 4 | 6.59 | 6.58 | 6.58 | 6.57 | 6.54 | 6.52 | 6.54 |
| 5 | 6.59 | 6.58 | 6.58 | 6.57 | 6.59 | 6.58 | 6.57 |
| 6 | 6.57 | 6.57 | 6.56 | 6.56 | 6.55 | 6.57 | 6.54 |
| 7M | 6.48 | 6.43 | 6.44 | 6.44 | 6.44 | 6.43 | 6.40 |
| 7 | 6.61 | 6.60 | 6.60 | 6.60 | 6.59 | 6.60 | 6.60 |

Lowering of the injection pressure resulted in a slight decrease in density in all positions of the casting. The mean density for the complete casting decreased from 6.59 to 6.56 g/cc (or in other words the per cent porosity increased from 1.58 to 2.03) with

TABLE 9 — INJECTION PRESSURE EFFECT OF GAS CONTENTS OF TAILLIGHT MOUNT CASTINGS

| Position | Gas Content (S), cc/100g (STP) | | | | | | |
|---|---|---|---|---|---|---|---|
| **Nonlubricated Castings** | | | | | | | |
| Casting No. | 1200-3 | 1100-5 | 1000-5 | 900-5 | 800-5 | 700-5 | 600-2 |
| Pressure, psi | 2140 | 1960 | 1780 | 1600 | 1420 | 1250 | 1070 |
| Weighted Mean | 5.29 | — | — | 4.88 | — | — | 4.09 |
| 4 | 6.49 | — | — | 5.94 | — | — | 5.64 |
| 5 | 4.45 | — | — | 3.96 | — | — | 3.26 |
| 6 | 6.69 | — | — | 5.62 | — | — | 4.48 |
| 7M | 5.21 | — | — | 6.04 | — | — | 5.21 |
| 7 | 4.05 | 3.73 | 3.78 | 3.89 | 3.47 | 2.08 | 3.05 |
| **Lubricated Castings** | | | | | | | |
| Casting No. | 1200-4 | 1100-6 | 1000-6 | 900-6 | 800-6 | 700-6 | 600-3 |
| Pressure, psi | 2140 | 1960 | 1780 | 1600 | 1420 | 1250 | 1070 |
| Weighted Mean | 5.65 | — | — | 5.58 | — | — | 4.35 |
| 4 | 7.46 | — | — | 7.11 | — | — | 6.92 |
| 5 | 4.68 | — | — | 5.35 | — | — | 3.47 |
| 6 | 6.76 | — | — | 6.32 | — | — | 4.37 |
| 7 | 4.37 | 4.16 | 4.54 | 4.29 | 4.73 | 3.89 | 3.26 |

Blank (—) indicates determination not done.

TABLE 10 — INJECTION PRESSURE EFFECT ON THE FUNCTIONS $V_v$, $V_g$ AND USS

| Casting No. | Pressure, psi | $V_v$, cc/100g | $V_g$, cc/100g | $\frac{V_g}{V_v}$, % | USS, cc/100g | USS, % of Theoretical |
|---|---|---|---|---|---|---|
| **Position 4** | | | | | | |
| Nonlubricated | | | | | | |
| 1200-3 | 2140 | 0.26 | 0.13 | 50.0 | 0.13 | 27.2 |
| 900-5 | 1600 | 0.36 | 0.16 | 44.4 | 0.20 | 41.9 |
| 600-2 | 1070 | 0.36 | 0.23 | 63.9 | 0.13 | 27.3 |
| Lubricated | | | | | | |
| 1200-4 | 2140 | 0.26 | 0.15 | 57.7 | 0.11 | 23.1 |
| 900-6 | 1600 | 0.31 | 0.19 | 61.3 | 0.12 | 25.2 |
| 600-3 | 1070 | 0.38 | 0.28 | 73.7 | 0.10 | 21.0 |
| **Position 5** | | | | | | |
| Nonlubricated | | | | | | |
| 1200-3 | 2140 | 0.24 | 0.09 | 37.5 | 0.15 | 31.4 |
| 900-5 | 1600 | 0.34 | 0.11 | 32.4 | 0.23 | 48.2 |
| 600-2 | 1070 | 0.34 | 0.13 | 38.2 | 0.21 | 44.0 |
| Lubricated | | | | | | |
| 1200-4 | 2140 | 0.26 | 0.09 | 34.6 | 0.17 | 35.6 |
| 900-6 | 1600 | 0.31 | 0.14 | 45.2 | 0.17 | 35.6 |
| 600-3 | 1070 | 0.31 | 0.14 | 45.2 | 0.17 | 35.6 |
| **Position 6** | | | | | | |
| Nonlubricated | | | | | | |
| 1200-3 | 2140 | 0.26 | 0.14 | 53.8 | 0.12 | 25.2 |
| 900-5 | 1600 | 0.34 | 0.15 | 44.1 | 0.19 | 39.8 |
| 600-2 | 1070 | 0.38 | 0.18 | 47.4 | 0.20 | 41.9 |
| Lubricated | | | | | | |
| 1200-4 | 2140 | 0.31 | 0.14 | 45.2 | 0.17 | 35.6 |
| 900-6 | 1600 | 0.34 | 0.17 | 50.0 | 0.17 | 35.6 |
| 600-3 | 1070 | 0.38 | 0.18 | 47.4 | 0.20 | 41.9 |
| **Position 7** | | | | | | |
| Nonlubricated | | | | | | |
| 1200-3 | 2140 | 0.19 | 0.08 | 42.1 | 0.11 | 23.1 |
| 1100-5 | 1960 | 0.22 | 0.08 | 36.4 | 0.14 | 29.3 |
| 1000-5 | 1780 | 0.22 | 0.09 | 40.9 | 0.13 | 27.3 |
| 900-5 | 1600 | 0.22 | 0.11 | 50.0 | 0.11 | 23.1 |
| 800-5 | 1420 | 0.24 | 0.11 | 45.8 | 0.13 | 27.3 |
| 700-5 | 1250 | 0.22 | 0.07 | 31.8 | 0.15 | 31.4 |
| 600-2 | 1070 | 0.24 | 0.12 | 50.0 | 0.12 | 25.2 |
| Lubricated | | | | | | |
| 1200-4 | 2140 | 0.22 | 0.09 | 40.9 | 0.13 | 27.3 |
| 1100-6 | 1960 | 0.24 | 0.09 | 37.5 | 0.15 | 31.4 |
| 1000-6 | 1780 | 0.24 | 0.11 | 45.8 | 0.13 | 27.3 |
| 900-6 | 1600 | 0.24 | 0.12 | 50.0 | 0.12 | 25.2 |
| 800-6 | 1420 | 0.26 | 0.14 | 53.8 | 0.12 | 25.2 |
| 700-6 | 1250 | 0.24 | 0.13 | 54.2 | 0.11 | 23.1 |
| 600-3 | 1070 | 0.24 | 0.13 | 54.2 | 0.11 | 23.1 |
| **Position 7M** | | | | | | |
| Nonlubricated | | | | | | |
| 1200-3 | 2140 | 0.51 | 0.11 | 21.6 | 0.40 | 83.9 |
| 900-5 | 1600 | 0.66 | 0.16 | 24.2 | 0.50 | 104.8 |
| 600-2 | 1070 | 0.73 | 0.21 | 28.8 | 0.52 | 109.0 |
| Lubricated | | | | | | |
| 1200-4 | 2140 | 0.53 | n/a | | | |
| 900-6 | 1500 | 0.63 | n/a | | | |
| 600-3 | 1070 | 0.73 | n/a | | | |

n/a indicates not available.

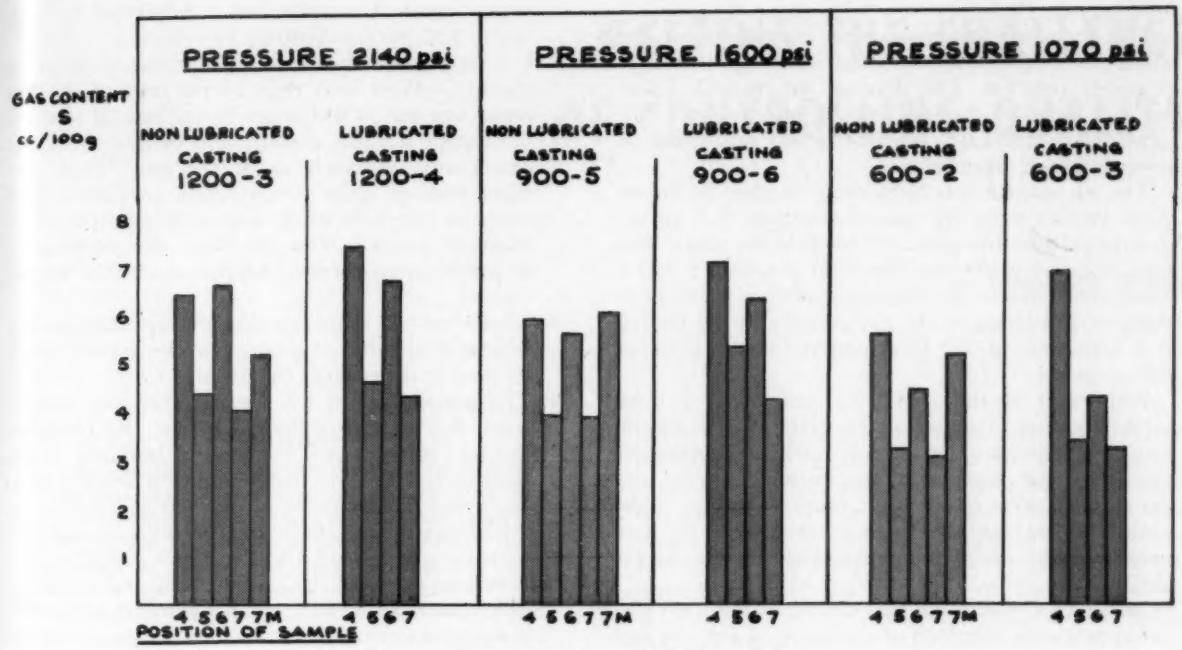



Fig. 11 — Variation of gas contents with position of castings produced at different injection pressures.

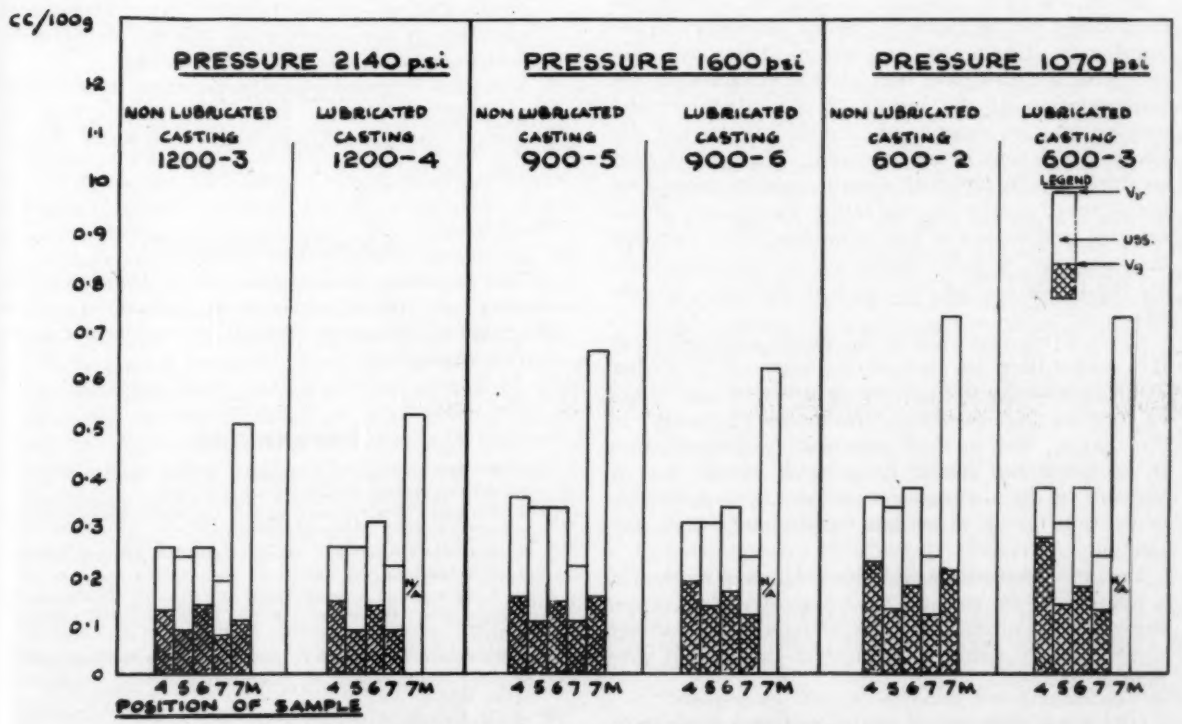



Fig. 12 — Functions $V_v$, $V_g$ and USS at various positions on castings produced at different injection pressures.

the decrease in pressure from 2140 to 1070 psi. The gas content decreased slightly with decreasing pressure, however, because of the lower pressure, the overall effect was a slight increase in its volume before freezing ($V_g$).

This increase accounted almost completely for the change in density, since the unfed shrinkage remained relatively constant. The decrease in gas with lower pressure indicates that the venting became more efficient. This apparently was due to less turbulence at lower injection pressures.

The gas content was considerably higher in the regions remote from the gate, indicating that filling took place from the gate area towards the vents. The highest gas contents were found at positions 4 and 6 which were adjacent to the major overflow wells. Thus the general pattern of the gas porosity in the casting is a reflection of the flow pattern of the alloy in filling the die.

Positions 1 M and 7 M were quite small sections at the location of bosses at the extreme sides of the castings. These samples showed much lower densities than the other portions of the casting, however, the gas contents were of the same order. Thus, as might have been expected, the high porosity level was due principally to a high unfed shrinkage in this thicker area.

As before, it was found that the use of lubricant had no appreciable effect on the density of the castings although it caused a slight increase in the gas content.

## CONCLUSION

By the use of the procedures described, commercial die castings can be examined in the laboratory to determine the distribution and nature of porosity present. Relationships have been developed allowing the interpretation of the results of the laboratory determinations to describe the conditions in the die cavity at the time of solidification. These techniques combined with standard metallographic procedures have proved useful for evaluating the quality of die castings, and may also find application for assessing the suitability of a die design and casting conditions for the production of castings with optimum properties.

Correct interpretation of the results obtained from the examination of isolated castings can be rather difficult, and the technique may prove of most value in studying and evaluating the effects of changes in die design. The method may also find application in assessing the effects resulting from the use of vacuum in die casting, and in fact some initial investigations along these lines have already been carried out.

Detailed results obtained from the examination of a number of die castings have been described. From these investigations a number of interesting findings concerning the nature of porosity in the castings were made:

1. Density measurements carried out on a large number of commercial die castings revealed a wide range of porosity levels. The densities of complete castings ranged from 6.59 to 6.36 g/cc, or, in other words, the volume per cent porosity ranged from 1.58 to 5.03. Individual sections cut from castings showed a wider range; the section showing the highest level of porosity had a density of 6.14 g/cc, i.e., 8.35 per cent porosity by volume.
2. For castings exhibiting a high degree of porosity (density 6.45 or less), over 50 per cent of the porosity was due to shrinkage. In the case of sections of extreme porosity (density 6.30 or less) shrinkage contributed as much as 70 per cent. Thus, the occurrence of gross porosity was associated with excessive unfed shrinkage due to an improper solidification pattern. This confirms the importance of maintaining correct thermal conditions in the die.
3. There was no apparent relationship between the incidence of surface porosity and the overall porosity level in the castings examined.
4. The gas content of the castings examined ranged from 4 to 8.5 cc/100 g (STP) for the complete castings, or, expressed in terms of standard conditions of temperature and pressure, it amounted to from about one-quarter to one-half the volume of the die cavity itself. Individual sections showed contents ranging from 2.5 to 15 cc/100 g (STP).
5. The gas present in the castings consisted mainly of deoxygenated air with some hydrogen and other minor constituents present. The hydrogen content (up to 25 volume per cent) can be accounted for as originating mainly from the reduction of water in the air and the decomposition of die lubricants.
6. The use of die lubricant even in excessive quantities had virtually no effect on the overall density of the castings. It did however cause appreciable increases in gas contents and changes in gas composition at some locations in the casting.
7. Variation in metal injection pressures in the range of 2140 to 1070 psi had only minor effects on density and gas contents of the two series of castings tested.

## ACKNOWLEDGMENT

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# MEETING THE CHALLENGE OF CONTROLLING QUALITY

by Lawrence S. Krueger


## ABSTRACT

**Although quality control has existed in the steel casting industry since the first castings were produced, there is need for stringent control, according to the author. He outlines the program which his company is using, highlighting the different facets. Melting, heat treating and sand quality control are covered, as are other parts of the program. The work of the Quality Control Committee is considered, as well as the training school program and the quality control records, cummunications and followup. Programs are needed that will increase the interest and responsibilities of hourly workers, establish supervision as the key to good quality control, use a minimum of paperwork and establish a mutually agreeable quality level with customers. This goal is necessary, according to the author, if the industry is to maintain a competitive position.**


## INTRODUCTION

In the past 18 months much has been said and written about quality control as it may or may not exist in the foundry industry. Local American Foundrymen's Society chapters are devoting more attention to this perplexing problem. The users of foundry products are speaking out critically concerning the inconsistency of the output as an industry. The Steel Founders' Society of American recently sent a marketing task force throughout the United States to discuss marketing problems, stressing inconsistent quality. Market research people are finding out that the resistance to greater usage of steel castings as an engineering material is twofold – cost and quality.

This is a severe criticism of present methods of production and quality control. What is done about eliminating or minimizing this criticism will determine the immediate and ultimate future of the steel casting industry. As steel foundrymen this challenge to the future growth of the industry cannot be ignored. Those who choose to ignore this problem may well find themselves in a struggle for survival.

The need for better quality control at the author's company was recognized over five years ago, and much has been done to improve the quality level of the castings produced. It is the intent here to present the procedures and techniques employed at the author's foundry which have aided in a substantial increase in quality level. Since pure technical answers do not adequately give a solution to controlling quality, the author will also present the philosophy that forms the foundation for the type of quality control program presently being employed.

## CUSTOMER FACTORS AFFECTING QUALITY CONTROL

Quality control in the industry has existed since steel castings were first produced. Admittedly, it existed in the most informal possible manner. If this is true, why devote so much of time and energy to this particular phase of the business? To find the answer to this question we must begin with the people who use our product. Excluding possible cost reductions within the foundry itself, the only motivating factor in controlling quality is the customer.

What has occurred in the customer's plants that has created all this concern for quality control? The writer believes a number of things have occurred within their plants that has enlarged the problems of quality control for steel foundries. Some of the more important points are:

1. The employment of statistical methods for receiving inspection.
2. Automatic machining.
3. Increasing use of standards and incentives.
4. The use of cost reduction committees.
5. Increasing costs for castings.
6. Establishment of tighter specifications.
7. Competitive materials and processes.

All customers have at least one of these factors present; and many have all seven factors present in their operations.

The proper use of statistics for establishment of quality control is an important useful tool for management. This is recognized to be true, since many use statistics in establishing internal controls. How-

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ever, it must be pointed out that statistical control can be abused and misapplied. Curves, graphs, charts and formulas can become more important than the engineering function of the part being subjected to this type of control. It has been the writer's experience that some customers have gone too far in this approach of using statistical quality control.

Misapplication of statistical data in quality control will accomplish two things for the customer – 1) it will build unnecessary quality into the castings being produced and 2) it will increase the foundry's cost of production, and ultimately the selling price for the castings. This feature of using statistical control warrants a much closer look than it is currently getting as many castings are being rejected on minor deviations. Both the foundry and the customer would be much better off if the establishment of receiving inspection were mutually agreed upon prior to the production of the castings.

### *Machining*

The other major change customers are making is in machining. Foundries no longer find the type of machining present where a customer can juggle a casting to compensate for variations. In the event machinery is present where compensations could be made for casting variations, objections will be raised from the customer's methods and standards procedures as established in their plant. Automatic machining is a real tartar; here the casting is virtually slammed into a machine, and completely machined in one operation. Any slight casting variation, such as warpage or core shifts, can cause the machining to be off with the end result being scrap. The writer's foundry is presently producing castings that weigh in excess of 400 lb that are completely machined in one operation in an automatic machine. This feature of automation is still in its infancy; castings users have only scratched the surface thus far. In the years to come this factor will affect quality control on all production castings that receive a machining operation.

As an industry, there are too many factors involved in production that little is known about. For example – 1) casting shrinkage as it pertains to pattern design, 2) dimension stability from casting to casting, 3) tolerances obtainable that can be held, 4) elimination of internal inclusions and defects and 5) machinability. These are all factors that are moving to the foreground in customer demands. These are the points that research and quality control procedures must conquer if the industry is to move ahead.

## MANAGEMENT'S RESPONSIBILITY IN QUALITY CONTROL

What is the industry or individual foundry going to do about these areas? Here are several possible answers:

1. The status quo can be maintained and gradually go out of business.
2. Customer demands can be yielded to without any discussion.
3. Comprehensive quality control programs can be established to obtain optimum quality.

Each of these answers by itself is probably wrong. The probable answer lies within a combination of these three answers, and is extremely difficult to determine.

Unfortunately, the greatest strides forward in quality have not been made by desire to improve product. Each time there has been an upward movement in quality, the motivating force was usually customer demands. Most foundrymen would have to look long and hard to find an area where quality was improved due to a desire to improve for improvement's sake. By this it should not be inferred that all quality improvements should have been initiated by the foundries; however, there are improvements that should be made, or have already been made, that the industry should have been responsible for.

The philosophy of following minimum quality control to get past our customer's inspection must be eliminated. Improving quality and reliability as an engineering material will ultimately increase customer faith in the foundry industry.

The establishment of a quality controlled steel foundry operation must start with the top management of the company. While the actual establishment of quality control must start at the bottom of an organization and be built up, the positive philosophy must exist at the top and filter down. Management must not only be 100 per cent in agreement, but must also insist on 100 per cent cooperation from every member of the team. This must include all phases of supervision. Negative thinking or negative actions anywhere within this group can undo all the efforts to create and guarantee a quality controlled operation.

### *Developing Company Enthusiasm*

How can foundry management create or develop enthusiasm for the improvement of quality necessary for survival in today's competitive market? There are many answers to this question – some of the more obvious ones are:

1. Management must honestly be trying to improve the quality of the castings being produced.
2. Management must provide proper facilities within which it is possible to produce quality castings.
3. Management must insist on the proper application of all scientific research results that can aid in quality control.
4. Supervision and staff personnel must get 100 per cent support in their efforts to produce better castings.
5. Management must be willing to scrap poor quality castings rather than use a calculated risk shipping policy.
6. Management must be willing to provide competent people.

These are only a few answers to this question. There are probably many more of equal importance and individual steel foundries would probably differ on the answers or importance of the answers.

Fortunately, most steel foundries are relatively small as measured in today's standards of company size. This creates the advantage that a more informal

quality control procedure can be followed. It is not only easier, but an absolute necessity, that top management in today's steel foundry take an active personal interest in all quality control problems that may exist.

The final point to be made under management's responsibilities is the role of line supervision in quality control. In the author's opinion, supervision is the key to quality control. Without the full cooperation of the supervisors, quality control will have trouble succeeding. The bulk of responsibility for quality control must be accepted by the supervisor. It has often been noted by the author that foundry supervisors are not always well informed on their own company operations. This lack of knowledge could prove detrimental in the maintenance of quality control procedures.

## BASIC QUALITY CONTROL CONCEPTS

Quality control in a steel foundry can be divided into two general categories – process quality control and operating quality control. Process quality control should include all procedures where relatively constant flow of a material is involved. These operations at a typical steel foundry might be found in this category – 1) the production of metal, 2) heat treating control and 3) the production of sands.

Operating quality control is a general description of everything that is done, exclusive of process control, to produce a quality casting. Every foundry will differ quite widely in what procedures will be followed in establishing their operating quality controls. For the purpose here, the author will discuss the procedures that have aided in attaining a measurable improvement in the quality of the castings produced. The steps taken to accomplish this improvement were:

1. Creation of a quality control team of supervisors to maintain the procedures that were established.
2. Establishment of training schools for foundry personnel to both improve their workmanship, and to increase the degree of their responsibility for quality.
3. Improvement of communications and records that pertain to quality control.
4. Development of a follow-up procedure on problems pertaining to quality control.

## MELT DEPARTMENT QUALITY CONTROL

Prior to attempting the solution of quality control problems created by increasing customer demands or requirements, today's steel foundry must have established positive quality control of the metal being produced. If inconsistent values of chemical and physical properties of the steel being produced exist, all other quality control procedures will prove to be of little or no value. The production of metal must be the starting point in the analysis or establishment of any foundry quality control procedure.

The author's company produces carbon and low alloy steels at a maximum rate of approximately 700 tons per month. Seven types of steels are produced in an acid electric furnace, ranging from dynamo steel to a molybdenum-manganese steel having 30 to 40 points of carbon. Off analysis chemistry, which could cause castings to be scrapped, amounted to 17,440 lb during 1960. This was accomplished while producing 2,616 heats of approximately 9000 lb each. The development of a high degree of quality control in melting can be attributed to:

1. Melters and helpers are paid on a quality control incentive system.
2. Purchasing of scrap is highly selective.
3. The size of all heats melted is approximately 9000 lb.
4. On all heats melting procedure is the same until additions are made to obtain the final chemistry.
5. A system of checks and records of procedures are maintained to give a history of every heat produced.

These factors, combined with the proper training of Melt Department personnel, have established a high degree of quality control in melting.

Figure 1 is a typical electric furnace log sheet employed by the author's foundry. The log sheet is filled out by the melter and contains a complete history of each heat produced. The log sheet is the first step in an adequate quality control procedure. The record of what has been done is important in the maintenance of a quality control procedure. In conjunction with this, a recording instrument is wired in the electric furnace curcuits to record the time on each heat, plus the time each tap of the transformer

| | 1 | 2 | 3 | 4 | 5 | 6 |
|---|---|---|---|---|---|---|
| Heat Number | 357 | 358 | 359 | 360 | 361 | 362 |
| Type of Steel | 25 Max. | 35/40 | 25 Max. | 40/50 | Dynamo | 35/40 |
| Meter Difference | 2640 | 2480 | 2320 | 2280 | 2320 | 2300 |
| K.W.H. per Ton | 660 | 620 | 580 | 570 | 580 | 575 |
| Electrodes | | 170 | | | 170 | |
| Number of Present Roof | 2 | 2 | 2 | 2 | 2 | 2 |
| Number of Heats Pres. Roof | 20 | 21 | 22 | 23 | 24 | 25 |
| Number of Heats Pres. Lin. | 20 | 21 | 22 | 23 | 24 | 25 |
| Number of Heats Pres. Ladle | 6 | 7 | 8 | 9 | 10 | 11 |
| Time power on | 6:45A | 8:11A | 9:42A | 11:02A | 12:28P | 2:09P |
| Time power off | 8:07A | 9:36A | 10:55A | 12:20P | 1:55P | 3:20P |
| Heat Tapped | 8:09A | 9:36A | 10:56A | 12:23P | 1:58P | 3:22P |
| | | | | | | |
| FURNACE ADDITIONS: | | | | | | |
| Purchased Scrap | 3900 | 4200 | 3700 | 3600 | 4400 | 3500 |
| Heads and Gates | 5100 | 4600 | 4300 | 4300 | 4600 | 4900 |
| Pigs and Spills | | 200 | 1000 | 1100 | | 500 |
| Scrap Castings | | | | | | |
| Sub-Total | 9000 | 9000 | 9000 | 9000 | 9000 | 8900 |
| Molybdenum Oxide | | | | | | |
| Mill Scale | | | | | | |
| Iron Ore | 250 | 250 | 250 | 250 | 450 | 250 |
| | | | | | | |
| Total Furnace Additions | 9250 | 9250 | 9250 | 9250 | 9450 | 9150 |
| Ladle Additions | | | | | | |
| Manganese - Lump | 73 | 68 | 73 | 67 | | 68 |
| Pulverized | | | | | | |
| Silicon | 53 | 52 | 55 | 50 | 125 | .30 |
| Ferro Molybdenum | | | | | | |
| Boron | | 2 | | 2 | | 2 |
| Aluminum | 9.5 | 9.5 | 9.5 | 9.5 | 40 | 9.5 |
| | | | | | | |
| | | | | | | |
| | | | | | | |
| Total Ladle additions | 135.5 | 131.5 | 137.5 | 128.5 | 165 | 129.5 |
| | | | | | | |
| | | | | | | |
| Furnace Operator | Felix | Felix | Felix | Felix | Felix | Stan. |

Fig. 1 — Electric furnace record.

was being used. This merely serves as a check point to the power on and off as recorded by the melter, plus the elapsed time between heats.

The quality plan under which the melters operate has three parts – chemical analysis of the heat, man-hours per heat for the department and kwh/ton melted. This plan was reported upon in AFS TRANSACTIONS in 1952 by Tetzlaff.[1] The portion of the plan to be considered here is the incentive on chemical analysis, which is the most important part both from a quality control standpoint and potential incentive earnings for the melters. In carbon steel heats three elements are held within narrow limits by the melters to obtain a bonus for a quality performance. These are – carbon, manganese and silicon. The bonus ranges for these elements on a 0.40/0.50 carbon steel heat are:

Carbon .....0.41 to 0.49
Manganese .....0.65 to 0.75
Silicon .....0.40 to 0.50

Any values above or below these ranges inflict a penalty to earnings where the range has been held on another of the elements. Obviously the heat would not be scrapped if manganese or silicon were somewhat out of range. The point here is that melters can hold these ranges when their incentive pay is dependent upon it. Figure 2 is a daily laboratory report on the chemistry of the previous day's heats. This report is available the morning after the day the heats were produced, and immediately indicates any tendency for a drop in the quality of the steel being produced.

| Date: 1-25-1961 | | | | | | | | | | | |
|---|---|---|---|---|---|---|---|---|---|---|---|
| Spec. | No. | Prel. C. | C. | Mn. | Si. | P. | Sul. | Cr. | Ni. | Mo. | Remarks |
| 25 Max | 357 | 16 | 24 | 65 | 38 | | 042 | | | | |
| 35/40 | 358 | 34 | 35 | 68 | 59 | | 043 | | | | |
| 25 Max | 359 | 12 | 25 | 67 | 47 | | | | | | |
| 40/50 | 360 | 35 | 42 | 67 | 55 | | | | | | |
| Dyn. | 361 | - | 09 | 13 | 62 | | 041 | | | | |
| 35/40 | 362 | 30 | 33 | 63 | 38 | | | | | | |
| 25 Max | 363 | 15 | 25 | 69 | 50 | | 044 | | | | |
| 25 Max | 364 | 15 | 23 | 70 | 50 | | | | | | |
| 25 Max | 365 | 14 | 27 | 75 | 55 | | | | | | |
| 35/40 | 366 | 30 | 36 | 70 | 52 | | | | | | |
| 25 Max | 367 | 15 | 22 | 67 | 49 | | | | | | |
| | | | | | | | | | | | |
| | | | | | | | | | | | |

Fig. 2 — Laboratory report.

### *Statistical Data Development*

To demonstrate the results of using a quality control incentive plan for melters, statistical data were developed for the first 60 heats of 0.40/0.50 carbon steel produced in 1960. Table 1 is a compilation of the actual chemistry of these 60 heats, and Figs. 3 and 4 are control charts developed from the data of Table 1. These charts were developed according to procedures described by Johnson and Fisher[2] in AFS TRANSACTIONS in 1948. While these charts indicate the process is in control they do not indicate whether or not customer chemical requirements are being met. To check this Table 2 was developed. This table is

TABLE 1 — C-Mn-Si CHEMISTRY OF 60 CONSECUTIVE HEATS OF 0.40-0.50 CARBON STEEL

| | Analysis, % | | | | Analysis, % | | |
|---|---|---|---|---|---|---|---|
| Heat No. | C | Mn | Si | Heat No. | C | Mn | Si |
| 217 | 0.43 | 0.73 | 0.50 | 561 | 0.44 | 0.69 | 0.40 |
| 236 | 0.42 | 0.67 | 0.53 | 570 | 0.46 | 0.67 | 0.44 |
| 243 | 0.46 | 0.71 | 0.48 | 592 | 0.39 | 0.72 | 0.35 |
| 266 | 0.44 | 0.68 | 0.42 | 619 | 0.43 | 0.66 | 0.35 |
| 271 | 0.46 | 0.72 | 0.43 | 644 | 0.47 | 0.65 | 0.37 |
| 274 | 0.43 | 0.67 | 0.44 | 666 | 0.48 | 0.72 | 0.53 |
| 282 | 0.48 | 0.65 | 0.50 | 669 | 0.46 | 0.70 | 0.35 |
| 302 | 0.42 | 0.72 | 0.47 | 691 | 0.43 | 0.68 | 0.42 |
| 330 | 0.46 | 0.66 | 0.50 | 697 | 0.34 | 0.63 | 0.44 |
| 337 | 0.46 | 0.72 | 0.37 | 704 | 0.45 | 0.74 | 0.45 |
| 355 | 0.43 | 0.66 | 0.46 | 724 | 0.48 | 0.73 | 0.43 |
| 363 | 0.47 | 0.66 | 0.36 | 727 | 0.48 | 0.73 | 0.45 |
| 393 | 0.42 | 0.69 | 0.50 | 731 | 0.46 | 0.68 | 0.46 |
| 397 | 0.44 | 0.65 | 0.45 | 745 | 0.45 | 0.67 | 0.33 |
| 419 | 0.43 | 0.70 | 0.47 | 749 | 0.46 | 0.66 | 0.49 |
| 422 | 0.46 | 0.71 | 0.42 | 759 | 0.48 | 0.66 | 0.43 |
| 425 | 0.44 | 0.73 | 0.41 | 770 | 0.43 | 0.65 | 0.48 |
| 440 | 0.44 | 0.73 | 0.47 | 784 | 0.43 | 0.68 | 0.49 |
| 446 | 0.42 | 0.77 | 0.58 | 789 | 0.43 | 0.77 | 0.48 |
| 449 | 0.46 | 0.71 | 0.43 | 825 | 0.41 | 0.65 | 0.40 |
| 452 | 0.47 | 0.71 | 0.47 | 829 | 0.43 | 0.72 | 0.54 |
| 457 | 0.48 | 0.73 | 0.52 | 855 | 0.46 | 0.65 | 0.47 |
| 462 | 0.44 | 0.70 | 0.52 | 887 | 0.42 | 0.66 | 0.44 |
| 468 | 0.43 | 0.66 | 0.44 | 898 | 0.43 | 0.71 | 0.47 |
| 473 | 0.44 | 0.67 | 0.39 | 927 | 0.41 | 0.69 | 0.46 |
| 480 | 0.42 | 0.69 | 0.42 | 945 | 0.44 | 0.70 | 0.44 |
| 483 | 0.42 | 0.70 | 0.49 | 949 | 0.43 | 0.68 | 0.48 |
| 492 | 0.41 | 0.68 | 0.40 | 967 | 0.48 | 0.73 | 0.50 |
| 515 | 0.46 | 0.80 | 0.54 | 985 | 0.42 | 0.72 | 0.61 |
| 531 | 0.41 | 0.67 | 0.37 | 1007 | 0.45 | 0.68 | 0.38 |

a complete list of specifications for 0.40/0.50 carbon steel under which the author's foundry is operating. Examination of Table 2 indicates that specification no. 3 is the severest from a chemical standpoint. It is the author's opinion that a jobbing steel foundry should establish quality controls based upon the most rigid melting specification for the particular steel being produced, and produce all heats of that analysis to that specification. The interpretation here should be based upon steels that are regularly produced. Exceptional specifications only occasionally produced could be handled in a special manner. Using this criterion a second review of the statistical data indicate that the heats produced and tabluated in Table 1 met this requirement.

TABLE 2 — CUSTOMER SPECIFICATIONS FOR 0.40-0.50 CARBON STEEL

| | Range, % | | |
|---|---|---|---|
| Customer Spec. No. | C | Mn | Si |
| ACM 180 | 0.40 - 0.50 | 0.60 - 0.90 | 0.60 Max. |
| CH 186-6 | 0.40 - 0.50 | 0.60 - 0.80 | 0.60 Max. |
| 1 E 3 | 0.40 - 0.50 | 0.65 - 0.85 | 0.60 Max. |
| PH 2 | 0.40 - 0.50 | 0.50 - 0.90 | 0.20 - 0.60 |
| SC 3 | 0.40 - 0.50 | 0.50 - 0.90 | 0.20 - 0.60 |
| 1045 S 3 | 0.40 - 0.50 | 0.60 - 0.85 | Not specified |

While melters are not paid incentives on physical properties of the steel produced, it would be possible to analyze the results in the same manner as the chemistry. Physical properties of the types of steel

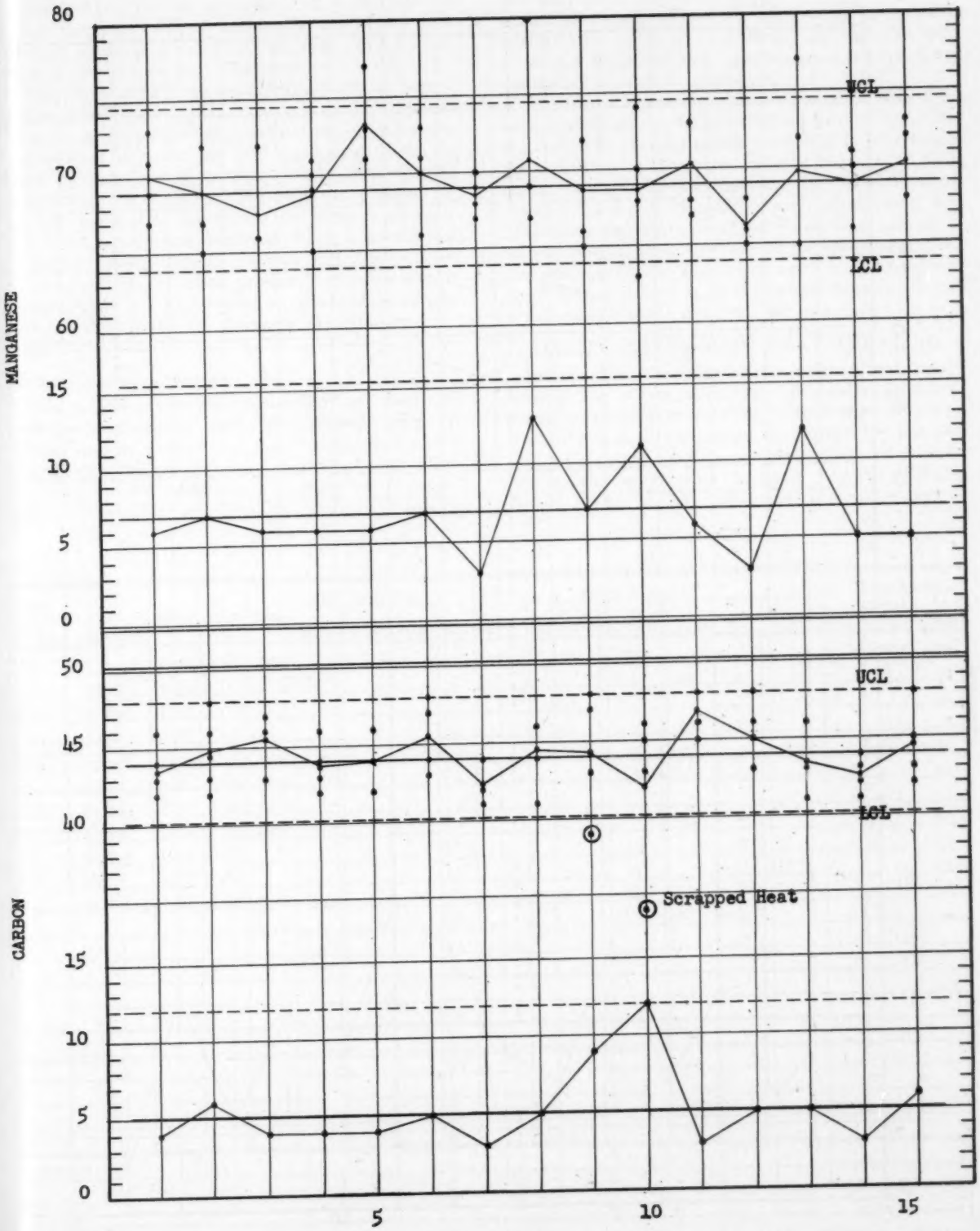



Fig. 3 — Control chart developed from data in Table 1.

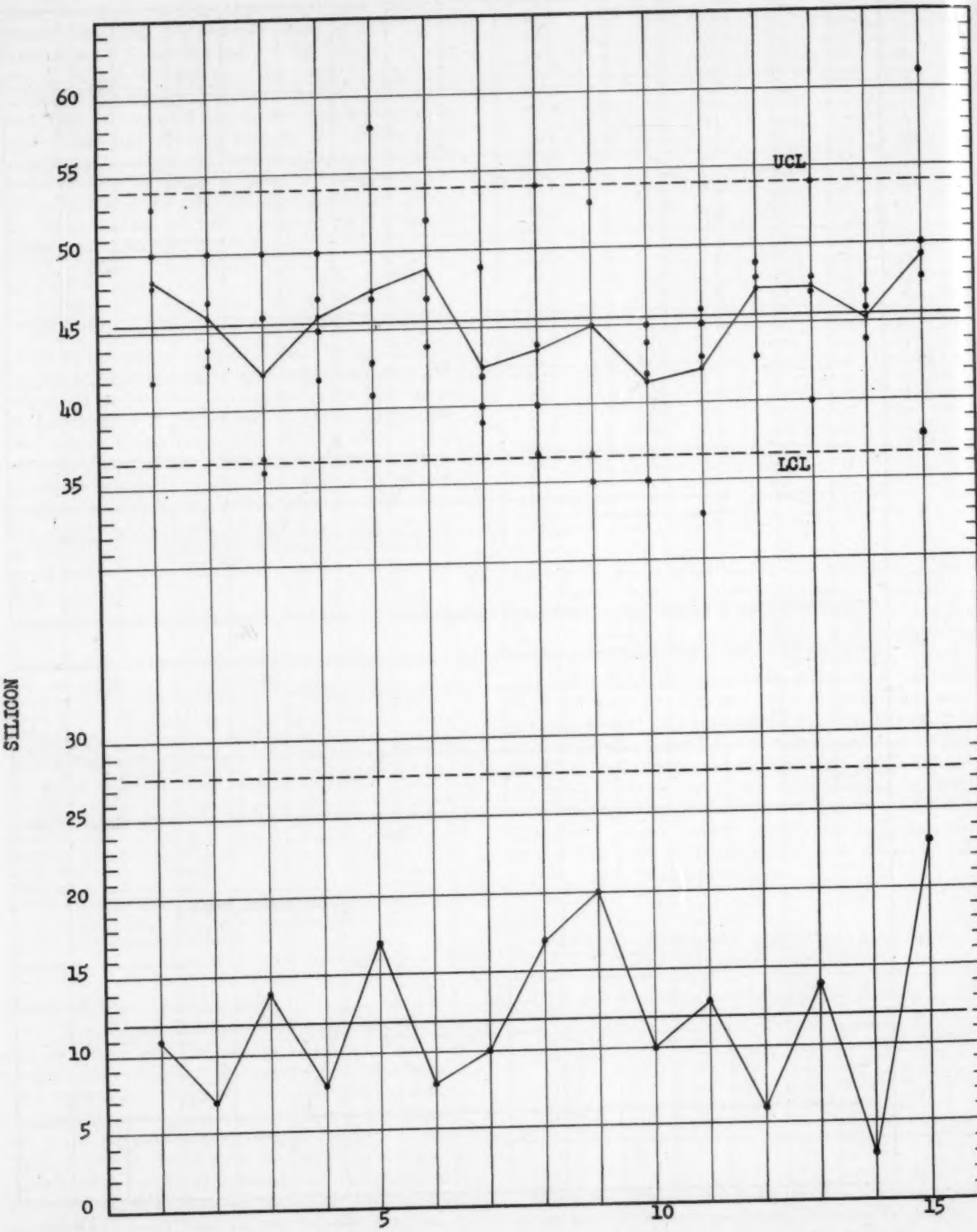



Fig. 4 — Control chart developed from data in Table 1.

being produced at the author's company are dependent upon two major factors – the heat treatment of the steel and the actual melting procedure in producing the heat. The point of concern to the Melting Department is one of procedure in producing the heats. Since procedures are firmly established and adhered to physical properties are controlled by employing the proper heat treatment.

The purchase of scrap for melting purposes cannot be minimized if good quality control is to be achieved. Each car of scrap received is visually inspected for size and type of scrap. If the purchased scrap does not meet specifications it is immediately rejected and returned to the supplier. The only other visual inspection made is to insure that the scrap is all plate and is not oily. A final chemical check is maintained on sulfur. This is done by running sulfur on at least 50 per cent of the heats produced. Since sulfur is accumulative, any high sulfur scrap would be caught within a day or two by observing the laboratory reports.

### *Melting Quality Control*

Two factors previously maintained are closely related to maintaining uniform quality control in melting—uniform heat size and standard melting procedures. Uniform heat size, combined with selective scrap purchasing, allows a standard melting procedure to be developed. This procedure is to reduce carbon, silicon and manganese to pre-determined low levels, then ferro additions of silicon and manganese are added to obtain the desired result. Carbon is added by dipping electrodes to a definite depth in the bath for a calculated period of time. This procedure has resulted in the high level of quality performance previously mentioned. Along with this it should be noted that operating costs have been controlled as well as quality.

The final checkpoint in melting quality control is done during the actual pouring of molds. The pouring crew uses the heat record form (Fig. 5) to indicate the condition of the ladles and the heat itself. This report, along with the furnace log sheet, is kept together as a daily history of melt production. An immersion thermocouple, Figs. 6 and 7, is also used by the pouring crew to check metal temperatures in the pouring ladles. This instrument is not used on every heat, it is used as an occasional checking device to insure hot heats from the melting unit.

## QUALITY CONTROL IN HEAT TREATING

In carbon and low alloy steels, designers and metallurgists are taking advantage of greatly increased strengths through closely controlled heat treating. Due to this need for higher strengths there has developed a strong trend to narrower allowable Brinell ranges. Customers who formerly allowed Brinell ranges of 50 to 60 points are currently specify-

| Melt Heat No. | Analysis | Time In | Time Out | Temp. | Ladle Condition | | | No. | Molds Poured | | | | | | Shell | Metal Pig. | Min. to Pour |
|---|---|---|---|---|---|---|---|---|---|---|---|---|---|---|---|---|---|
| | | | | | Bull | 1 Ton | 500# | | Bench | Squeezer | Slinger | P.L. | Fl. | D.S. | | | |
| 357 | 25 Max | 8:12 | 8:40 | 2900 | | | | 1 | 9 | 103 | 26 | | | | | 150 | 28 |
| 358 | 35/40 | 9:40 | 10:28 | 2890 | | | | 2 | 17 | | 4 | 64 | 2 | | | | 48 |
| 359 | 25 Max | 11:00 | 11:45 | 2910 | | | | 3 | 9 | 253 | | | | | 49 | | 45 |
| 360 | 40/50 | 12:25 | 1:00 | | | | | 4 | | 132 | 4 | 43 | 1 | | | 75 | 35 |
| 361 | Dynamo | 2:05 | 2:35 | 2895 | | | | 5 | | 1 | 19 | | | | | 150 | 30 |
| 362 | 35/40 | 3:24 | 3:50 | 2920 | | | | 6 | 11 | | 9 | | 4 | | | | 26 |
| 363 | 25 Max | 7:42 | 8:25 | 2920 | | | | 7 | | 216 | | | | | | 125 | 43 |
| 364 | 25 Max | 8:55 | 9:35 | 2930 | | | | 8 | 12 | 62 | | 30 | 6 | | | | 40 |
| 365 | 25 Max | 10:17 | 10:50 | | | | | 9 | | | 11 | | 7 | | | | 33 |
| 366 | 35/40 | 11:37 | 12:00 | | | | | 10 | | | 16 | | | | | | 23 |
| 367 | 25 Max | 12:52 | 1:30 | 2920 | | | | 11 | | 96 | 14 | | | | | | 38 |
| | | | | | | | | 12 | | | | | | | | | |
| | | | | | | | | 13 | | | | | | | | | |
| | | | | | | | | 14 | | | | | | | | | |
| | | | | | | | | 15 | | | | | | | | | |
| | | | | | | | | 16 | | | | | | | | | |
| | | | | | | | | 17 | | | | | | | | | |
| | | | | | | | | 18 | | | | | | | | | |
| | | | | | | | | 19 | | | | | | | | | |
| | | | | | | | | 20 | | | | | | | | | |
| | | | | | | | | 21 | | | | | | | | | |

REMARKS:

Fig. 5 — Heat record of January 25, 1961.



Figs. 6 and 7 — Immersion thermocouple in use by pouring crew.

TABLE 3

| Casting No. | Bhn | Casting No. | Bhn |
|---|---|---|---|
| 1 | 311 | 26 | 306 |
| 2 | 311 | 27 | 306 |
| 3 | 311 | 28 | 311 |
| 4 | 311 | 29 | 311 |
| 5 | 302 | 30 | 285 |
| 6 | 302 | 31 | 285 |
| 7 | 302 | 32 | 311 |
| 8 | 285 | 33 | 302 |
| 9 | 302 | 34 | 302 |
| 10 | 311 | 35 | 302 |
| 11 | 302 | 36 | 302 |
| 12 | 302 | 37 | 311 |
| 13 | 285 | 38 | 285 |
| 14 | 285 | 39 | 302 |
| 15 | 302 | 40 | 302 |
| 16 | 302 | 41 | 311 |
| 17 | 302 | 42 | 302 |
| 18 | 302 | 43 | 311 |
| 19 | 302 | 44 | 311 |
| 20 | 302 | 45 | 311 |
| 21 | 302 | 46 | 285 |
| 22 | 285 | 47 | 302 |
| 23 | 302 | 48 | 302 |
| 24 | 302 | 49 | 285 |
| 25 | 302 | 50 | 311 |

ing ranges of 30 to 40 points. In order to meet this increasing demand, it is necessary to establish a quality controlled heat treating operation.

The first essential in a quality controlled heat treating department is a thorough knowledge of the chemistry of the steel. Rigid melting controls will substantially reduce difficulties in heat treating. It is the author's opinion that heat treating facilities should be an integral part of a steel foundry operation. This will accomplish two things—control of quality will be more easily accomplished and the cost of heat treating will be reduced.

Heat treating to Brinell specifications lends itself to statistical quality control. Table 3 contains the Brinell readings of 50 castings of a molybdenum-manganese steel that is liquid quenched and drawn to a Brinell range of 280-320. The plotting of statistical curves would prove to be of little value, since the customer specification will determine the upper and lower control limits. In this type of application, after a 100 per cent inspection demonstrates all Brinell readings are within the specified range a sampling procedure should be established to maintain control.

At the author's company, two methods of sampling are employed. If the customer has established a receiving inspection standard of sampling a given percentage of each shipment, the same procedure will be followed by the author's company's Inspection Department. If there is a lack of information available sampling will be done on a percentage basis, usually in the area of 10 per cent. One hundred per cent Brinell inspection has an extremely high cost factor, and is only employed in these instances:

1. On the first order or two on a new job until a proper sampling procedure has been established.
2. On the first order or two on an existing job where the specifications are changed.
3. On any job where a heat treating problem has developed. This will remain 100 per cent until the trouble is eliminated.
4. Where customer specifications require it, and an additional cost is charged to the job in quoting.

The remaining heat treating procedures include the usual standard operating procedures employed by most steel foundries. All heat treating equipment contains recording temperature instrumentation to serve as a check point on the heat treating operators. Since heat treating is a function of temperature and time based upon chemistry all heat treating specifications are established by the Metallurgical Department.

## QUALITY CONTROL OF SANDS

The necessity for adequate quality control of sands and bonding materials is known by all steel foundrymen. On a weight or volume basis this is the largest single material used in today's steel foundry. Therefore, it is logical that a great deal of time and effort has been expended in research to produce the best possible economic sand mixes. The maintenance of the performance of the sand mixes is a function of quality control. The degree of quality control applied to sands must, to a great extent, be dependent upon the class of work being produced by the individual steel foundry. Because of this, it is

| Molding Station | Moisture | Perm. | Green Strength | Hardness | Where Tested | Remarks |
|---|---|---|---|---|---|---|
| Squeezer | 2.50 | 170 | 6.3 | 82 | Station | Reclaimed Facing |
| Squeezer | 2.40 | 180 | 6.1 | 83 | Station | Reclaimed Facing |
| Squeezer | 2.60 | 170 | 6.3 | 83 | Station | Reclaimed Facing |
| Squeezer | 2.60 | 170 | 5.9 | 82 | Station | Reclaimed Facing |
| Squeezer | 2.30 | 180 | 6.4 | 83 | Station | Reclaimed Facing |
| Squeezer | 2.40 | 180 | 6.2 | 81 | Station | Reclaimed Facing |
| Floor | 3.60 | 130 | 5.6 | 81 | Station | Reclaimed Facing |
| Floor | 3.40 | 140 | 5.4 | 82 | Station | Reclaimed Facing |
| Floor | 3.60 | 130 | 5.3 | 81 | Station | Reclaimed Facing |
| Floor | 3.30 | 140 | 5.7 | 82 | Station | Reclaimed Facing |
| Blower | 5.40 | 60 | 3.3 | 72 | Mill | New Sand |
| Blower | 3.80 | 100 | 2.8 | 69 | Station | New Sand |
| Blower | 3.20 | 110 | 2.2 | 66 | Station | New Sand |
| Blower | 3.00 | 110 | 2.1 | 65 | Station | New Sand |
| | | | | | | |

Fig. 8 — Daily sand report form, of January 25, 1961.

impossible to have firmly estalished quality control procedures that will apply to all steel foundries.

The author's foundry produces castings up to approximately 3,500 lb in weight, using loose and production pattern equipment on snap flask molding units and tight flask molding units. On all molding a dry reclaimed facing sand is used back up with heap sand. All core sands are made with new sand which is an AFS 63 to 68 screened and dry Portage sand. The desired properties of molding sand vary with the type of casting being produced, and with the type of molding being employed to produce the castings. The basic tests employed on molding sands are — 1) moisture, 2) green strength, 3) pH, 4) standard ram hardness and 5) permeability. Core sand mixes are checked more frequently than molding sands, and in addition to green properties, core sands are checked for dry properties.

Once again it can be pointed out that records of what has been done are essential to a good quality control procedure. As in melting, the author's company maintains adequate records of sand mill performance to insure that proper operating procedures are followed. Two daily reports are maintained by the Sand Technician; 1) the daily sand report (as illustrated in Fig. 8) which is a random check of facing sand properties, the samples to be checked are usually obtained at the molding station and 2) a record of the total sand production by types of sand (Fig. 9). In conjunction with this a recording strip chart is wired into each sand mill electrical circuit as a check point on mulling time for each batch of sand. Figure 10 is a chart taken off of a sand muller.

| Type of Mix | 1st Shift | 2nd Shift | 3rd Shift | Total No. Mixes |
|---|---|---|---|---|
| Heap Sand | 169 | 85 | | 254 |
| Facing | 38 | 12 | | 50 |
| Strong Oil Sand | 1 | 1 | | 2 |
| Collapsible Oil Sand | 3 | | | 3 |
| Blower Sand | 5 | | | 5 |
| Zirconite Core Sand | | | | |
| Ladle Sand | 1 | | | 1 |
| Special Sand | | | | |
| Down Time Mill No. 1 | 28 Min. | 1 Hr. | | 1 Hr. 28 Min. |
| Down Time Mill No. 2 | 26 Min. | 1 Hr. 14 M. | | 1 Hr. 40 Min. |
| Remarks: | | | | |

Fig. 9 — Sand laboratory record of daily mill mixes, of January 25, 1961.

The amount of checking is fairly constant as pertains to sands; approximately 50 per cent of the batches of core sand are checked for green properties, while approximately 25 per cent of the batches of reclaimed facing sands are checked. The best criterion of adequate quality control in sands is the casting quality in the cleaning room. Therefore, sand technicians are required to constantly check castings prior to head and gate removal, to prevent any tendency for a sand defect becoming a major problem. This is pointed out here because the author feels it is much too easy to blame quality problems on sands when the casting is partially processed by the cleaning room. Oftentimes a casting with headsand gates on it will point to the fact that a quality problem may be one that involves the level of workmanship rather than the quality of the sand.

### *Sand Screen Analysis*

The final point to be made concerning sands is the screen analysis checks being made by the author's company. The screen analysis of every car of new sand received is checked to determine that it meets the specification. Reclaimed sand and heap sand are regularly checked to insure that screen distribution and screen analysis are within the operating ranges that have been established. Figure 11 is a visual type of report that is posted in the sand laboratory where everyone can see it. This gives the AFS number as well as the screen distribution of the reclaimed and heap

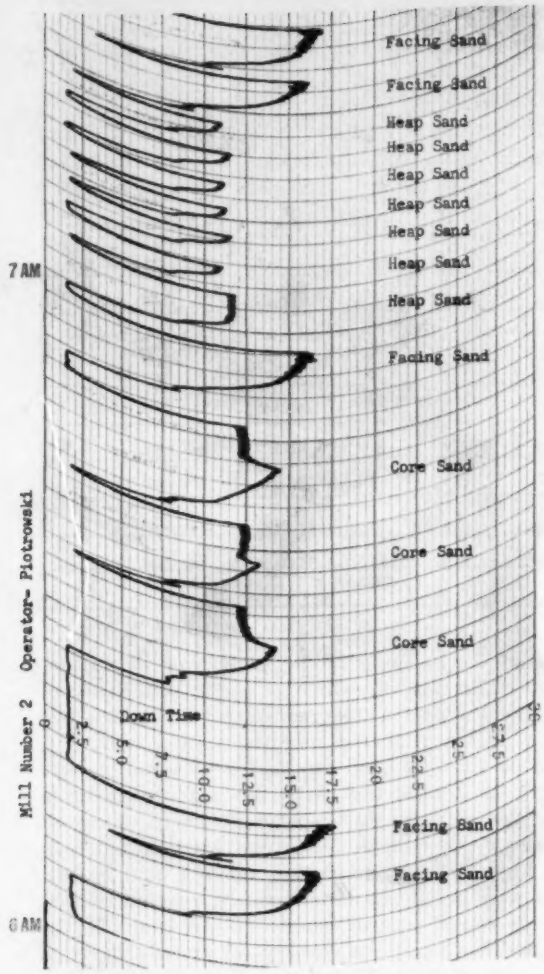



Fig. 10 — Chart taken off sand muller.

sands. It is felt that a visual type of report is better, since the data are more easily read. Another advantage is the ease of comparison to previous reports to observe any tendency for a shift in screen analysis.

While sand control at the author's foundry is not as brief as described here, it is felt that great detail on this subject would be repetition of many previous papers and reports by others. It is hoped that indicating the type of control employed will suffice on the subject of sand quality control.

## QUALITY CONTROL COMMITTEE

The establishment of quality control responsibilities can vary quite widely from company to company. It is believed that controlling quality is a prime responsibility of supervision; however, most steel foundry supervisory groups are fairly large, and would be difficult to work with in its entirety. Because of this problem, it is felt that a quality control team or committee should be established from within the supervisory and management group. The group should be as small as possible, and should consist of those people who would normally have a wider knowledge of the overall quality control problems.

With this premise in mind the author's company established a quality control committee to act as an advisory group in establishing and maintaining quality control procedures. The members of this committee are the works manager, plant metallurgist, chief inspector and quality control supervisor.

The purpose of this committee is to act as a central clearing house for supervision in the matters of quality control. Each individual on the committee watches specific areas of the quality control program, and handles the problems that develop within that area. The committee in no way circumvents supervision, nor does the committee relieve the responsibility of supervision for quality control.

This committee approach has resulted in improved quality control. The paper work involved in estab-

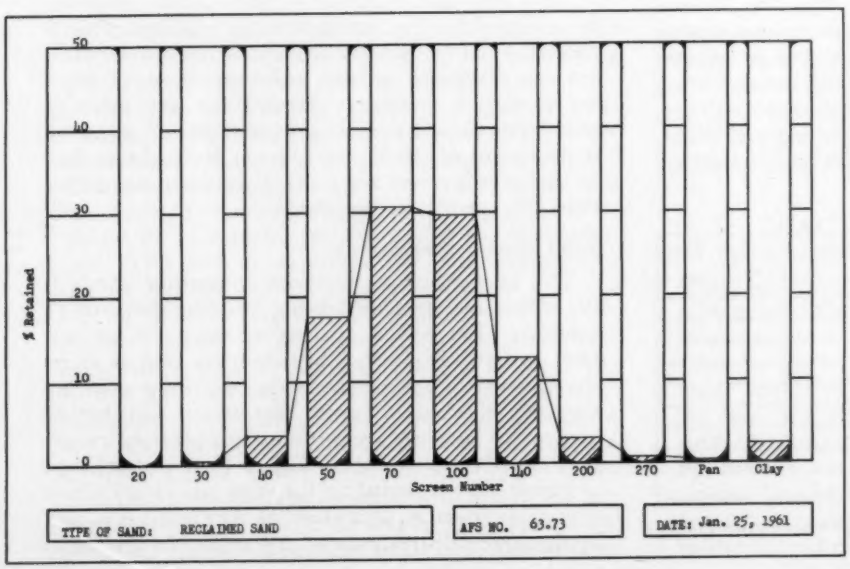



Fig. 11 — Visual type report posted in sand laboratory.

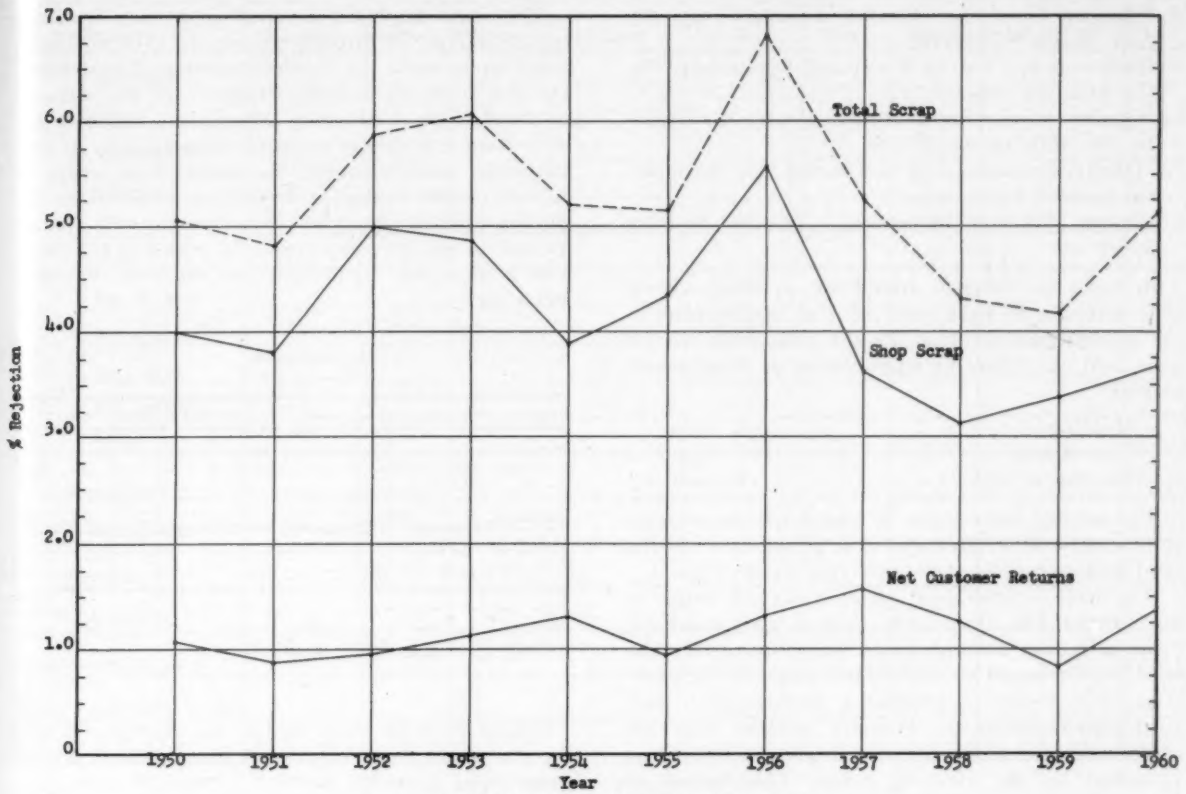



Fig. 12 — Chart showing need for quality control.

lishing and maintaining quality control procedures has been minimized. The results of this effort are hard to evaluate in dollars and cents, since customer relations cannot be evaluated in terms of money. To demonstrate the need for quality control, Fig. 12 was developed. This shows the percentage of net customer returns, shop scrap and total scrap for the author's company for the last ten years. The general trend prior to 1956 was up while the trend since 1956 has been down; 1960 shows a reverse in trend again.

In the author's opinion, the reversal in 1960 was due to a substantial increase in customer requirements. It has been the author's experience that customer returns will closely follow shop scrap—as internal scrap increases, customer returns will also increase.

## TRAINING SCHOOLS FOR IMPROVED QUALITY

One of the first problems that arose in establishing improved quality control was the need to increase the interests and amount of responsibility of the hourly workers. It was felt that although supervision is the key to quality control the hourly worker has a high degree of responsibility for the level of quality that can be obtained. Any method that will increase his desire to do a better job, or increase his pride of workmanship, is highly desirable.

It was decided to run a training school on a voluntary basis to determine the amount of interest the hourly man had in his work. The school was established on Saturday mornings, with attendance strictly optional on the part of the men involved. No one was paid for attending, although free lunch was served at the conclusion of the sessions. Since the cleaning room cannot improve much on the castings it received from the foundry, the decision was made to concentrate the effort into three departments—coremaking, molding and pouring. Incidentally, no one within the organization was excluded from attendance at any session regardless of the department they worked for.

Four sessions were planned for four consecutive Saturday mornings—no. 1—sand and its use, no. 2—molding practice, no. 3—core room practice and no. 4—pouring.

All lectures were prepared and delivered by the supervisors from the quality control committee and departments involved. Each session was allowed ample time for a question and answer period, and all questions asked were answered.

The following is a complete list of the lectures presented at the four classes held.

1. "Why We Use the Sands We Do," by the Plant Metallurgist.
2. "The Proper Use of Sands for Quality," by the Quality Control Supervisor.
3. "Essentials to Good Cores," by the Plant Metallurgist.

4. "Quality Cores for Quality Castings," by the Core Room Supervisor.
5. "Selection and Use of Tools and Equipment," by the Molding Supervisor.
6. "Quality Workmanship Equals Quality Castings," by the Molding Supervisor.
7. "Quality Considerations in Pouring," by the Quality Control Supervisor.
8. "Proper Pouring Procedures," by the Pouring Supervisor.

In order to correlate attendance at these schools it is necessary to have some idea of employment in the departments involved. At the time these schools were held, the following employment by departments existed.

Molding ..... 35 men
Core Room ..... 15 men
Pouring & Indirect ..... 45 men

The average attendance at school sessions was approximately 50 hourly men, with a high of 65 at the sand session to a low of 42 at the core room session.

The lecture series paid off in a greater desire to do a better job. Shop scrap showed an appreciable drop after the lecture series was concluded. The hourly men began to understand some of the problems encountered in producing castings that meet customer requirements. Foundry workers were observed taking a keener interest in the castings being processed by the cleaning room. The feeling of being an important part of the organization was imparted to the men who attended the lecture series. All of these observations has sold the author's company on the need for putting more trust and responsibility in the hourly workers' hands. An informed group of hourly people under a qualified supervisor cannot help but improve the products produced.

## QUALITY CONTROL RECORDS, COMMUNICATIONS AND FOLLOWUP

One of the most important areas of quality control is in the necessary records being maintained, and the means of communications. If a truly workable quality control procedure is to be established, the records kept must be the minimum required to do the job. Communications must be such that everyone concerned with any problem is immediately informed about the problem. To accomplish this is extremely difficult, and the author cannot say the present system employed meets these requirements. What has been developed thus far has aided in reducing confusion; however, the system does not always perform to perfection.

At the author's company all records pertaining to shop scrap and customer returns are kept on a monthly basis. Daily reports are made, and the cumulative records are maintained. While some foundries advocate posting shop scrap results at the various molding stations, the author's company maintains the records on a bulletin board at the shop entrance. Each casting returned from a customer is reviewed by the supervisors involved in the production of the particular casting.

The forms employed to report shop scrap are illustrated in Figs. 13 through 15. Figure 13 is the daily scrap report made out by the Inspection Department. On this form all castings scrapped for the day are recorded. The Production Department transfers the data from this report to production records. If any particular pattern number is running high scrap, an A.V.O. (Avoid Verbal Orders) form is issued to the quality control supervisor for investigation. Figures 14 and 15 are the scrap report as posted in the shop, and is illustrated to demonstrate the type of report being used.

DAILY SCRAP REPORT
DATE_____

| Customer | Pattern | No. Pcs. | Reason | #/Pc. | Total Weight | Molder | Remarks |
|---|---|---|---|---|---|---|---|
| | | | | | | | |
| | | | | | | | |
| | | | | | | | |
| | | | | | | | |
| | | | | | | | |
| | | | | | | | |
| | | | | | | | |
| | | | | | | | |

Fig. 13 — Daily scrap report form.

Figure 16 is the form used to list castings returned by customers. Four copies are made out—1) inspection copy, 2) works manager's copy, 3) sales copy and 4) accounting copy. In addition, the Inspection Department keeps a record of returns by customer and pattern number. Figure 17 is the form employed, and tells at a glance the customer rejection rate on any particular job during the year. This record has proved to be of great value in establishing quality control requirements for an individual job.

In a jobbing operation the type of production does not lend itself to statistical scrap analysis. The cost of developing the data would not be warranted when compared to the results obtained. In the author's opinion, it is the responsibility of supervision to watch the causes for internal rejection, and immediately take steps to correct the situation. The chief inspector and Inspection Department should be relied upon to catch abnormal situations and should immediately call attention to the problem.

The final form used in evaluating and controlling shop scrap and customer returns is illustrated in Fig. 18. This form is used for internal as well as external quality complaints. On internal problems the form is filled out by inspection in duplicate, the original kept for the Quality Control Committee for follow up while the copy is forwarded to the quality control supervisor for action on the problem. Each week the reports are reviewed, and corrections are reported in writing as a part of the permanent records.

### *Job History File*

The establishment of a job history file has been a great aid in quality control. A job history file starts with the blueprint of the part and contains all pertinent data on the part. If it is a new job the sample processing form (Fig. 19) and a copy of the sample

| DEPARTMENT | | Squeezer | Bench | Floor | Slinger | Pin Lift | Core | Shell | Total Weight | % |
|---|---|---|---|---|---|---|---|---|---|---|
| Weight of Scrap To Date | Pouring: Miss Run | | | | | | | | | |
| | Pouring: Slag | | | | | | | | | |
| | Pouring: Poured Short | | | | | | | | | |
| | Pouring: Shifted | | | | | | | | | |
| | Pouring: Misc. | | | | | | | | | |
| | Pouring: Total | | | | | | | | | |
| | Melting: Pin Holes | | | | | | | | | |
| | Melting: Off Analysis | | | | | | | | | |
| | Melting: Cold Heat | | | | | | | | | |
| | Melting: Wet Ladles | | | | | | | | | |
| | Melting: Total | | | | | | | | | |
| | Core Room | | | | | | | | | |
| | Cleaning Room | | | | | | | | | |
| | Inspection | | | | | | | | | |
| | Supervision | | | | | | | | | |
| Total Scrap To Date | | | | | | | | | | |
| Total Production To Date | | | | | | | | | | |
| % Scrap To Date | | | | | | | | | | |

Fig. 14 — Production scrap report form.

| MOLDING STATIONS | | PRODUCTION TO DATE POUNDS | SCRAP TO DATE POUNDS | % SCRAP |
|---|---|---|---|---|
| SQUEEZER MOLDING | 6. Pomeroy | | | |
| | 8. Serano | | | |
| | 11. Worak | | | |
| | 12. Blakesmith | | | |
| | 13. Manske | | | |
| | 17. Dziedzic | | | |
| | 21. Gondek | | | |
| | 23. Czajkowski | | | |
| | 25. Wischowski | | | |
| | 30. Banach | | | |
| | 31. Pogorselski | | | |
| | 34. Wiskowski | | | |
| | 35. Cichosz | | | |
| | 42. Banaszynski | | | |
| | TOTAL | | | |
| BENCH | 3. Ilovshek | | | |
| | 32. Ledesma | | | |
| | TOTAL | | | |
| FLOOR MOLDING | 5. Baker | | | |
| | 9. Schultz | | | |
| | 15. Ferry | | | |
| | 16. Jaster | | | |
| | 22. Schultz | | | |
| | 48. Wagner | | | |
| | | | | |
| | TOTAL | | | |
| MACHINE MOLDING | Day Shift | | | |
| | Night Shift | | | |
| | TOTAL | | | |
| | Day Shift | | | |
| | Night Shift | | | |
| | TOTAL | | | |
| TOTAL MACHINE SCRAP | | | | |
| Dry Sand Molding | | | | |
| Shell Molding | | | | |

Fig. 15 — Production scrap report form.

**RETURNED CASTING REPORT**

Customer _____ Date _____

Pattern No. _____ No. Pieces _____

Selling Price _____ Shipping Weight _____

Reason For Rejection _____

Disposition _____

By _____

Fig. 16 — Returned casting report form.

shipping letter (Fig. 20) will be filed with the blueprint. Any excessive shop scrap as reported by inspection, or any returned castings data recorded on the forms illustrated in Figs. 16 and 18, will also be filed in the job history folder. The records take several years to develop before they can be of any real value. A point that should be made here concerning the blueprint and sample forms is that the level of quality control required for any new job is established at this point. Much too often this factor is overlooked until production has been started on the job.

One of the greatest problems in maintaining quality control procedures is to establish a method to guarantee that a solution to a quality problem will be carried out on all subsequent orders for the part

REJECTION REPORT

CUSTOMER _____ PATTERN NUMBER _____

| Date Received | No. Pcs. | Reason for Rejection | No. Scrpd | No. Salv. | No. Pcs. | Remarks |
|---|---|---|---|---|---|---|
| | | | | | | |
| | | | | | | |
| | | | | | | |
| | | | | | | |
| | | | | | | |
| | | | | | | |
| | | | | | | |
| | | | | | | |

Fig. 17 — Customer rejection report form.

CUSTOMER: _____ DATE: _____

PATTERN NO. _____ CUSTOMER REPRESENTATIVE: _____

QUANTITY: _____

QUALITY COMPLAINT: _____

DECISION REACHED WITH CUSTOMER: _____

BY: _____

CORRECTION & REVIEW BY INSPECTION DEPARTMENT: _____

BY: _____

DATE: _____

SKETCH: _____

Fig. 18 — Quality complaint form.

CUSTOMER _____

SAMPLE FORM OR NOTE BY _____

Old New Revised

Pattern No. _____ Loose ☐ Loose ☐ Loose ☐

Revised ☐ Permanent ☐ Master ☐

Amount of Pcs. to One Mold _____

Pattern Received _____

Month_____Day_____Year_____

Date Patt. Rigged & Ready for Cstg. _____

Month_____Day_____Year_____

Date Cast _____

Month_____Day_____Year_____

Started to Process _____

Month_____Day_____Year_____

Amount of Pcs. Required to Submit as Samples:

Customer Check _____ Our Check _____

Appearance of Risers _____

Gates _____ Surface _____

Molding Supv. Informed ..... YES ☐ NO ☐

Gamma Ray Before Welding ..... YES ☐ NO ☐

Gamma Ray Report ..... GOOD ☐ BAD ☐

Burn Out Anneal ..... YES ☐ NO ☐

Reweld ..... YES ☐ NO ☐

Prints for Checking ..... YES ☐ NO ☐

Dimensional Check ..... YES ☐ NO ☐

New Samples Reqd.: YES ☐ NO ☐

Weld & Anneal After: YES ☐ NO ☐

Anneal: YES ☐ NO ☐

Ship ☐ Hold ☐ By Whom: _____

Sample Forms Contain Needed Information for Customer:

YES ☐ NO ☐

Heat Treatment Reqd. ..... YES ☐ NO ☐

BRN Reqd. ..... YES ☐ NO ☐

Actual BRN _____

Customer Analysis: C ☐ M ☐ S ☐ Moly ☐

DATE TO SHIPPING DEPARTMENT _____

DATE SHIPPED _____

Fig. 19 — Sample processing form for new job.

_____

_____

_____

Attention: Mr. _____

Subject: Sample Casting No. _____

Gentlemen:

We have included in our shipment of _____ sample castings, plainly marked,

Attention: Mr. _____

We would appreciate your comments regarding this / these casting / castings at your earliest convenience.

Very truly yours,

Chief Inspector

CC: _____

LIST OF SAMPLE CASTINGS

Fig. 20 — Sample shipping letter files with blueprint.

involved. To accomplish this a special card was designed, as illustrated in Fig. 21. This card is filled out and filed with the Sales Department records for the job. When a new order is placed on the part, the sales clerk pulls the card and forwards it to the person indicated. This procedure of keying quality corrections to incoming orders has accomplished the job desired with a minimum amount of clerical effort. It is the author's opinion that this procedure of keying only critical jobs eliminates the need for a job order system.

The final area of records involves the heading and gating of pattern equipment. In a jobbing operation, both production and loose pattern equipment are employed. One of the problems of quality control is to maintain a record of any changes in heading and gating. To accomplish this the form in Fig. 22 was developed, for the purpose of recording original risering, and any subsequent changes to the

CUSTOMER _____ PATTERN NO. _____

MEMORANDUM FOR: _____

NOTE: This card is to be given to the above immediately upon the receipt of a new order on this pattern.

NOTES

_____

_____

_____

_____

_____

_____

_____

_____

Fig. 21 — Card designed to carryover any quality problem solution to subsequent orders.

CHANGE IN RIGGING FORM

Customer _____ Pattern Number _____

Molding Unit: _____ Shipping Weight: _____

No. per Mold: _____ As-Cast Weight: _____

Size of Flask: _____ As-Cast Yield: _____

Cope Ht: _____

Drag Ht: _____

Remarks: _____

Sketch on other side

Fig. 22 — Rigging change form.

risering. This eliminates the requirement of relying on someone's memory as to what has been previously tried when a new problem develops on a job. It is also extremely valuable in loose work, since the sketch of what was successful on previous orders will indicate the proper risering and gating to the supervisor.

One of the important features of the system employed by the author's company has been regular quality control visits to customer plants by operating personnel. This procedure has proved to be extremely helpful in the reduction of quality problems. Only those plants where the foundry personnel can see the customer's inspection and shop personnel are visited. In this manner it is possible to determine any minor problems that are irritating to the customer. When the awareness to the problem exists, action is taken immediately to eliminate it. To control this a report form (Fig. 23) was developed. The man making the visit is required to fill out the form, and when the problems indicated are eliminated the form is filed alphabetically by customer. This establishes a record of the visits made and a record of the problems encountered. Incidentally, the customers have generally expressed approval of this type of call by operating personnel.

CUSTOMER _____ DATE OF VISIT _____

CUSTOMER PERSONNEL CONTACTED:

_____ _____ _____

_____ _____ _____

_____ _____ _____

_____ _____ _____

REPORT OF CALL:

BY: _____

Fig. 23 — Quality control customer visit form.

## NEW EMPHASIS FOR VARIABLES CONTROL

Whenever a casting is produced, the foundry producing the casting is striving to accomplish one thing —reproducibility. The entire effort is being expended to have every casting produced from any given pattern exactly like every other casting produced from that pattern. Thus far this has not been accomplished while maintaining a reasonable cost of production at the same time. The foundry industry has an exceptionally high human element in its production. This factor presents a difficult problem to scientific and quality control efforts. Once a satisfactory quality level has been achieved on any job, how can reproducibility at a competitive cost basis be achieved?

The author's company does not purport to know the answer to this problem. However, there are some steps a steel foundry can take that will aid in approaching a solution to reproducibility. One of the first steps should be to develop adequate laboratory facilities to control the production of metals and sands. Figures 24 through 28 are views of recently installed laboratory facilities at the author's company. One of the reasons the industry controls the production of metal and sand as well as it does is due to the means in the companies for checking the results of production.

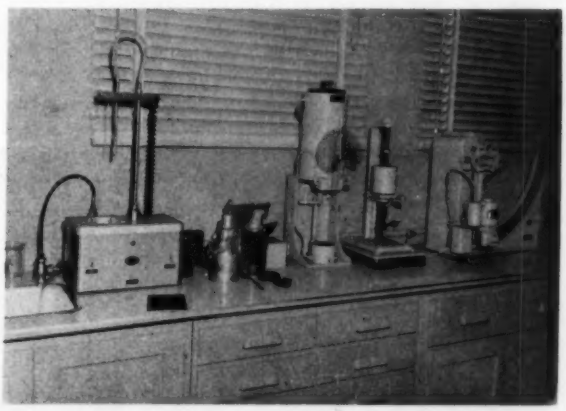
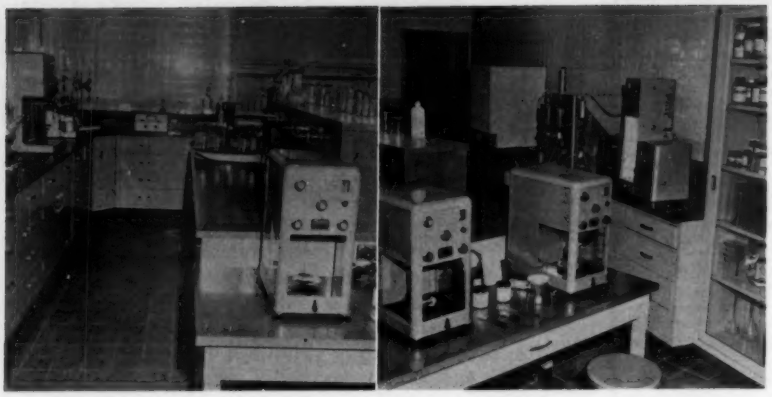


Fig. 24 — View of sand laboratory.

Chemical and physical properties of metal, as well as sand properties, are not subjected to opinions, the results are determined by laboratory inspections. Can today's steel foundry continue with only the barest rudiments for these facilities? The author thinks not, not if integrity as producers of an engineering material is to be established.

How many steel foundries have as an integral



Figs. 25 and 26 — Views of chemistry laboratory.

part of their facilities nondestructive testing? Facilities such as x-ray, cobalt and magnetic particle inspection equipment. Recently the author heard a quality control manager of a large user of castings cite an annual rate of rejection of $847,000 on casting purchases of 18 to 20 million dollars with a machining loss of $147,000. The causes for rejection were:

61 per cent.....Internal defects
25 per cent.....Visual defects
12 per cent.....Dimensional variations
2 per cent.....Miscellaneous defects

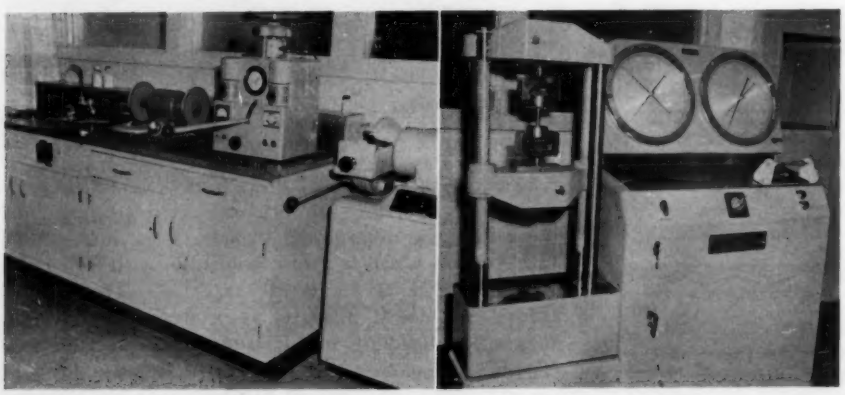
This indicates the industry must do much more in eliminating the problem of internal defects; nondestructive facilities is a step in that direction. At the author's foundry Cobalt 60 and magnetic particle testing are part of operations (Figs. 29 and 30). In the example cited, if one-half the rejections for internal defects could be eliminated, the dollars saved for the foundries involved would exceed a quarter of a million dollars. This type of savings will pay for a lot of nondestructive testing.

Dimensional control is becoming a critical factor in final inspection for a steel foundry. At the author's company the use of straightening processes, gages and checking fixtures is increasing at a rapid rate. These tools are a necessity if proper quality is to be maintained. Figure 31 is a recently acquired 500 ton press for straightening. Figures 32 and 33 are illustrations of the employment of gages and fixtures in final inspection. The production of castings to conform to inspections requiring the use of gages and fixtures is a function of quality control. Proper steps taken in core making and molding will minimize the problems in the cleaning room.

Can the money spent on quality control be afforded? Perhaps the best answer is, can we afford not to spend it? Quality control should save money if it is properly conceived and executed. In the author's opinion, the savings in quality control can be derived from two things, a reduction in internal scrap and a reduction in customer returns and allowances. Values can be placed on each of these items, and direct savings calculated from the reduction obtained. Dollar values are not quoted here, since they are a function of an individual foundry's operating costs and selling price.

## CONCLUSION

Results of the program the author has described have been encouraging. However, what has been done is not enough. After several years of honest effort to improve quality it can be noted that the author's company is just beginning to catch up to customer demands. It is not sufficient to just catch up, but must move ahead of customer requirements in quality. The effort to move ahead cannot be accomplished in a short period of time. Installation of systems or people will not suddenly make a foundry a



Figs. 27 and 28 — Views of physical laboratory.




Fig. 29 — Nondestructive testing.



Fig. 30 — Magnetic particle testing.

high quality shop. Education or re-education through a gradual step-by-step program will accomplish an improvement in quality; however, a vital factor is an aggressive dynamic management that is not content to be second best.

The steel foundry industry needs quality control, even though it currently has quality control to varying degrees. What the industry needs to do is establish programs that put the responsibility where it belongs. Programs are needed that will increase the interest and responsibilities of hourly people, establish supervision as the key to good quality control, use a minimum of record keeping and clerical work and, finally, to establsh a mutually agreeable quality level with customers. This must be the goal of quality control if the industry is to maintain a competitive position.

## REFERENCES

1. E. G. Tetzlaff, "Melt Department Incentive Plan," AFA Transactions, *vol. 60*, p. 432 (1952).
2. H. H. Johnson & G. A. Fisher, "Statistical Quality Control a New Tool for the Foundryman," AFA Transactions, *vol. 56* (1948).

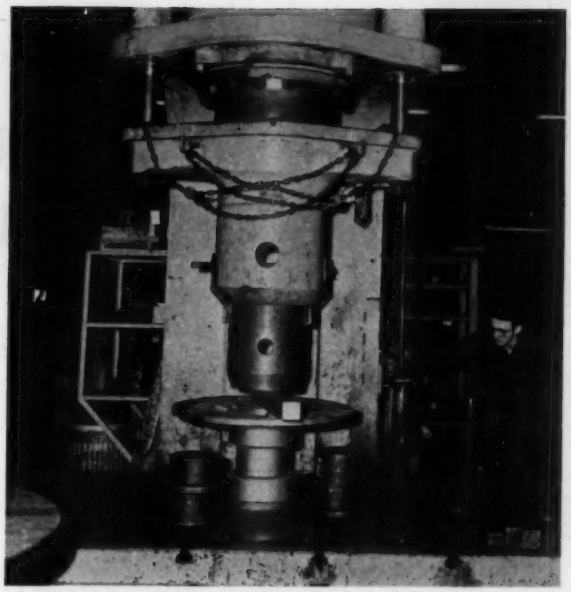


Fig. 31 — 500 ton straightening press.

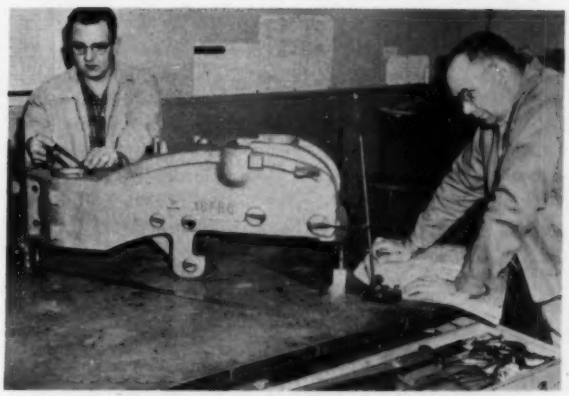


Fig. 32 — Typical layout inspection.



Fig. 33 — Final template inspection.

# MOLD VARIABLES INFLUENCE ON FLUIDITY OF ALUMINUM

## *Doubling casting ductility by fluidity control*

by M. C. Flemings, F. R. Mollard and H. F. Taylor


### ABSTRACT

**Results are presented of a study of the influence of various molding materials on the fluidity of aluminum-4.5 per cent copper alloy. Molding materials studied include green sand with various water contents and additives, core sands (bonded with linseed oil and phenol-formaldehyde), $CO_2$ sand and zircon sand. Effects of variables such as sand grain size were also studied. Results indicate significant influences of the molding materials on fluidity. For example, fluidity is 20 to 27 per cent less in core sands than in green sand, and 22 to 55 per cent less in zircon sand than in green sand. Data obtained are analyzed in detail.**

**Fluidity improvement due to mold coatings such as hexachloroethane is shown to result from reduction in the mold-metal heat transfer coefficient. These mold coatings are effective in improving fluidity of a wide range of metals and alloys in sand molds (not just aluminum alloys). Use of mold coatings can result in significantly improved properties in aluminum castings, by permitting substantially reduced pouring temperatures. For example, in an aluminum-4.5 per cent copper alloy test casting, coatings permitted lowering the pouring temperature 100 F. As a result, elongation doubled (increased from 4 per cent to 8 per cent) with a simultaneous increase in tensile strength.**


### INTRODUCTION

Work conducted during the third year of a research program on fluidity of aluminum alloys conducted at Massachusetts Institute of Technology, sponsored by the U.S. Army Ordnance Department through Pitman-Dunn Laboratory, Frankford Arsenal is described. Publications and reports issued earlier have described work conducted to date on fundamental and applied aspects of the program.[1,2]

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Much of the work conducted has been on basic investigations of heat and fluid flow, and of solidification in fluidity test channels. In addition, an applied phase of the study was the investigation of fluidity in sand molds, and of the improvements in fluidity that could be obtained through use of mold coatings.[2,4] Work reported herein is a continuation of that portion of the study concerned with fluidity in sand molds. It is concerned primarily with (1) influence of various molding materials on fluidity of aluminum alloys and (2) influence of mold coatings on mechanical properties of thin aluminum castings.

### MOLD VARIABLES INFLUENCE ON ALUMINUM-4.5 PER CENT COPPER ALLOY FLUIDITY

#### *Test Apparatus and General Procedure*

This study was conducted using the double spiral sand mold test developed and described in detail earlier.[2-4] Figures 1 and 2 illustrate the test casting and mold design. Essential features of the test are:

1) An overflow pouring basin to maintain constant metal head (pressure) during pouring.
2) A tapered sprue and sprue well to minimize turbulence and dross formation.
3) A large runner and bottom gated spiral, to insure that no metal enters the test section until full pressure (metal head) is developed in the pouring basin.
4) A screen to filter entrapped dross.
5) Two test spirals.

The two spirals in a single mold permit study of different mold (and mold coating) variables. For example, one spiral can be rammed in green sand and the other in $CO_2$ sand to determine relative



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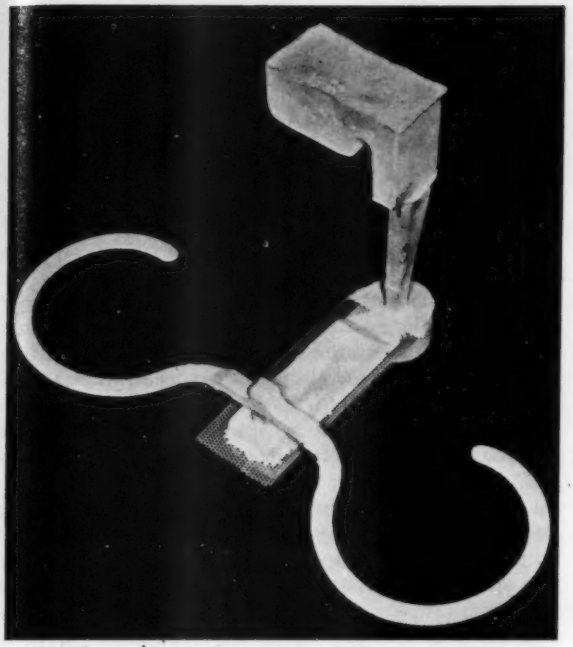


Fig. 1 — Sand mold fluidity test casting with gating system.

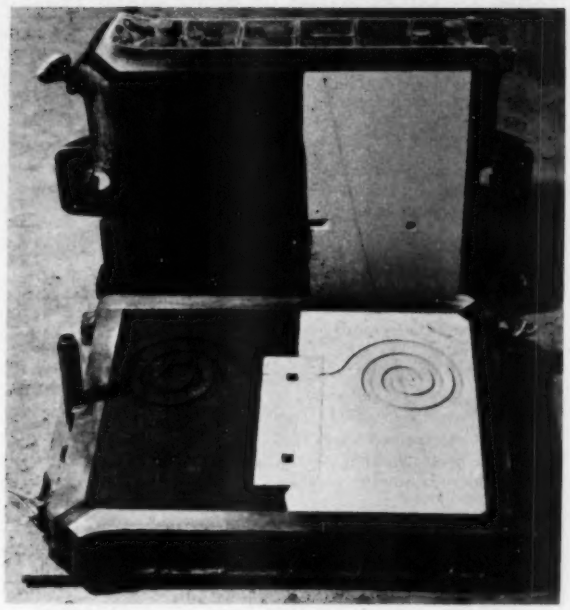


Fig. 3 — Cope and drag of sand mold fluidity test casting, showing technique used to compare fluidity in green sand (left) with that in $CO_2$ sand (right).

fluidities in these two molding media. Figure 3 illustrates such an arrangement.

The general procedure used for testing mold variables was to produce six test molds for each trial run. One spiral in each test mold was faced (covered to a depth of ½-in.) with the molding media to be studied, and the other spiral was faced in a standard foundry sand as a control piece (110 AFS, clay bonded green sand, Mix A, Table 1). The standard foundry sand was then used as backing sand for the entire mold. Fluidity tests are of such short duration (metal flows in the spiral for only about one sec) that ½-in. of sand is more than adequate as an "infinite" mold. For some tests, minor modifications of the above molding procedure were employed and these are noted where used.

Each set of fluidity tests was conducted as described earlier,[2,4] except that in this work grain refined metal was used to simulate commercial practice. In previous cases the metal was not grain refined. A gas fired furnace was used to melt metal of nominal composition 4.5 per cent copper, 0.14 per cent titanium, balance aluminum. Charge materials were aluminum ingot (99.9 per cent), aluminum-copper master alloy (50 per cent Cu), aluminum-titanium master alloy (6 per cent Ti) and revert (pigged

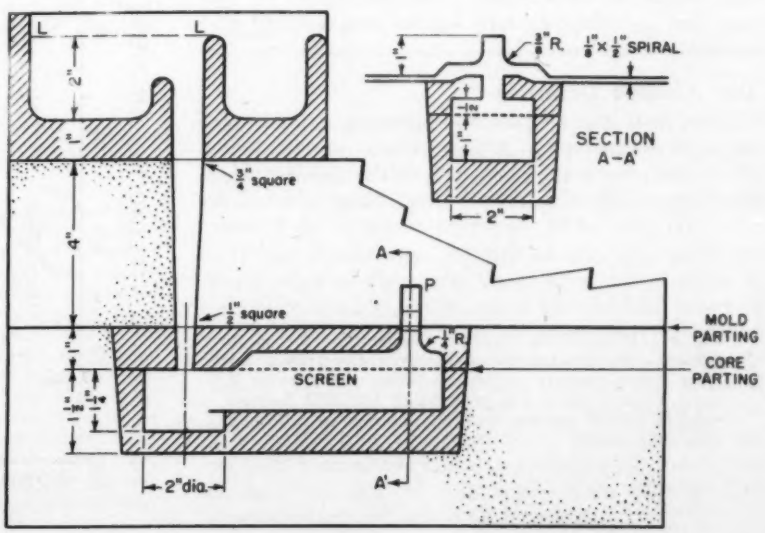



Fig. 2 — Schematic diagram of gating system for sand mold fluidity test.

TABLE 1 — COMPOSITION OF SAND MIXES TESTED

| Composition | Pounds |
|---|---|
| Mix A (Standard green sand) | |
| Silica sand (110 AFS) | 100 |
| Southern bentonite | 5.0 |
| Cereal | 0.75 |
| Water | 3.5 |
| Mix B (Zircon green sand) | |
| Zircon sand (110 AFS) | 100 |
| Southern bentonite | 2.5 |
| Cereal | 0.40 |
| Water | 1.75 |
| NOTE: On a volumetric basis, this sand has approximately the same quantity of additives as Mix A. | |
| Mix C (Fine green sand) | |
| Silica sand (140 AFS) | 100 |
| Southern bentonite | 5.0 |
| Cereal | 0.75 |
| Water | 3.5 |
| Mix D (Coarse green sand) | |
| Silica sand (30 AFS) | 100 |
| Southern bentonite | 5.0 |
| Cereal | 0.75 |
| Water | 3.5 |
| Mix E | |
| Silica sand (110 AFS) | 100 |
| Proprietary sodium silicate binder | 3.0 |
| Mix F | |
| Silica sand (110 AFS) | 75 |
| Silica sand (80 AFS) | 25 |
| Proprietary sodium silicate binder | 3.0 |
| Mix G | |
| Silica sand (110 AFS) | 100 |
| Linseed oil | 2.8 |
| Cereal | 1.5 |
| Water | 5.0 |
| Baked at 400 F, 3 - 4 hr | |
| Mix H | |
| Silica sand (110 AFS) | 100 |
| Proprietary phenol-formaldehyde resin binder | 1.0 |
| Cereal | 1.5 |
| Water | 4.0 |
| Baked at 400 F, 3 - 4 hr | |

aluminum from gates and risers of previous test castings which were analyzed before use) in quantities up to 75 per cent of the total charge. Chemical analyses were made on most heats which proved to be 4.65 per cent ± 0.20 per cent copper and 0.13 to 0.14 per cent titanium.

### *Dry Nitrogen Degassing*

Each melt was degassed by bubbling dry nitrogen through the metal for at least 10 min at 1300-1350 F (704-732 C) while the metal was in the furnace. Melts were then tapped at 1400 F (760 C) and checked for gas with the reduced pressure tester. If any gas was found to be present, the degassing operation was repeated. Samples were taken from each heat for chemical analysis. After tapping, all molds were poured consecutively from the same crucible. Care was taken to insure accurate pouring temperatures.

Since pouring temperature is of great importance in determining metal fluidity, it is necessary to measure temperatures precisely. In these experiments, a chromel-alumel thermocouple, enclosed in an iron protection tube, was immersed in the crucible at the time of pouring. Melt contamination was prevented by coating the iron tube with silica-bentonite wash. The temperature of the melt at the instant of pouring was indicated on an electronic recorder with an external cold junction (ice water).

After shakeout, the test spirals were measured by comparison with a reference spiral graduated every 0.2 in. Readings were made at the extreme tip of the spiral and, although the spiral tip was slightly rounded, the readings may be considered accurate to ± 0.1 in. The shape of the tip was uniform in all the spirals, and there were no irregular surface protrusions.

Thickness of each spiral was measured carefully to determine if there was any accidental variation of thickness, but accuracy was maintained at ± 0.005 in. A few spirals exceeded this tolerance and were discarded. Thickness measurements were taken every 2 in. along the axis of the spirals, and a trend was noted for them to be slightly undersize at mid-length and oversize near the tip. The extreme tip was again undersize and smooth, due to surface tension and film effect which prevented complete filling of the mold. Similar results have been reported previously.[2] Aver-

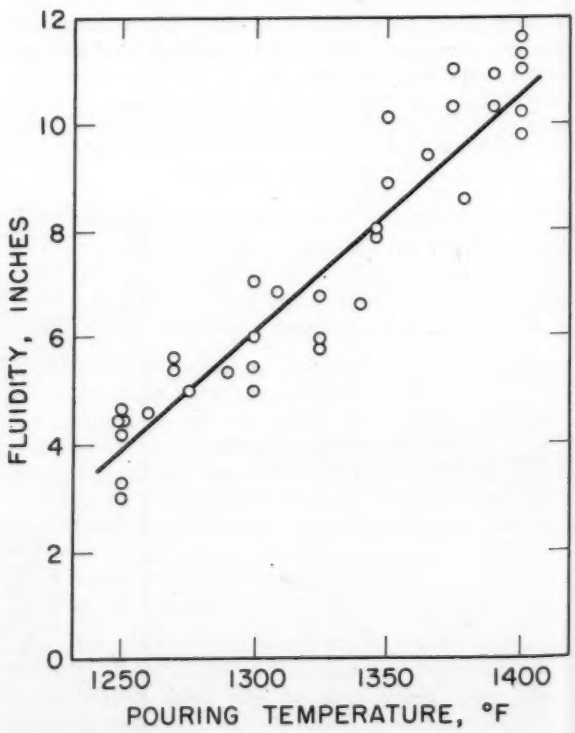



**Fig. 4 — Fluidity vs. pouring temperature for grain refined aluminum-4.5 per cent copper alloy. Data are from control spirals of present work, representing 8 heats. Control spiral was green sand, 110 AFS fineness number. Grain refining was by addition of 0.15 per cent titanium.**

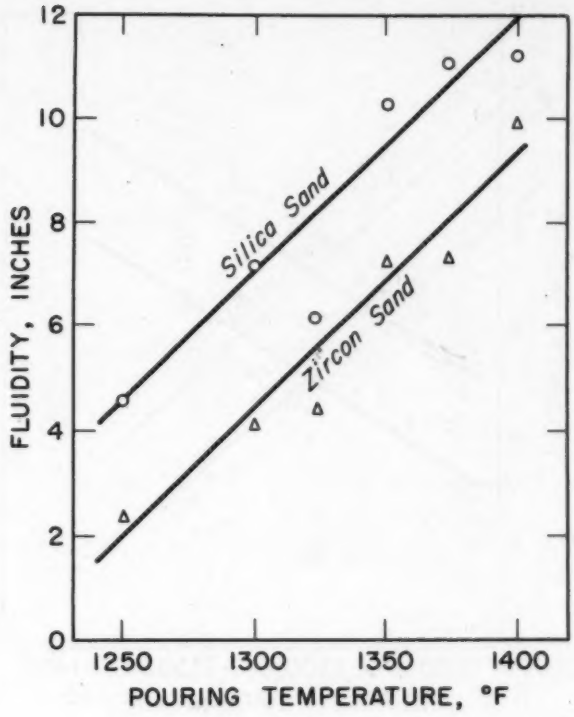



**Fig. 5 — Fluidity of aluminum-4.5 per cent copper alloy in silica sand and zircon sand. Both sands were bentonite bonded, green and of approximately 110 AFS fineness number.**

age thickness of all test castings produced was 0.123 in. ± 0.005 in. (except where otherwise noted).

Figure 4 summarizes data obtained from the control spiral of eight different heats. The data fall generally within ± one in. of a "standard" curve, with only 4 points falling outside this band. The standard curve for this study falls slightly below that reported earlier,[2,4] probably because grain refined metal was used for the studies described but not for those reported previously. It is to be expected that grain refinement would reduce fluidity in this test, as it does in the vacuum fluidity test.[3]

### *Fluidity in a Chilling Sand*

In order to determine if molds made of a "chilling" sand (sand with a high thermal diffusivity) would reduce fluidity significantly, tests were made using zircon and regular silica sand in undried (green) sand molds. The zircon sand was similar in grain distribution to the 110 AFS standard silica sand, Table 2. Since the density of zircon sand is approximately twice that of washed silica sand, the amount of additives (clay, cereal, water) mixed with zircon sand was only half as much by weight as the amount mixed with the standard sand. Thus, both sands had approximately equal volumetric compositions. Mix B, Table 1, is the mixture of zircon sand used.

Figure 5 is a comparison of the effects of zircon and silica sand molds on fluidity. Zircon sand lowers fluidity significantly (approximately 2.5 in. at all temperatures). Spirals cast in the zircon sand showed no apparent difference from those cast in silica sand (as regards surface finish, spiral thickness, etc.), and it appears certain that the lower fluidity in zircon sand is directly attributed to the greater chilling power* of this material.

### *Sand Grain Size Effect on Fluidity*

The influence of sand grain size on fluidity was determined using two different sands in each test mold.

**TABLE 2 — GRAIN SIZE DISTRIBUTION OF SANDS TESTED**

| | Weight Retained | | | | |
|---|---|---|---|---|---|
| U.S. Sieve Number | Silica sand (110 AFS) | Silica sand (110 AFS + 80 AFS) | Zircon sand (110 AFS) | Fine silica sand (140 AFS) | Coarse silica sand (30 AFS) |
| +20 | | | | | 2 |
| +30 | | | | | 57 |
| +40 | | | | | 40 |
| +50 | | 1 | | | |
| +70 | 5 | 17 | | 1 | |
| +100 | 29 | 40 | 11 | 13 | |
| +140 | 34 | 23 | 66 | 29 | |
| +200 | 18 | 11 | 19 | 31 | |
| +270 | 6 | 3 | 1 | 17 | |
| Pan | 5 | 3 | — | 8 | |
| Total (Grams) | 97 | 98 | 97 | 99 | 99 |

To emphasize possible effects, the standard 110 AFS sand was not employed, and in its place a still finer sand (Mix C, Table 1) was used on one side of the mold. The spiral on the other side of the mold was molded in a coarse sand (30 AFS, Mix D).

It was found that fluidity in the fine sand was lower than in the coarse sand. This was contrary to preliminary expectations, since the chilling power of coarse sand was thought to be somewhat greater than that of fine sand.[5,6] The explanation for results obtained (Fig. 6) lies in the fact that significant metal penetration was obtained in the coarse sand spirals. The penetration amounted to as much as 0.025 in. on each face of the spiral (approximately one grain diameter for No. 30 AFS sand). This amounted to approximately 10 per cent by weight.

*"Chilling power" of molding sands can be expressed quantitatively as the square root of the product of thermal conductivity, density and specific heat of the bulk sand. Chilling power of zircon sand is approximately 50 per cent greater than that of silica sand. Under conditions ordinarily encountered in solidification of simple sand castings (negligible resistances to heat flow in the metal and at metal-mold interface), the above means that in a given time zircon sand will extract 50 per cent more heat from a solidifying casting than will silica sand. It means, further, that a simple casting will solidify in zircon sand in less than half the time required in silica sand.

In the case of a fluidity test casting, solidification time is so short that heat flow resistance at the metal-mold interface cannot be neglected.[7] Here, zircon sand will not exert the full influence calculated, but will nonetheless extract heat more rapidly than silica sand.

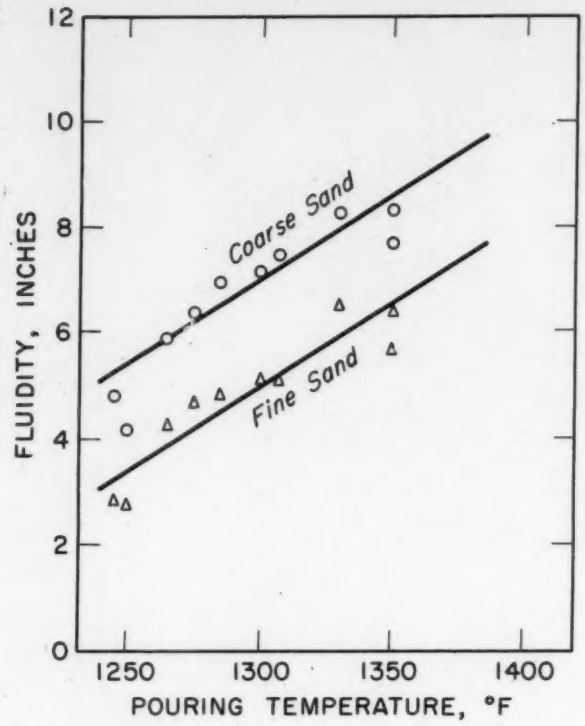



Fig. 6 — Fluidity of aluminum-4.5 per cent copper alloy in coarse (30 AFS) and fine (140 AFS) silica sand, clay bonded, green. Data from two heats.

The added heat of fusion that must be extracted to solidify an extra 10 per cent of metal can readily account for the increase in fluidity observed. As concerns the effect of grain size on thermal diffusivity, this effect would be expected to be small particularly at temperatures of interest in aluminum founding.[8] Even at steel temperatures, differences in grain size per se have not been found to affect fluidity.[9]

While the higher permeability of coarse sand is sometimes considered to promote fluidity, this does not appear to be a factor in these tests. Vents (½-in. diameter) were cut at the extreme ends of all spirals in this study to avoid any danger of back pressure. Also, in order to check if back pressure was significant, a series of tests were conducted with and without vents (standard 110 AFS green sand). No effect of venting on fluidity was found.

### *Moisture Content*

To determine the effect of moisture content on fluidity in clay bonded sand, a series of molds was poured in one heat in which the first and last molds poured had been made of standard green sand (containing 3.5 per cent water). Intermediate molds were of the same sand but had been baked 3 - 4 hr at 400 F (204 C) and allowed to cool. Baked molds gave the same results as green sand molds (Fig. 7), and it may be assumed that fluidity in clay bonded sands is not influenced by moisture content at least in amounts up to 3.5 per cent.

### *New Versus Used Sand; Effects of Additives*

It was felt that carbonaceous residues and/or other changes in molding sands as a result of repeated use might have some effect on fluidity. Hence, tests were conducted with the standard 110 AFS green sand in the new and re-used condition, with moisture content of both sands held at 3.5 per cent. Results of fluidity tests showed no measurable effects.

Fluidity was next compared in two green sand mixes which were standard, except that one mix contained 0.75 per cent cereal. Cereal proved to have no effect on fluidity.

To determine if larger amounts of organic material might affect fluidity, 10 per cent sawdust was added to the molding sand. Even with this large amount of

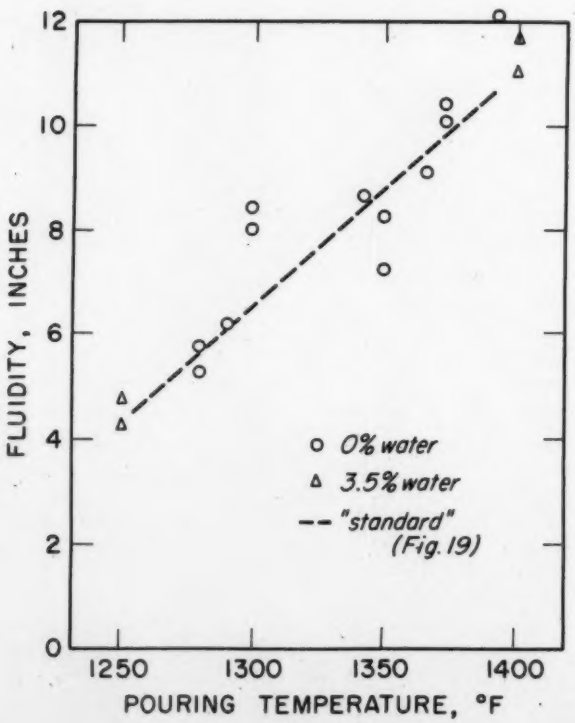



Fig. 7 — Fluidity of aluminum-4.5 per cent copper alloy in clay bonded silica sand, showing fluidity is the sand in dried molds (0 per cent water) as in green molds (3.5 per cent water). Sands were 110 AFS fineness number.

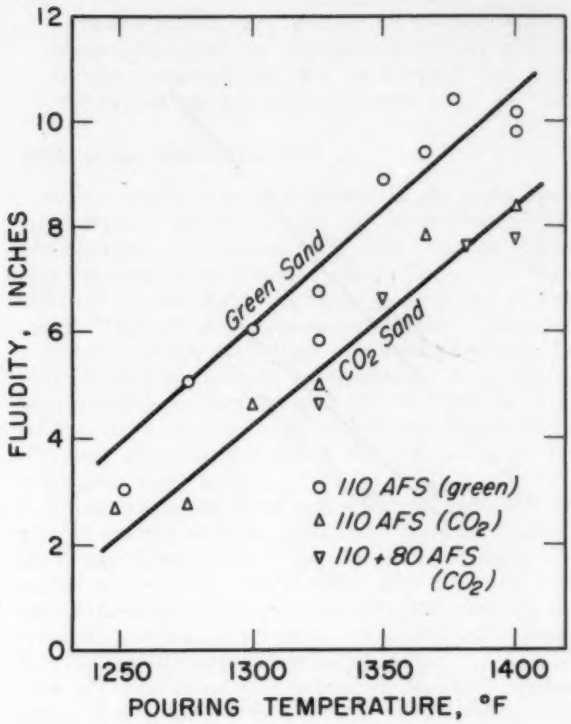



Fig. 8 — Fluidity of aluminum-4.5 per cent copper alloy in sodium silicate bonded ($CO_2$) sand and in green clay bonded sand. Data from two heats. Green sand was 110 AFS fineness number.

additive, only relatively small improvement (10 to 13 per cent) was detected.

*Binder Effect on Fluidity*

**Sodium Silicate.** Two heats were made to compare fluidity in sand molds bonded with sodium silicate to that of regular clay bonded green sand. The standard 110 AFS green sand (Mix A) was used as control, and two sodium silicate bonded mixes were tested:

1) A mix based on 100 per cent 110 AFS sand (Mix E).
2) A mix based on 75 per cent 110 AFS sand and 25 per cent 80 AFS (Mix F).

Results of these tests (Fig. 8) show that fluidity in the sodium silicate (so-called $CO_2$) bonded sands is somewhat less than in green sand. Fluidity was the same in the two sodium silicate bonded mixes, indicating again that sand grain size had no effect on fluidity. In this test, spiral surface finish, weight, etc. were the same for castings produced in different sands, and so metal penetration did not influence results as it had in the tests summarized in Fig. 6.

**Linseed Oil.** Results of tests to determine fluidity in linseed oil bonded (dry sand) molds are summarized in Fig. 9. Molds poured at the highest and lowest temperatures were control molds, made with standard 110 AFS green sand. Molds poured at intermediate temperatures were composite molds, consisting of linseed oil bonded sand facing (Mix G), on one spiral, and standard 110 AFS sand on the other spiral. These molds were baked at 400 F (204 C) for 3-4 hr and cooled to room temperature before pouring. Fluidity in the linseed oil bonded sand was lower than in the green sand.

**Phenol Formaldehyde.** Fluidity in molds bonded with phenol formaldehyde, like that in other core sands, was lower than in green sand, Fig. 10. In this series of tests, the molds poured at the highest and lowest pouring temperatures were of the standard 110 AFS green sand (Mix A). Molds at intermediate temperatures were faced with phenol formaldehyde bonded sand (Mix H). These were baked at 400 F (204 C) for 3-4 hr and cooled to room temperature before pouring.

**Binders Effect on Fluidity Discussion.** There appears to be no simple explanation why fluidity in

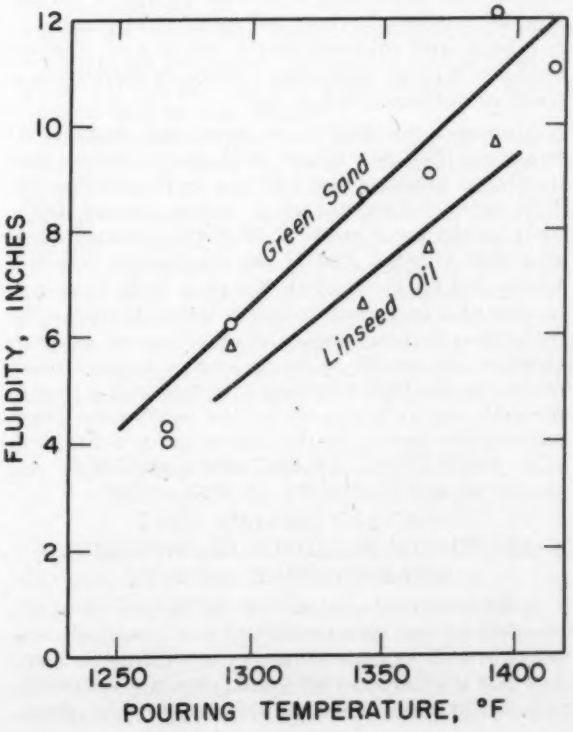



Fig. 9 — Fluidity of aluminum-4.5 per cent copper alloy in clay bonded green sand. Both sands were 110 AFS fineness number.

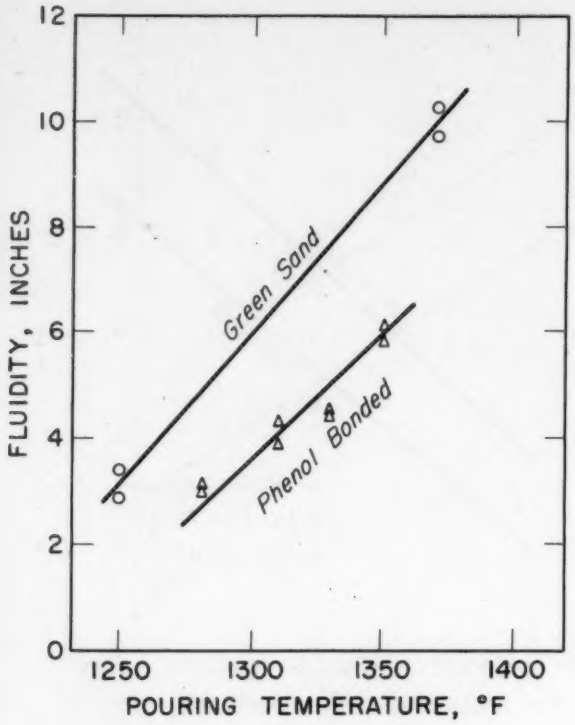



**Fig. 10 — Fluidity of aluminum-4.5 per cent copper alloy in phenol-formaldehyde bonded sand and clay bonded green sand. Both sands were 110 AFS fineness number.**

various core sands (including those bonded with sodium silicate, linseed oil and phenol formaldehyde) is less than fluidity in green clay bonded sand. Differences in moisture content alone cannot account for the results obtained, since it was shown that moisture content has no effect on fluidity in clay bonded molds. Factors such as changes in surface tension or surface films are unlikely to have influenced results, and surface finish and thickness of the spirals were checked carefully with no significant differences detected as a result of different binders used.

It appears that fluidity in green sand molds must be greater than that in core sand molds because heat transfer is somehow less effective in the green sand. This could be because of (1) lower thermal Diffusivity in the green sand, or (2) higher contact resistance due to a gas film at the metal-mold interface in the case of green sand. There is little reason to suppose that green sand possesses lower thermal diffusivity than do core sands, and experimental evidence indicates the reverse to be true.[10, 11] A more likely reason for the higher fluidity in green sand is that in thermal contact resistance at the mold-metal interface may be greater in the case of green sand. However, there is no apparent explanation why this should be so.

## ALUMINUM FLUIDITY IN SAND MOLD QUANTITATIVE ASPECTS

A mathematical analysis for the fluidity of aluminum alloys has been developed and presented elsewhere.[3] This analysis is based on a model of metal flow and solidification behavior in fluidity spirals that is in accord with current understanding of the process, and can be applied with relative ease to solution of a wide range of problems of practical interest. Some results of this analysis are considered herein, as they apply to, and aid in, understanding experimental observations on fluidity in sand molds.

Specifically, the analysis referred to above applies to fluidity tests on mushy freezing alloys, such as 195 alloy, in which:

1) Solid particles form during flow in a fluidity spiral and travel downstream with the liquid.
2) Flow stops when the mean solid concentration reaches a certain value (critical solid concentration).
3) Flow velocity is essentially constant until flow stops.

The analysis shows, quantitatively, that these factors are important in determining fluidity:

1) Test variables, such as geometry of the flow channel and effective pressure (metal head) at the spiral entrance.
2) Mold variables including mold temperature and thermal diffusivity of the mold.
3) Metal-mold variables, specifically, thermal contact resistance between metal and mold.
4) Metal variables including superheat, amount of solid required to stop flow (critical solid concentration), and heat of fusion of the solidifying metal.

For fluidity tests of the kind usually employed (and for most sand castings) factors such as surface tension, surface films or viscosity may be neglected.

In addition to delineating many factors that are known by intuition or experiment to affect fluidity in sand molds, the analysis shows that two less well understood factors can have an important effect. These are (1) changes in critical solid concentration and (2) changes in mold-metal (thermal) contact resistance.

In most fluidity spirals (and in most sand castings) effective metal head is low. For example, in the sand spiral used in this work the metal head over the spiral was 7 in.; however, due to friction losses in the runner system the "effective" metal head, in terms of pressure at the spiral entrance, was somewhat less than 2 in. Under these conditions a small amount of solidification near the tip of the flowing stream is sufficient to stop flow – about 5 per cent solid suffices. At higher pressures (not usually obtained in sand casting) substantially more solidification can occur before flow ceases.

Thermal diffusivity (chilling power) of molds is

known to affect fluidity by altering the rate of heat extraction. In addition, thermal contact resistance at the mold-metal interface must also be considered, and analysis shows it to be a significant factor. This is different from the case of solidification of relatively large castings where (because of the much longer times involved) mold-metal contact resistance can be neglected.

### *Heat Transfer Coefficient*

Analysis of data obtained in this work, and earlier,[3] indicates the coefficient of heat transfer at the mold-metal interface in sand molds is about 0.05 cal/cm² sec° C, and shows this value to be significant in determining total flow.

As reported earlier, certain mold coatings, such as hexachloroethane and carbon black increase fluidity in green sand molds by as much as a factor of three.[2, 4] On the other hand, the use of certain core sands instead of green sand result in decreases of fluidity from 20 to 40 per cent. Changes in the heat transfer coefficient that would account for such changes in fluidity were calculated, and are shown in Table 3. Critical solid concentration was assumed to be 5 per cent.

**TABLE 3 — HEAT TRANSFER COEFFICIENT AT MOLD-METAL INTERFACE EFFECT ON FLUIDITY IN SAND MOLDS**

| Heat transfer coefficient h, cal/cm² sec° C | Fluidity at 1350 F, $L_f$, in. | Change in fluidity,* % |
|---|---|---|
| 0.015 | 30 | +200 |
| 0.053 | 10 | 0 |
| 0.071 | 8 | −20 |
| 0.110 | 6 | −40 |

*Compared to fluidity at h = 0.053; this is the value obtained for uncoated green sand molds.

It is of interest to note that the heat transfer coefficient need be changed for only an extremely short time to influence fluidity. For example, in a typical test in a fluidity spiral, metal velocity (V) is 80 cm/sec and the choking zone ($\Delta X$) is one cm. The average time the tip of the metal is in contact with any one portion of the mold is $1/80 = 0.018$ sec.

To substantiate that mold coatings did improve fluidity by changing heat transfer and not by some other mechanism, a study was made of the effect of mold coatings on flow velocity in the sand mold fluidity test. The reason for studying flow velocity to determine the mechanism of fluidity improvement is that analysis[1,2] and intuition indicate:

1) If a mold coating improves fluidity by changing factors such as surface tension, oxide films, etc., this improvement (from a fluid flow standpoint) should be the same as would be achieved by an increase in pressure head. This, flow velocity (particularly initial flow velocity) would be markedly increased. Duration of flow (fluid life) would be increased only slightly, or even decreased, depending on alloy content, superheat, etc.
2) If a mold coating improves fluidity by changing the rate (or mode) of solidification, fluid flow variables in the early portion of the test will remain unchanged. In this case, the initial velocity of the stream should not be affected, but fluid life should be increased markedly.

### *Fluid Flow Determination*

Flow velocities were determined in the green sand mold fluidity spirals through use of motion picture photography (64 frames/sec). The double spiral (Fig. 3) had been designed especially so the flow channel was wholly in the drag (bottom half of the mold). This allowed substitution of glass plates for a portion of the sand cope (top half of mold). In practice, two plates, each approximately 8 in. square were molded so they covered the two spirals completely. After the cope was rammed and turned over, the sand covering the plates was cut away so the two spirals were visible.

Before closing the mold, one spiral was treated with a mold coating and the other left untreated. During the actual fluidity test, a motion picture camera was placed so it would record metal flow into the two spirals simultaneously. Occasionally the glass plates cracked, but only after flow was completed (actual flow time in all cases was one sec or less).

In all tests using the technique described above, initial flow velocity was the same in the coated and uncoated spirals. Therefore, it can be concluded that the hexachloroethane changes heat transfer and not surface phenomena such as surface oxide films, surface tension, etc. Figure 11 shows a plot of distance versus time (in the spiral test) for an aluminum-4.5 per cent copper alloy poured at 1325 F (719 C). In three other tests similar results were obtained. In all cases, initial velocity of the metal (initial slope of the curve of Fig. 11) was constant at 30-35 in./sec regardless of pouring temperature. In other words, the coated spiral flowed not faster but longer, and therefore farther. A similar test conducted with a mold coating of amorphous carbon also showed improvement in fluidity with no effect on flow velocity.

Further confirmation that mold coatings improve fluidity by changing the heat transfer coefficient is given by the fact that hexachloroethane mold coatings have resulted in improved fluidity of all metals and alloys tested to date. It appears unlikely that the coating could effect improvement in all alloys by any other mechanism than changing the heat transfer coefficient. Table 4 lists results obtained for a series of ferrous and nonferrous alloys, including two aluminum alloys, a magnesium alloy, two copper base alloys, a low alloy steel and a stainless steel.

## MOLD TREATMENT EFFECT ON MECHANICAL PROPERTIES OF THIN WALLED CASTINGS

One practical result of work conducted earlier in this program[2,4] was the development of mold coatings for improving fluidity of aluminum alloys in practice, and the establishment by various tests that mold coatings had a strong influence on fluidity. Many foundries are now using these coatings, and as a result it is of practical interest to determine what

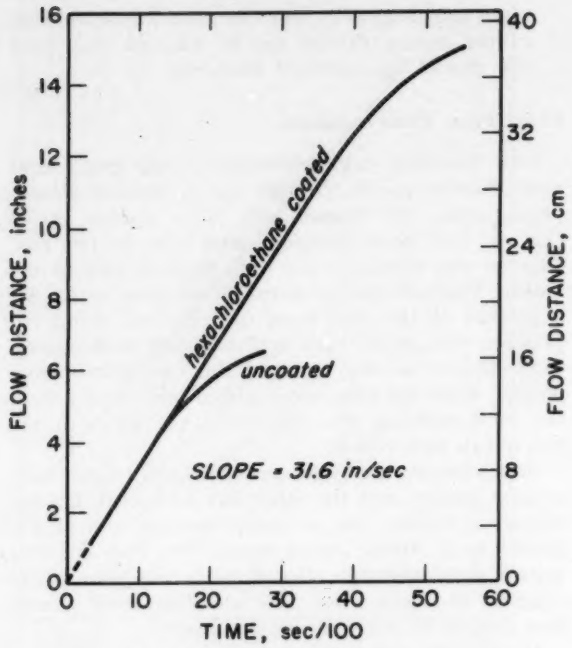



Fig. 11 — Distance vs. time for hexachloroethane-coated and uncoated spiral. Aluminum-4.5 per cent copper alloy. Pouring temperature 1325 F.

effect, if any, the coatings have on mechanical properties of sand castings.

To determine the above, a test casting was constructed with two plates mounted on opposite sides of a runner, each plate being 5 x 5 x 1/8-in. thick (Fig. 12). To determine the effect of mold coatings, the mold cavity of one plate was coated and the other left uncoated. A series of castings was then produced at different pouring temperatures using

TABLE 4 — HEXACHLOROETHANE EFFECT ON FLUIDITY OF SOME COMMERCIAL NONFERROUS AND FERROUS ALLOYS

| Alloy | Pouring temp., F | Fluidity control spiral, in. | Fluidity coated spiral, in. | Fluidity increase due to coating, in. | Increase due to coating, % |
|---|---|---|---|---|---|
| **Aluminum Alloys** | | | | | |
| 195 Al-4.5 Cu | 1318 | 6.3 | 17.9 | 11.6 | 184 |
| | 1249 | 3.8 | 11.0 | 7.2 | 190 |
| 356 Al-7 Si-0.3 Mg | 1318 | 11.9 | 22.1 | 10.2 | 86 |
| | 1246 | 7.2 | 14.1 | 6.9 | 96 |
| **Magnesium Alloy** | | | | | |
| AZ91C Mg-8.7 Al -0.72 Zn-0.15 Mn | 1300 | 12.3 | 23.5 | 11.2 | 91 |
| **Copper-Base Alloys** | | | | | |
| Manganese-Bronze 58.2 Cu-39 Zn 1.4 Fe-1 Sn-0.1 Mn | 1750 | 4.8 | 10.7 | 5.9 | 122 |
| 85-5-5-5 85 Cu-5 Sn -5 Pb-5 Zn | 2000 | 3.5 | 5.7-7.7* | 2.2-4.2 | 63-125* |
| **Ferrous Alloys** | | | | | |
| 4340 Steel 0.40 C-1.75 Ni 0.80 Cr-0.25 Mo | 2820 | 3.3 | 5.0-8.4* | 1.7-5.1 | 56-154* |
| CA-40 (420) Stainless Steel | 3100 | 11.0 | 12.2-13.8 | 1.2-2.8 | 11-25 |
| 0.20 C-13.4 Cr 0.40 Mn -0.30 Si | 3000 | 9.3 | 11.8-12.6* | 2.5-3.3 | 27-35 |

*The smaller figure refers to the length of spiral in which the metal completely filled the mold cavity. The larger figure gives the maximum length to which any metal flowed in the spiral.

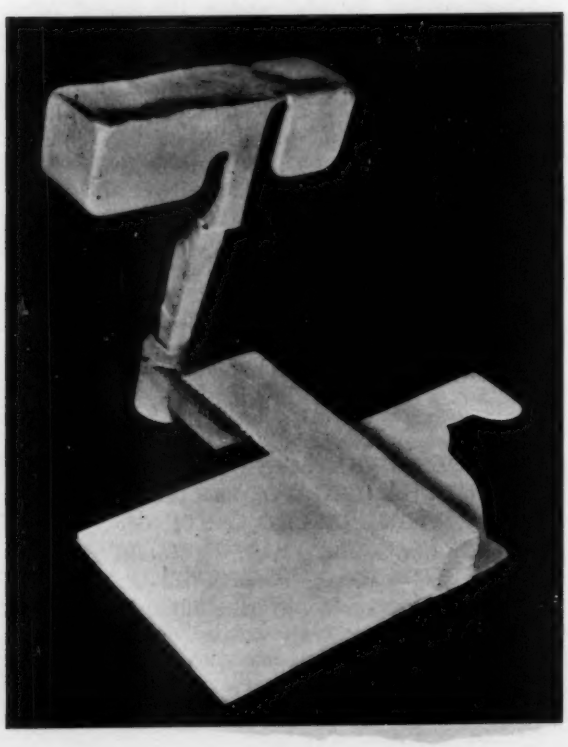


Fig. 12 — Test plate used for studying effect of mold coatings on soundness and mechanical properties of test plates. Identical plates on each side of the runner are 5 x 5 x 1/8-in. thick. In test casting, brown hexachloroethane mold coating promoted complete filling of the left plate. The plate on the right is not filled. No venting was employed.

different coatings. These were radiographed and heat treated, and mechanical properties were determined by cutting test bars from the castings.

### *Castings Poured; Coatings Effect on Fluidity and X-ray Quality*

Two coatings were examined (hexachloroethane and carbon black), and the alloy poured was aluminum-4.5 per cent copper with 0.18 per cent titanium added as a grain refiner. Melting, degassing and temperature measurements were as previously described.[2] Molds were hand-rammed in 12 x 12 in. flasks, using bentonite bonded green sand of 110 AFS fineness number. Hexachloroethane was sprayed from an ether solution to a thickness of about 0.004 in. Carbon black was applied with a smoking acetylene torch.

Eleven castings were poured at temperatures ranging from 1305 F to 1190 F (707 to 593 C). At temperatures below 1272-1300 F (688-704 C) misruns were apparent on the uncoated plates. At 1300 F (704 C) radiography indicated the plates did not completely fill, and so temperatures higher than 1300 F (704 C) would be necessary in practice to produce a sound casting if no coating were used. The hexachloroethane and carbon black coatings permitted filling the plates completely (visually and radiographically) at pouring temperatures as low as 1200 F (649 C). Both coatings improved fluidity to about the same extent, and neither had any deleterious effects on X-ray soundness or surface quality of the cast plates. Figure 13 is a sketch of three typical plates produced.

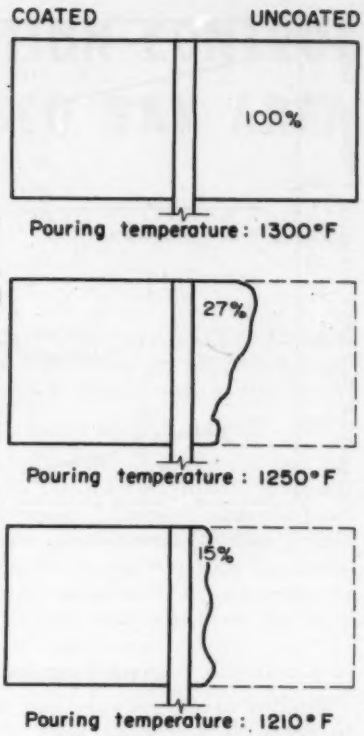



Fig. 13 — Appearance of typical test plate castings. Left side coated with carbon black; right side uncoated.

### *Coatings Effect on Mechanical Properties*

Flat test bars (0.100 in. thick x 0.500 in. x 1.00 in. gage) were cut from the plates in two locations. One location was adjacent to the runner, and the other was one in. from the outer edge of the plates (parallel to the runner). All test bars were cut from areas of plates that were radiographically sound, and were heat-treated:

Solution treatment 960 F (516 C), 16 hr, water quench.
Hold at room temperature 24 hr.
Age at 310 F (154 C) for 15 hr.

Location of the test bars had only a relatively small effect on mechanical properties. Bars farthest from the runner had slightly higher properties, but the average increase was only about 1.4 per cent in elongation and 2000 psi in ultimate tensile strength. The coatings had little or no effect, per se, on mechanical properties. Properties were the same at a given pouring temperature regardless of whether molds were coated or not (Figs. 14, 15). However, the coatings had a major beneficial effect on properties indirectly by permitting substantial reductions in pouring temperature.

For example, castings poured in coated molds at 1200 F (648 C) possessed mechanical properties of the order of 55,000 psi ultimate tensile strength, 36,000 psi yield strength and 8-10 per cent elongation. Without the mold coating, temperatures in excess of 1300 F (704 C) are essential for mold filling, and resulting mechanical properties are decreased to less than 50,000 psi tensile strength, 36,000 psi yield strength and 4 per cent elongation.

The above results are readily explainable in view of the concepts presented previously. Briefly, these coatings affect heat transfer conditions only during the first small fraction of a second. Afterward (during nearly all the solidification process) their effect on heat transfer is negligible. Thus, the coatings act to slow heat transfer at a desirable time (during filling of the mold), but later do not prevent the rapid heat extraction necessary to produce a sound, strong casting.

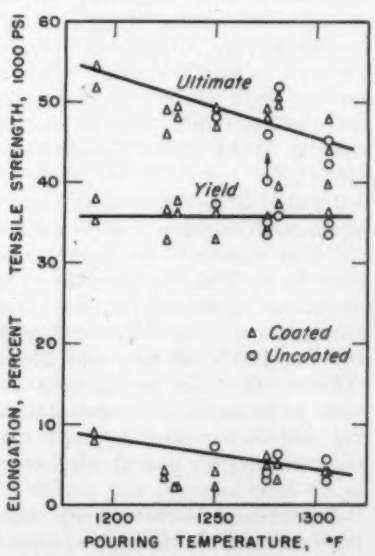



Fig. 14 — Hexachloroethane mold coating effect on mechanical properties of test plate castings (from two different heats).

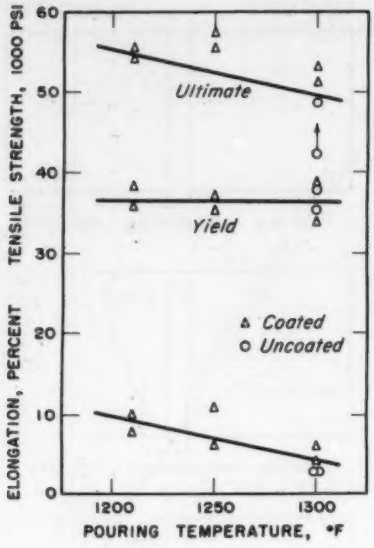



**Fig. 15 — Carbon black mold coating effect on mechanical properties of test plate castings (from one heat).**

## CONCLUSIONS

Fluidity of Al - 4½ - per cent copper alloy poured in zircon sand is lower than in silica sand. Reductions found were 55 per cent at 1250 F (677 C) and 22 per cent at 1400 F (760 C). The lower fluidity in zircon sand is primarily due to the greater heat diffusivity of zircon as compared to silica. Variations in grain size affect fluidity only when the grains become sufficiently coarse that rough surfaces and significant penetration result. The extra effective thickness of spirals in very coarse sand increases apparent fluidity by providing more metal which takes longer to freeze.

Moisture content of clay bonded sand (up to 3.5 per cent) has no effect on fluidity. Cereal (up to 0.75 per cent) has no effect on fluidity in green clay bonded sand. Fluidity in re-used clay bonded green sand is the same as fluidity in new sand. Ten per cent sawdust added to green clay bonded sand increases fluidity 10 per cent at 1250 F (677 C) and 13 per cent at 1400 F (760 C). Fluidity in molding sand bonded with sodium silicate is less than in green clay bonded sand. The decrease found was 45 per cent at 1250 F (677 C) and 20 per cent at 1400 F (760 C).

Fluidity in molding sand bonded with linseed oil is less than in green, clay bonded sand. The decrease is approximately 20 per cent at 1250-1400 F (677-760 C). Fluidity in molding sands bonded with phenol-formaldehyde is less than in green, clay bonded sand. The decrease is approximately 50 per cent at 1275 F (692 C) and 27 per cent at 1375 F (746 C). The fluidity in core sands (as compared with green sand) is decreased because the core sands are able to extract heat more rapidly, probably because of a higher heat transfer coefficient at the mold-metal interface.

Comparison of theory with experiment shows that improvements in fluidity obtained by use of mold coatings such as hexachloroethane or carbon black are due to reduction of the heat transfer coefficient at the metal-mold interface, rather than to any effect of these coatings on surface tension or surface oxide films. Differences in fluidity in silica sand and zircon sand can be largely accounted for by the different thermal properties of these materials.

Hexachloroethane mold coatings are effective in improving fluidity of a wide range of metals and alloys (not just aluminum alloys). Significant improvement has been found in all alloys tested, including magnesium alloys, manganese-bronze, 85-5-5-5 copper base alloy, low alloy steel and stainless steel. In 85-5-5-5 fluidity improvement was as much as 125 per cent and in the low alloy steel it was up to 154 per cent.

Mold coatings (hexachloroethane and carbon black), properly applied, have no deleterious effects on x-ray quality, surface appearance or mechanical properties of aluminum castings. Mold coatings can significantly improve mechanical properties of aluminum castings by permitting substantially reduced pouring temperatures. For example, in an aluminum-4.5 per cent copper alloy test casting, coatings permitted lowering the pouring temperature 100 F. As a result, elongation was doubled (increased from 4 per cent to 8 per cent) with a simultaneous increase in tensile strength.

## ACKNOWLEDGMENT

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# FOUNDRY AIR POLLUTION CONTROL IN THE SAN FRANCISCO BAY AREA

by B. Linsky and G. Aase


## ABSTRACT

A discussion on the basic need of an Air Pollution Control District, covering such items as localized effects and terrain of the District, the meteorological and climatic conditions that prevail in the area, pollutant measurements which have been made and a survey of total emissions is presented. An initial study of the problem was a survey of air pollution control regulations adopted in other areas, including discussions with eminent professional people in the air pollution field. There was an organization of concepts and principles, and these were talked about in many meetings with interested industry groups ending with the first draft of Regulation 2.

A condensed version of Regulation 2 is given, covering the five classes of source operations, general limitations and some of the features not found in other regulations, to tell why sampling methods were included as a part of the regulation and to give a review of sections of particular interest to foundry operators.

A literature review of the main source operations in foundry practice of foundry emissions is presented, giving a chemical and physical description of these emissions and, where possible, a rate or way to estimate the amount of emissions. Controls being installed to meet the emission limits are shown, including a breakdown of foundry source operations with a description of the control method of methods being used or installed. Some of the problems which have been encountered are included.

The control of organic vapors from core ovens and similar operations is part of future trends. An increased coverage of foundry floor operations, a discussion on expected technical advances in operating practices and control methods and devices is presented.


## INTRODUCTION

Air pollution control regulations are not new in the industrial sections of the United States. While at first they were applied to such areas as townships or within city limits, we now see them applied to whole counties and in some instances a multi-county area. Entire states are beginning to be subject to certain laws relative to air pollution.

The Bay Area Air Pollution Control District is the first of the truly metropolitan regional agencies to be established in the air pollution control field. It has recently issued what may be referred to as a West coast type of air pollution control regulation, Regulation 2. This regulation has a decided effect upon metal melting operations within its six county jurisdiction, some of which will be shown in this paper.

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## AIR POLLUTION REGULATION

On the whole there was little disagreement over the need for air pollution control regulations. However, many foundrymen were not convinced as to the need for the degree of control which was placed in the regulation. Similar legislative activity with the establishment of limits on the emission of air contaminants is under way in many new areas which are not now subject to this type of regulation.

Prior to the establishment of an air pollution control agency there has to be a need shown that either a potential problem exists or that a problem is present. This is often brought out in public hearings, usually backed with detailed technical information describing the local situation. There may be presented reports containing data from studies of meteorology, local terrain, air pollution measurements, visibility studies, etc. Evidence of damage to property or vegetation may be shown.

Evaluation of air pollution and its effects is specialized, and requires study so that intelligent analysis can be made of the data presented. Leaders and speakers for the foundry industry should have a good speaking knowledge of air pollution in order that their words are received with proper recognition.

## INVERSION CONDITIONS

Meteorological studies had an important role in setting up the San Francisco Bay Area District and its program. Two factors became apparent, the first concerning the inversion conditions which are prevalent along most of the West coast of the United States, and second, the frequency and duration of periods with light variable winds. An inversion is a stable atmospheric situation which favors the buildup of air pollutants by allowing them to pass through the inversion layer into the upper atmosphere. When this upper level air flow is shut off only horizontal wind movement remains as a means by which air pollutants can be diluted and/or removed.

Inversions in the San Francisco Bay Area are most frequent during the summer and fall months. Average frequency over this period is 86 per cent with variations from 98 per cent in July to over 70 per cent in November. It is evident that a day without an inversion is unusual during this period.

Light variable winds are considered as those winds in the zero to three mph range with no definite direc-

tion. In the San Francisco Bay Area these stagnant or light variable winds are present 22 per cent of the time, ranging from 5 per cent during July to as high as 44 per cent of the time during November. The combination of inversion layers, low surface wind speeds and mountain ranges or long valleys can lead to a serious buildup of air contaminants (see map).

## AIR CONTAMINANT MEASUREMENT

Direct measurement of some of the air contaminants or groups of contaminants are made routinely. These are associated with undesirable air pollution effects such as visibility reduction, eye irritation, odor, vegetation damage and other property damage. Among the general group most measured are the oxidants which include ozone and organic peroxides. Oxidant measurements are used as an index showing relative amounts of eye irritating compounds in the atmosphere. The available data show that the peak oxidant concentrations in the San Francisco Bay Area are about half those found in the Los Angeles Basin.

An important factor considered in establishing the Bay Area Air Pollution Control District was the expected population and industrial growth of the area. Experts are showing a doubling of the Bay Area population every 20 years, with the total industrial output doubling each 12 to 15 years. From these facts and projections it is apparent that any air pollution problems in existence at present would become increasingly serious as time goes on unless active program for control were initiated.

Some of the pertinent sections of the legislative act establishing the Bay Area Air Pollution Control District (B.A.A.P.C.D.) must be highlighted as background information that determines the form of any regulation that can be written. The law, approved in July, 1955, made no provision for a permit system in the district program. The use and issuing of permits has been a basic operating tool of air pollution control agencies for many years. District regulations have been written to be as effective as possible without the permit procedure.

The basic law sets up the legal procedures to be followed when handling violations of the regulations. The common practice of working along misdemeanor lines is not a part of the legislative act. The decision of the Hearing Board and then a Superior Court is required to curb a violation of the regulation. The legislative act also states that no order, rule or regulation of the district shall specify the design of equipment, type of construction or particular method to be used in reducing the release of air contaminants.

## NEW REGULATION

The District staff, after consultation with the Advisory Council, was directed by the Board of Directors to write a regulation controlling the emission of particulate matter from industrial sources and combustion including incineration and also controlling pollutant gases related to combustion.

Historically, all but the world's first air pollution control regulation have been used by control agencies and their legislative bodies as bases upon which to build new regulations. This was also true to a degree in the development of regulations by the B.A.A.P.C.D. A study was made of regulations already in effect in various parts of the United States. This has had double importance from an engineering viewpoint as proof that certain standards can and are being met, and that design criteria are available.

In addition to the study of existing regulations, contacts were made with many highly respected persons, considered as experts in the field of air pollution, as to new ideas, new approaches, new concepts, new standards, etc., which might be used to produce a better regulation. Another step was an inventory and classification of emissions from all types of sources within the six county area.[1] This survey was used to determine the objectives of the District program; i.e., the degree of control and cleanup required to satisfy the demands of the general citizenry for a cleaner atmosphere.

Prior to the writing of the first draft of the regulation, meetings were held with industrial and other groups in the Bay Area to obtain additional suggestions, approaches and details, and to verify the surveys. Considerable time and deliberation was taken between the first preliminary draft and the adoption of the regulation. This entailed study by industry groups, public hearings and the final resolving of the regulation through combined efforts of the Advisory Council and the District staff.

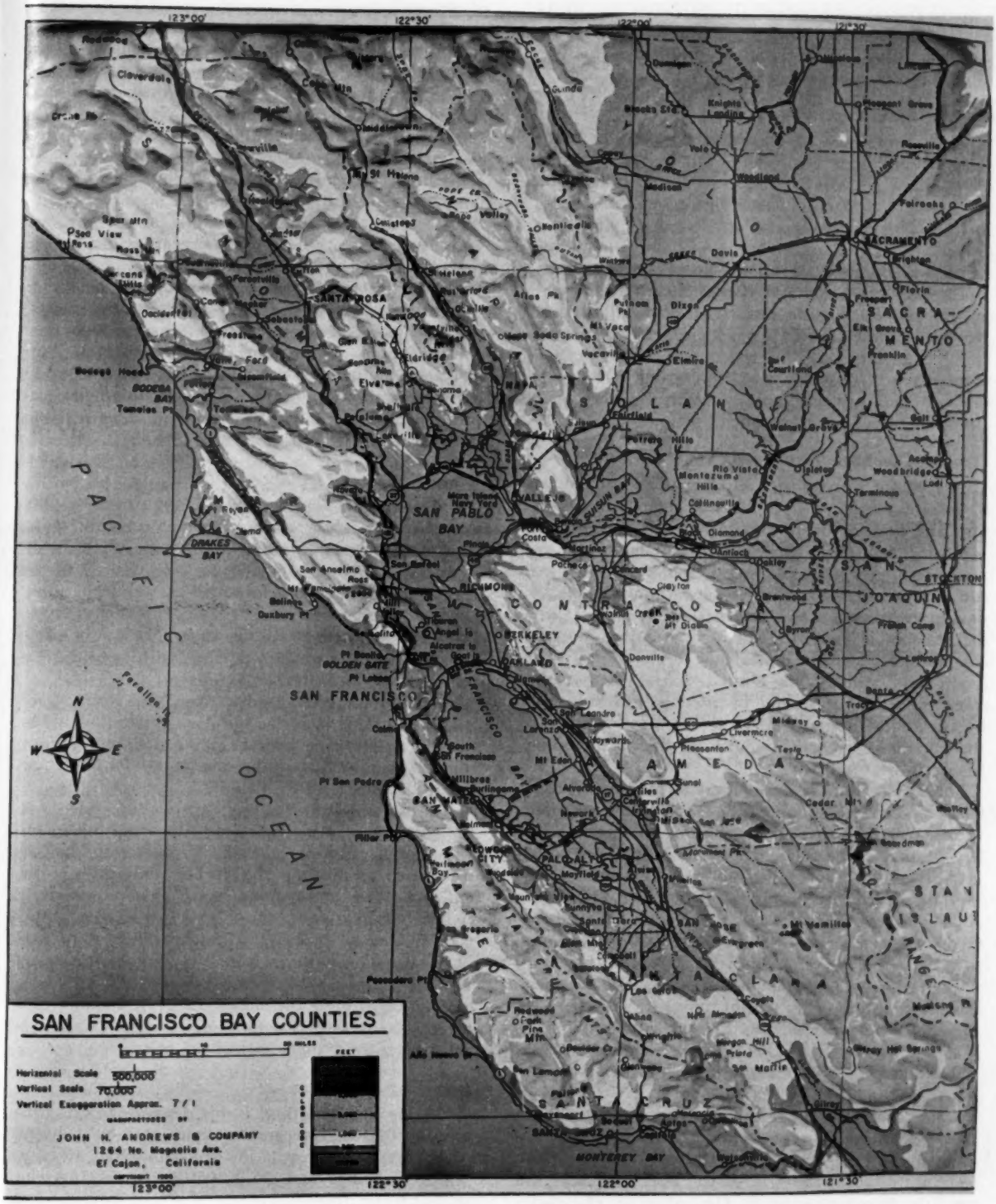
## FOUNDRY REGULATIONS

The regulation affecting metal melting and auxiliary operations is considered by some to be quite complex. Part of this can be attributed to demands for greater precision in technical interpretations and legal procedures. The regulation divides industrial and commercial operations into five classes. Operations within these classes are grouped to allow maximum coverage with a minimum number of controlling rules. The regulation places emission limits on these classes of operation; incineration, salvage, heat transfer, general combustion and general operations.

Incineration hardly needs defining because it denotes burning to reduce the bulk in the disposal of combustible waste material. Salvage operations are of some importance to foundry people, as in this category are placed such operations as the burning of insulation from copper wire or armatures. It also includes the recovery of lead or aluminum in those furnaces where a portion of the charge remains unmelted.

A heat transfer operation involves the combustion of a fuel for the principal purpose of utilizing the heat of combustion through the indirect transfer of such heat to a process material. Prime examples are the steam boiler or indirect fired ovens and driers. There is no direct contact between the products of combustion and the material being processed.

A general combustion operation consists of any operation where combustion is carried on which is not by definition an incineration, salvage or heat transfer operation. In general combustion operations, combustion gases are in direct contact with the material being processed. Examples of general combustion operations are a cupola operation, an electric arc




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**Map of area under the jurisdiction of the San Francisco Bay Area Air Pollution Control District.**

furnace, reverberatory furnace, cement kiln and most dryer operations.

General operations cover all operations not utilizing combustion. Examples are foundry shakeouts, sand handling and screening. Induction melting usually comes under this heading.

### *Opacity Limitations*

All of the classes of operation are subject to the Ringelmann and Equivalent Opacity limitations. These place a 3 min limitation in any hour on emissions as dark or darker than Ringelmann No. 2, or of such opacity as to obscure an observer's view to a degree equal to Ringelmann No. 2. Additional reference will be made to this section of the regulation.

The regulation contains specific requirements concerning sulfur dioxide emissions from all classes. Normally sulfur dioxide is not a problem with foundry people, but those interested will find this new approach to a problem most interesting. Sulfide ore roasting and used automobile battery recovery operations are directly concerned with this problem.

### *General Combustion and Operation*

The major percentage of operations carried on within the foundry will fall into the general combustion and general operations groups. The regulation places the same basic air pollution abatement requirements on both classes. Measured emissions from these two classes are not subject to a carbon dioxide or oxygen correction. This should be remembered when making comparisons with existing regulations in other communities, because it has long been a subject of controversy around the United States whether the same rules can or should be used to govern steam boilers and a cupola. Under the Bay Area regulation the cupola is general combustion, the steam boiler furnace is heat transfer.

Foundry operations, like others in the general combustion and general operations classes are required to meet all the provisions pertaining to these classes. One is that particulate matter in the exhaust gases cannot exceed 0.30 grains/standard dry cu ft. However, in nearly all instances other, more restrictive, rules will govern; opacity or the process weight rule. Opacity has already been explained, but a related rule in the regulation stipulates that opacity determinations cannot be made within a bonafide building. Basically, the process weight type of rule sets up the allowable emission rates for particulate matter as a small percentage of the amount (weight) of material being processed in a unit of time, usually per hour.

The process weight rule has an interesting origin.[2] It was based on many measurements made principally on metallurgical operations. Two additional factors were introduced; the size of the operation and the required degree of control. The collection efficiencies required were those that could be obtained with commercial equipment that was then within economic reach of industry. Large industrial units constitute a greater point source of atmospheric contamination, and can usually handle higher efficiency collection more economically than small units.

The process weight concept does not take into consideration any relationship between the volume of effluent gases and the weight of solids discharged at the stack. It simply and effectively limits the amount of solids discharged and makes it unnecessary to consider dilution in control standards.

The first process weight rule that is known of was adopted by the Los Angeles County Air Pollution Control District in March 1949. This means that for over ten years the designers and manufacturers of air pollution control equipment have been doing the required engineering and have been installing equipment which meets this rule. This more than ten years' experience is a strong deterrent against changing any of the figures in the rule.

About the only change that has been made is for those industries with a process weight rate over 60,000 lb/hr. In the San Francisco Bay Area regulation the emission limits have been increased over the Los Angeles allowance for processes beyond the 30 ton an hour size with the numbers fitted into a mathematical formula. Table 1 shows the allowable rates of emission based on the process weight rate.

### *Other Parts of Regulation*

A discussion of the regulation is not complete without describing some sections written into the regulation at the specific request of the Advisory Council. A five year "Grandfather Clause" was placed in the regulation, giving an additional 25 per cent allowance on the process weight rule. This pertains to processes that were in operation at the date of adoption, and expires on May 4, 1965 with certain filing requirements by May 4, 1963.

A minimum "grain loading floor" has been placed in conjunction with the process weight rule. The process weight rule thus modified does not require the reduction of particulate matter in the source gas below the limits set in Table 2.

A minimum grain loading floor was also placed in the regulation in conjunction with the Ringelmann and Equivalent Opacity rule. An operation shall be considered as meeting the visible limitations if the emission does not contain more than $n$ grains of particulate matter/standard cu ft, where $n$ equals 0.12 divided by $L$, and $L$ is the significant dimension of the emission point in feet. $L$ is equal to the square root of the area of the emission point.

The use of these two minimum grain loading rules are subject to several qualifications which can best be interpreted by actual study of the regulation.

### *Sampling Procedure*

Among the definitions is one describing particulate matter. This is given greater significance by including as a part of the regulation the sampling procedure which is being used. In actuality this arrangement gives greater precision as to what can be measured, and what constitutes particulate material. Every attempt is made to determine particulate matter as it exists at the point of sampling with a special provision for gases over 500 F (260 C).

Another feature of the regulation is the initial allowance of time to bring an existing operation into

compliance. The Air Pollution Control Officer was authorized, until Oct. 1, 1960, to receive detailed program schedules and grant time beyond the effective date of Jan. 1, 1961 for necessary engineering, procurement, fabrication, installation and adjustment when there was and continued to be a showing of satisfactory progress toward compliance with the regulation. Most of the foundry compliance schedules were straightforward engineering task schedules. The problems of foundries that failed to make their moves are being dealt with now to avoid giving them competitive advantages over their complying neighbors.

Core ovens are a source of fumes and odors. These units are now subject to the visible emission limits which usually, but not always, result in controlling the odorous gases. A future regulation is planned to provide a general coverage of hydrocarbon emissions from general combustion and general operations. Additional formal limitations will then be placed on core oven operations.

During the time when meetings were being held with various industry people including foundrymen, it became apparent that there was a general lack of knowledge among these groups as to what was being emitted from either their individual processes or their plants. This became important when they were evaluating the economic effect of the proposals on degree of control, for the degree of control decides the final cost of air pollution abatement installations.

### *Fine Particle Control*

The regulation of the B.A.A.P.C.D. has been called a "fine particle control" regulation. This means that a knowledge of the physical as well as the chemical properties of the emission must be known. Particle size determinations are expensive and time consuming and are often given the role of a research project. This information is not always readily available. Some of it is in the files of the designers and manufacturers of air pollution control equipment. A few large operating corporations with specific engineering divisions that handle dust and fume problems also have such data.

The largest concentrated sampling program of emissions from industrial sources of air pollution has been conducted by the Los Angeles County Air Pollution Control District. Some of their findings have been released as technical progress reports and reports of investigations. Many of the original studies[3] included particle size distribution as well as emission rates. They have found that an average cupola will discharge about 17 lb of particulate matter for every ton of metal processed, and that about 14 per cent by weight of this particulate matter will be below 10 micron in size with 7.5 per cent by weight below 5 micron diameter.

Studies on electric arc furnaces for steel production show an average of 8 lb of particulate matter per ton of metal processed with practically all under 3 microns in diameter. Zinc oxide fume is extremely fine, most of it being submicron in size. Table 3 gives a review of emission rates from various metal melting operations in Los Angeles County.

The American Foundrymen's Society has issued two publications that adequately list the operations in the foundry that are a source of air contaminants. The "Foundry's Attack on Air Pollution" published in the Sept. 1956 issue of MODERN CASTINGS, and the FOUNDRY AIR POLLUTION CONTROL MANUAL, are both recommended reading for anyone as an aid to evaluation of air pollution control legislation or the control of a specific foundry air pollution problem.

**TABLE 1* — ALLOWABLE RATE OF EMISSION BASED ON PROCESS WEIGHT RATE[a,b]**

| Process Weight Rate | | Rate of Emission, | Process Weight Rate | | Rate of Emission, |
|---|---|---|---|---|---|
| Lb/Hr | Tons/Hr | lb/hr | Lb/Hr | Tons/Hr | lb/hr |
| 100 | 0.05 | 0.551 | 16,000 | 8.0 | 16.5 |
| 200 | 0.10 | 0.877 | 18,000 | 9.0 | 17.9 |
| 400 | 0.20 | 1.40 | 20,000 | 10.0 | 19.2 |
| 600 | 0.30 | 1.83 | 30,000 | 15.0 | 25.2 |
| 800 | 0.40 | 2.22 | 40,000 | 20.0 | 30.5 |
| 1,000 | 0.50 | 2.58 | 50,000 | 25.0 | 35.4 |
| 1,500 | 0.75 | 3.38 | 60,000 | 30.0 | 40.0 |
| 2,000 | 1.00 | 4.10 | 70,000 | 35.0 | 41.3 |
| 2,500 | 1.25 | 4.76 | 80,000 | 40.0 | 42.5 |
| 3,000 | 1.50 | 5.38 | 90,000 | 45.0 | 43.6 |
| 3,500 | 1.75 | 5.96 | 100,000 | 50.0 | 44.6 |
| 4,000 | 2.00 | 6.52 | 120,000 | 60.0 | 46.3 |
| 5,000 | 2.50 | 7.58 | 140,000 | 70.0 | 47.8 |
| 6,000 | 3.00 | 8.56 | 160,000 | 80.0 | 49.0 |
| 7,000 | 3.50 | 9.49 | 200,000 | 100.0 | 51.2 |
| 8,000 | 4.00 | 10.4 | 1,000,000 | 500.0 | 69.0 |
| 9,000 | 4.50 | 11.2 | 2,000,000 | 1000.0 | 77.6 |
| 10,000 | 5.00 | 12.0 | 6,000,000 | 3000.0 | 92.7 |
| 12,000 | 6.00 | 13.6 | | | |

a. Sections of major importance with reference to this table are ss 2024, 2027, 3213, 3214 and 6112.2.

b. Interpolation of the data in table for process weight rates up to 60,000 lb/hr shall be accomplished by use of the equation $E = 4.10\ P^{0.67}$, and interpolation and extrapolation of the data for process weight rates in excess of 60,000 lb/hr shall be accomplished by use of the equation;

$E = 53.0\ P^{0.11} - 40$, where E = rate of emission in lb/hr and P = process weight rate in tons/hr.

* Bay Area Air Pollution Control District, Regulation 2, May 4, 1960.

**TABLE 2* — MINIMUM CONCENTRATION TO BE REQUIRED[a,b]**

| Source Gas Volume, SCFM | Concentration GR/SCF | Source Gas Volume, SCFM | Concentration GR/SCF |
|---|---|---|---|
| 7,000 or less | 0.100 | 140,000 | 0.038 |
| 8,000 | 0.096 | 160,000 | 0.036 |
| 9,000 | 0.092 | 180,000 | 0.035 |
| 10,000 | 0.089 | 200,000 | 0.034 |
| 20,000 | 0.071 | 300,000 | 0.030 |
| 30,000 | 0.062 | 400,000 | 0.027 |
| 40,000 | 0.057 | 500,000 | 0.025 |
| 50,000 | 0.053 | 600,000 | 0.024 |
| 60,000 | 0.050 | 800,000 | 0.021 |
| 80,000 | 0.045 | 1,000,000 | 0.020 |
| 100,000 | 0.042 | or more | |
| 120,000 | 0.040 | | |

a. Sections of major importance with reference to this table are: ss 2024, 2030, 6112.3 and 6112.4.

b. Interpolation of the data in this table shall be based on linear interpolation between adjacent values.

* Bay Area Air Pollution Control District, Regulation 2, May 4, 1960.

TABLE 3 — EMISSIONS OF AEROSOLS FROM METALLURGICAL OPERATIONS IN LOS ANGELES COUNTY*

| Source | Emissions in lb/ton of Metal Processed | |
|---|---|---|
| | Uncontrolled | Controlled |
| Open Hearth Steel Furnaces (58 ton capacity[a] | 17.8 | 0.36 |
| Gray Iron Melting Cupolas (Avg.) | 17.1 | 0.26 |
| Less than 48 in. I.D. | 12.9 | — |
| 48-60 in. I.D. | 19.5 | — |
| Greater than 60 in. I.D. | 18.9 | — |
| Electric Steel Melting Furnaces (Avg.) | 8.6 | 0.17 |
| Less than 5 ton capacity | 10.6 | — |
| 5-20 ton capacity | 5.7 | — |
| 50-75 ton capacity | 9.6 | — |
| Melting of Red Brass | | |
| Crucible or Pot Furnaces | 3.3 | — |
| Rotary Furnaces | 21.3 | 10.1[b] |
| Reverberatory Furnaces | 16.8 | — |
| Electric Furnaces | 3 | — |
| Melting of Yellow Brass | | |
| Crucible Furnaces | 14 | 5.1[b] |
| Rotary Furnaces | — | 22.8[b] |
| Reverberatory Furnaces | — | 5.7[c] |
| Electric Induction Type Furnaces | 0.7 | — |
| Melting of Bronze | | |
| Crucible Furnaces | 3.8 | — |
| Rotary Furnaces | 30.6 | 4.7 |
| Melting of Aluminum | | |
| Crucible Furnaces | 1.9 | — |
| Reverberatory Furnaces | 5.2 | 2.1 |

a. Principally scrap remelt.
b. Using slag cover as the only control method.
c. With baghouse control.
* Air Pollution Control District County of Los Angeles, Control of Stationary Sources, Technical Progress Report, *vol. 1*, (1960) p. 31, Los Angeles, Calif.

## FOUNDRY INSTALLATIONS

Some of the foundries in the San Francisco Bay Area have not completed their engineering studies, but apparently most of them are going to use conventional, highly effective methods in bringing their operations into compliance with the regulation. The major cupola operations are installing baghouses with either stack igniters or afterburners. As far as is known, all will be using glass cloth as the filter media. Some plants will be installing heat reclaiming equipment for heating tuyere air, which will allow some utilization for production heat of the natural gas required to complete the combustion of the cupola gases for air pollution control. This exchange of heat can also help in lowering the temperature of the gases going to the collectors.

Some new ideas in gas cooling have been proposed but so far vertical towers with water sprays have been selected. A few gray iron foundries are looking to induction melting or gas fired reverberatory melting for reducing their emissions, while at the same time improving their production factors. At least two foundries have made commitments to go to gas melting.

In many instances the installing of control equipment brings added problems to the foundry operator. Space may be a problem where areas for baghouses, cooling towers, ducting and fans are not readily available and some rearrangement of facilities may be required. Electrical distribution can also be troublesome. The addition of electrical horsepower in a foundry through the past years has sometimes reached the point where the existing electrical system is operating at maximum load. The additional kw needed for a control system may require a new substation, switchgear and other auxiliary equipment.

One San Francisco Bay Area foundry encountered a situation that required an entirely new cupola charging system to be designed and installed. This, in itself, has turned out to be a major engineering undertaking.

## NONCONTROLLED SOURCES

The present regulations of the B.A.A.P.C.D. leave some types of air pollution from some types of industrial sources yet to be controlled. These concern organic vapors, odorous gases and particulates and some inorganic gases. A special study will be made of these sources including methods of control to determine equitable limits on invisible as well as visible emissions. Core oven operations will be a part of this study.

Increased coverage of foundry floor operations can be expected at some future time. Modernization and mechanization of mold handling and pouring will be a part of this study program. Any future increase in air pollution control in the foundry should be more than balanced by improvements in methods and equipment used to collect dust and fumes. Great advances have been made in the last decade and this trend can be expected to continue along with the other technical developments.

More efficient gas cooling and higher operating temperatures for control equipment will reduce the size of air pollution abatement equipment. The allowable temperatures for baghouse operation have gone from 275 F to 550 F (135 to 288 C), with test work on new filter material at 750 F (399 C) being reported. Great developments are expected along these lines, as new synthetic fibers are discovered.

Changes in melting practice, better scrap preparation and, in some cases, a decrease in scrap use, will contribute new benefits to the metal melting industry including relief from air pollution problems. Alternate alloys are being developed to replace some of the problem metals now being cast.

Comparable progress is shown for core room practice and sand molding as new materials are introduced and new techniques developed. Some have caused a major reduction in the amounts of air contaminants being emitted from these equally important foundry operations.

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# GRAPHITE FORMATION DURING SOLIDIFICATION OF CAST IRON

## *Flake, spheroidal, lacy, film and compact graphite*

by C. R. Loper, Jr. and R. W. Heine


### ABSTRACT

**The principles of graphite formation during solidification of cast irons containing flake, spheroidal, lacy, film-type and compact graphite of the type occurring in as-cast malleable iron are described. Cooling curves and quenching of small samples from important temperatures during cooling were used to show the sequence of events in the development of the various types of graphite. With respect to flake graphite, agreement with the work of previous investigators was reached. Graphite flakes were found to grow in contact with the melt and as part of the two phase austenite-graphite eutectic.**

**Compact graphite of the type present in a currently produced as-cast malleable iron was found to nucleate and grow after solidification and at temperatures at and below the $A_{cm}$ solubility curve. Film-type graphite is formed in the same way. Spheroidal graphite was observed to nucleate at temperature above the eutectic temperature range in both hypo and hypereutectic irons. Growth occurred only upon reaching the upper eutectic temperature. Growth of spheroids continues down to temperatures well below the normal lower eutectic temperature. Growth of spheroids occurs while they are encased in austenite, i.e., they are not in contact with the liquid.**

**The number of spheroids was found to increase during cooling only to the eutectic start temperature, not in the eutectic range. Substantial increase in the eutectic solidification temperature range was demonstrated for ductile irons. The principles of graphite formation set forth describe the conditions under which the various types of graphite must form and explain why mixed types are found in some compositions. The effect of magnesium, cerium and other elements on the graphite formation process is described.**


### INTRODUCTION

The solidification and graphitization of cast irons has attracted many investigators over the past years, but is still not well understood. The subject is complex, and such problems as the metastability of iron carbide, the many types of graphite which develop, inherent variables of melting and casting and the effect of small amounts of additional elements other than iron, carbon and silicon must be recognized in such a study. Because of the complexity of the problem and the variety of the forms of graphite which occur, most investigators have limited their study to narrow ranges of chemical composition, and to the investigation of only one type of graphite as being distinct from the others.

As a result, the broader viewpoint of the solidification and graphitization process over wide variations in chemical analysis and graphite form has been overlooked. The purpose here is to consider this broader viewpoint, and to show that the forms of graphite are in the end all related but differ in the mechanism of precipitation from the melt or solid.

During solidification and cooling of the iron, these forms of graphite have been observed to occur:

1. Eutectic flake graphite. The graphite present in most gray irons typifies this graphite and may include types A, B, D and E according to the current AFS classification.
2. Kish or proeutectic flake graphite in hypereutectic gray or pig irons.
3. Spheroidal graphite. The true spheroids found in ductile iron typify this form.
4. Compact graphite. This form is found in certain types of as-cast malleable irons (and graphitic steels).[10,35,48]
5. Film graphite. This type is not intentionally produced in any commercial cast iron, but is sometimes observed.[10,48]
6. Lacy graphite.

Some forms other than those listed may seem to occur, but according to present knowledge these appear to be variations of the basic graphite forms listed. The graphite forms develop during cooling of the iron from the casting temperature. This study will not be concerned with graphitization which occurs as a result of some heat treatment of solidi-

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fied and cooled cast iron such as annealing or malleablizing.

## REVIEW OF PREVIOUS WORK

Extensive reviews of the literature on the solidification and graphitization of cast irons have been compiled by Parthasarathi, Srikantiah, and Nijhawau,[17] Morrogh and Williams[15] and Gries and Maushake.[18] In this literature review, the authors present what seems to be the consensus of the various investigators as to the solidification mechanism of cast irons, and, in particular, the method of graphite formation.

The literature is fairly well agreed that eutectic flake graphite is formed while in direct contact with the liquid iron. A hypoeutectic gray cast iron would solidify in the pattern:

1. Austenite dendrites in the liquid, formed above the eutectic arrest.
2. At the eutectic arrest, the eutectic of austenite and graphite solidifies on a spheroidal crystallization front (cellular growth). Ends of the flake graphite are in contact with the liquid.
3. On completion of the eutectic solidification, the austenite of the eutectic become continuous.

A hypereutectic gray cast iron would solidify in a related fashion:

1. Above the eutectic arrest, primary crystals of kish graphite form in the liquid.
2. The eutectic solidification is similar to that of a hypoeutectic gray cast iron, except that the cells are nucleated as kish graphite.
3. On completion of solidification of the eutectic, the structure consists of long flakes of kish graphite and eutectic flake graphite similar to that obtained in the hypoeutectic gray cast iron.

The literature is divided, however, on the solidification mechanism involved in a cast iron of lower carbon and silicon content or a cast iron cooled at a faster rate. Graphite in these alloys would be present as type E or type D[19] undercooled graphite. Morrogh,[15] Eash[24] and others state that undercooled graphite forms on decomposition of iron carbide solidified in the eutectic. Hultgren, Lindbolm, and Rudberg[29] and others have shown that type D graphite may form as a cellular structure with the graphite flakes in direct contact with the liquid.

The solidification of cast irons containing flake or undercooled graphite involves many anomolies which require explanation, correlation or proof. Forty-one of the more important anomolies have been listed by Morrogh and Williams,[15] and will not be presented here.

There are almost as many theories in the literature concerning the solidification of ductile irons as there are investigators. Gries and Maushake[18] group these theories according to three problems:

1. The nuclei from which the graphite spheroids originate. Twelve different theories are presented as to the structure of the nuclei. The majority of the authors believe this nucleus to be some type of carbide.
2. The crystal structure of the spheroids. The majority of the investigators believe spheroids are crystal aggregates, the base planes of which are normal to the radius.
3. The manner of crystallization of the spheroids and the matrix and the relationship to them. Eighteen theories on the mode of crystallization of spheroids and matrix are advanced, the majority of which state that supercooling is observed. The theories are equally divided as to whether it is necessary for the graphite nodule to be surrounded by austenite or whether the graphite nodule develops while in intimate contact with the liquid.

In general, then, the mode of solidification of ductile iron is not well established, at least as far as the formation of graphite nodules is concerned. There is no mention in the literature of the solidification mechanism of cast irons containing compact or film or lacy type graphite.

## EXPERIMENTAL PROCEDURE

The solidification and graphitization process was studied using the technique of interrupted slow cooling of small samples. Several iron-carbon-silicon alloys were selected for study according to the type of graphite expected in their microstructures. These alloys were:

1. Alloy A — 0.98% C, 3.52% Si. A cast iron which contains compact graphite.
2. Alloy CS — 1.55% C, 2.33% Si. An as-cast malleable iron containing compact graphite.[10,35]
3. Alloy F — 3.12% C, 1.70% Si. A hypoeutectic flake graphite cast iron.
4. Alloy GR — 3.90% C, 2.16% Si. A hypereutectic flake graphite cast iron.
5. Alloy GR′ — 4.00% C, 2.30% Si. A commercially produced ductile cast iron, hypereutectic in composition.
6. Alloy GSW — 3.54% C, 2.14% Si. A commercially produced ductile cast iron, hypoeutectic in composition.

Additional details of chemical composition of these alloys are given in Table 1. Samples of these alloys to be used in the interrupted cooling experiments were obtained by drawing a sample of the molten iron into a ¼-in. inside diameter glass tube and immediately quenching it into water. Samples were then broken off into sections about 1⅜-in. long.

The interrupted slow cooling of samples was carried out in the furnace shown in Fig. 1. It consists of a cylindrical furnace heated by means of a hollow cylindrical globar heating element, 2 in. inside diameter x 16 in. long. The electrical connections on either end of the vertically mounted heating element were water cooled and were surrounded by 6 in. of insulating material. The power input to the furnace could be varied so that the heating and cooling rate of the furnace could be controlled.

Samples were melted down in fused silica tubes, ⅝-in. inside diameter and fused closed on one end.

TABLE 1 — CHEMICAL COMPOSITION

| Alloy | Analysis, % C | Si | S | P | Mn | Graphite Type | Iron Type |
|---|---|---|---|---|---|---|---|
| A | 0.98 | 3.52 | 0.061 | — | 0.601 | Compact | As-cast Malleable |
| CS[2] | 1.55 | 2.33 | 0.099-0.127 | 0.051 | 0.58-0.63 | Compact | As-cast Malleable |
| F | 3.12 | 1.70 | 0.061 | — | 0.601 | Flake | Gray Iron |
| GR | 3.90 | 2.16 | 0.012 | 0.0161 | — | Flake | Gray Iron |
| GR[3] | 4.00 | 2.30 | 0.008 | 0.016 | — | Spheroidal | Ductile Iron |
| GSW[4] | 3.54 | 2.14 | 0.011 | 0.0161 | — | Spheroidal plus Flake | Ductile Iron |

1. Estimated analysis.
2. 0.01% B also present.
3. About 0.050% Mg also present.
4. 0.051% Mg present before re-melting, 0.028-0.031% Mg present in re-solidified samples.

These fused silica tubes containing the samples were shielded from the direct radiation of the heating element by means of a one in. inside diameter zircon combustion tube. A 2¼-in. long zone over which the temperature gradient was ± 3 F was produced by this arrangement.

Several platinum 10 per cent rhodium-platinum thermocouples were used in the assembly and are shown in Fig. 1. One, connected to the furnace controller, was placed between the combustion tube and the heating element. The second couple was placed inside the zircon combustion tube in such a manner that it would be within ⅛-in. of the sample in the fused silica tube during the experimental runs. This couple was connected to a recorder having a 1000 to 3000 F scale and a chart speed of 4.8 in./hr.

### *Cooling Curves*

Cooling curves for thermal analysis of the alloys investigated were obtained by placing another platinum 10 per cent rhodium-platinum thermocouple directly into the melted sample. A fused silica tip was fixed onto the end of a high temperature refractory insulator to protect the couple from the molten metal. The cooling curve obtained from this couple immersed in the molten metal was compared with the cooling curve of the couple adjacent to the sample. Figure 2 shows the results obtained with these two couples for one of the cast irons studied (3.12% C, 1.70% Si). After having obtained these cooling curves, the couple adjacent to the fused silica tube was then used to determine the temperature of the samples in subsequent experimental runs.

The samples to be studied were placed in the fused silica tubes, held in the furnace at temperature (generally 2700-2750 F) for 10 min, and then cooled in the furnace at 10-15 F/min. After the samples had furnaced cooled to various specific temperatures, they were quenched into cold water. The samples were then examined metallographically. Determination of the number of graphite spheroids per sq mm or per cu mm were made by the method described in a reference.[5]

The importance of cooling rate of the specimen should be recognized. Although the specimen is small,

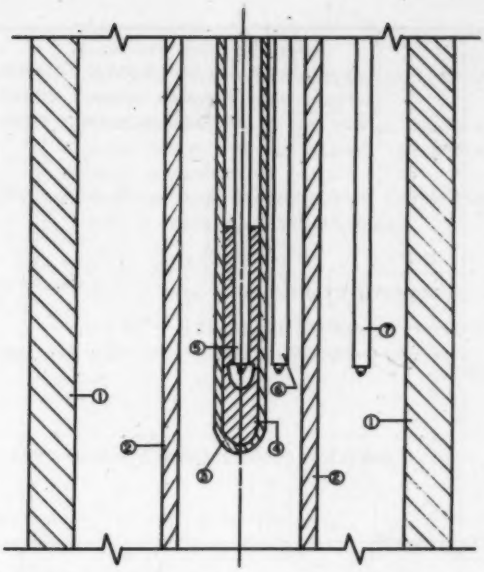



Fig. 1 — Schematic diagram showing globar tube furnace, tube holding molten sample and arrangement of three thermocouples for measuring and controlling temperature of the sample. (1) 2 in. I.D. globar heating element. (2) One in. I.D. zircon combustion tube. (3) 5/16-in. fused silica tube sealed at one end. (4) Molten cast iron sample approximately 8 grams in weight. (5,6,7) Thermocouples.

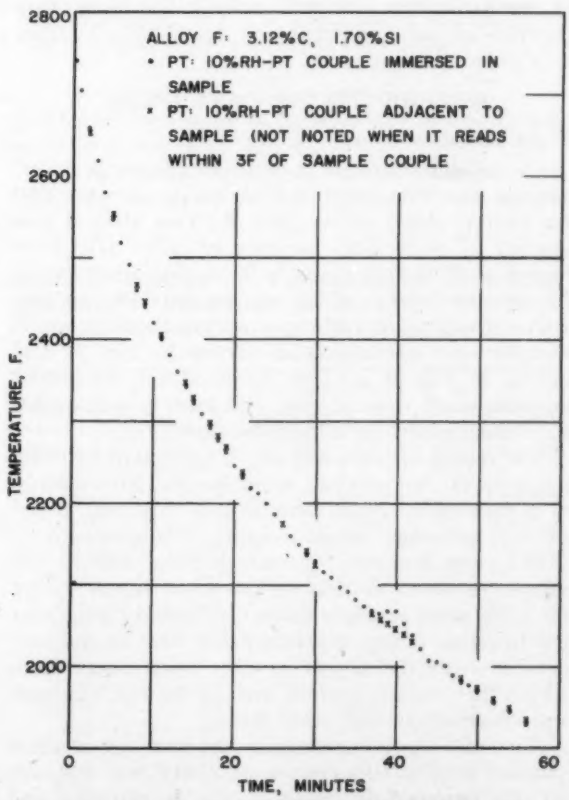



Fig. 2 — Comparison of cooling curve of molten sample with cooling curve of thermocouple immediately outside tube holding sample, TC 5 vs. TC 6 in Fig. 1.

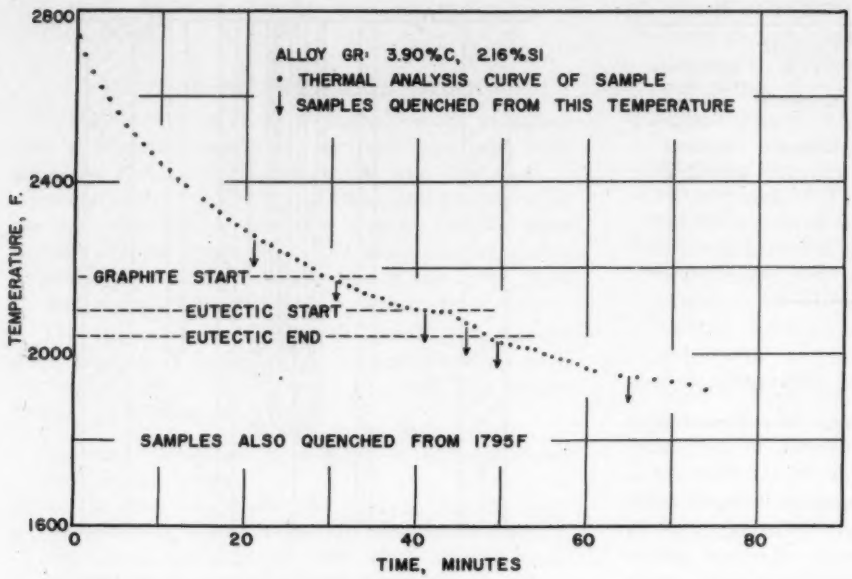



Fig. 3 — Cooling curve of alloy GR.

approximately 8 grams, it was cooled from the 2700-2750 F range to just below 2000 F in about 40 to 50 min for most samples, and from 2250-2300 F in about 13 min in other cases. This is equivalent to the cooling time in an infinite 4 in. square bar or 2 in. plate section of cast iron in molding sand.

In addition to the solidification studies performed in the laboratory, thermal analysis and quenching experiments were performed on ductile iron castings in a commercial foundry.

## HYPEREUTECTIC CAST IRONS

### *Flake Graphite Irons*

The solidification and graphite formation in hypereutectic cast irons was studied in an alloy of 3.90 per cent C and 2.16 per cent Si. This alloy is designated as alloy GR. Samples of alloy GR were heated to 2700-2750 F and held at that temperature for 10 min before cooling was started. The cooling curve of this alloy, and the temperatures from which samples were quenched, are shown in Fig. 3. Also shown in Fig. 3 are the temperatures designated graphite start, eutectic start and eutectic end which were determined metallographically.

The results of this study are in agreement with the majority of the previous investigators. No evidence of a proeutectic constituent of any type was noted in the quenched samples until a temperature of 2190 F was attained. Quenched from 2190 F, the sample consisted entirely of quenched liquid except for a few short graphite flakes distributed throughout the structure. These graphite flakes were in intimate contact with the liquid at the temperature from which they were quenched, and no shell of austenite was observed around these flakes.

Austenite was not present in the structure of these samples until a temperature of 2100 F was reached. At this temperature, eutectic cells of austenite and flake graphite were nucleated (by the proeutectic graphite flakes) and grew with decreasing temperature. These eutectic cells were observed to grow radially and with the ends of the graphite flakes in contact with the liquid metal. Growth of both the kish graphite (proeutectic graphite) and the eutectic flake graphite took place with the graphite flakes in direct contact with the liquid metal. The growth of the eutectic in this manner is the same as that described by Boyles.[23]

Solidification of the austenite-flake graphite eutectic took place over a temperature range of about 60 F. The eutectic freezing range was established metallographically at 2100-2040 F, and is in agreement with the thermal analysis data presented in Fig. 3. When eutectic cells nucleate at 2100 F and grow down to 2040 F at the 2.16 per cent Si level, random type A graphite flakes are expected and were observed. Since the microstructure of kish and eutectic cell growth during solidification have been well illustrated with photomicrographs,[15,23,24,49,50] these microstructures will not be shown again here.

In summary, hypereutectic gray cast irons, as characterized by alloy GR at 3.90 per cent C and 2.16 per cent Si solidify in the manner:

1. Formation of flakes of proeutectic graphite (kish) in the liquid melt at about 2190 F.
2. Growth of kish with decreasing temperature and subsequent possible flotation of this graphite due to the difference in specific gravity between the graphite and the liquid metal.
3. Nucleation of the cellular eutectic by the kish graphite at about 2100 F and growth of the eutectic cells and the kish graphite between 2100 and 2040 F. At 2040 F, solidification of alloy GR was completed.

### *Spheroidal Graphite Irons*

The formation of graphite as spheroids in cast irons was studied using alloy GR' (4.00% C, 2.30% Si). This alloy is a magnesium-treated, commercially-produced ductile cast iron of essentially the same analysis

as alloy GR, the hypereutectic gray cast iron. In order to prevent the loss of magnesium which occurred while holding the sample at 2700-2750 F for 10 min, it was necessary to modify the normal procedure followed. Without modifying the procedure, a flake graphite structure, like that of alloy GR, was obtained.

In addition to the effect of exposure time at 2700 F, loss of magnesium also occurred in those samples subjected to elevated temperatures of long periods of time while they were slow cooled. On the other hand, it was found that heating to temperatures lower than 2700 F for short times (4 min) was insufficient to dissolve the graphite spheroids in the structure. Although the sample was completely free of graphite initially, graphite spheroids developed during heating. This effect has been shown in Fig. 4, where it may be noted that the number of spheroids in the structure decreased as the temperature at which the sample was held for 4 min was increased. A temperature higher than 2600 F was necessary to dissolve all of the spheroids in 4 min.

In order to study the nucleation of graphite spheroids in this alloy, it was necessary to select combinations of temperatures and exposure times which would be sufficient to dissolve all graphite spheroids present in the sample, and at the same time retain an adequate amount of magnesium to insure spheroidal graphite formation during subsequent cooling. The number of spheroids retained in the melt after different temperature time cycles is given in Table 2. Table 2 also demonstrates the well-known fact that time is required to dissolve graphite when cast iron is heated into the molten state, and that this solution time is temperature dependent.

Using this necessary modification of the procedure already set forth, it was found that small, round graphite nuclei dissolve in 4 min at 2700 F and then reappear on cooling to about 2500 to 2450 F. These nuclei are present in all samples of alloy GR′ quenched from lower temperatures. Figures 5 and 6 show examples of this graphite particle in the unetched and etched condition. These graphite nuclei have been observed only in cast irons which finally solidify with graphite spheroids, as in ductile iron. It is also noteworthy that the carbides in which these nuclei are found are coarser than other carbides in the structure, and are also coarser than the carbides present in alloy GR′, the hypereutectic gray cast iron of approximately the same analysis.

### *Graphite Spheroid Formation*

At lower temperatures these graphite nuclei develop into graphite spheroids, as shown in Figs. 7 and 8, for samples quenched from 2200 F and Fig. 9 for a sample quenched from 2150 F. The spheroids present in these samples are not in direct contact with the liquid at the quenching temperature, but are instead separated from the liquid by a thin shell of austenite. Evidence of this austenite shell is shown in Fig. 8, and is best seen in Fig. 9. The quenched liquid appears as rods of carbide with areas of prior austenite. These photomicrographs show that the spheroids begin to grow rapidly between 2250 and

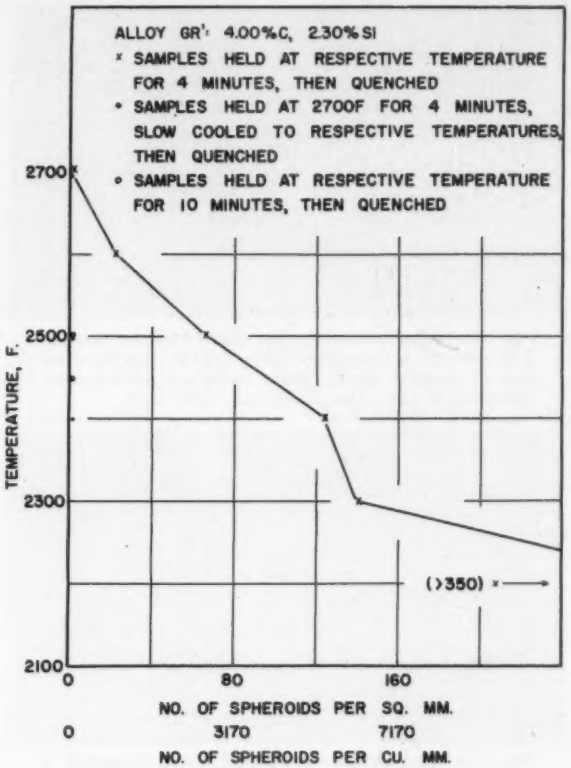



**Fig. 4 — Effect of molding melt 4 min at temperature on number of undissolved spheroids in melt. Also showing that holding 10 min at temperature is long enough to completely dissolve spheroids above 2400 F.**

2200 F and even more so between 2200 F and 2150 F and 2150 F, Fig. 8 vs. Fig. 9.

The number of spheroids formed increases with decreasing temperature for alloy GR′ until the

**TABLE 2 — EFFECT OF TEMPERATURE AND TIME FOR DISSOLVING GRAPHITE IN A 4.00 PER CENT C, 2.30 PER CENT Si, MAGNESIUM-TREATED DUCTILE CAST IRON**

| Temperatures at which samples were held, F | Time at Temperature, min | Temperature to which samples were slow cooled, F | No. of Retained Graphite, Spheroids/mm² |
|---|---|---|---|
| 2700 | 4 | — | 0 |
| 2700 | 4 | 2500 | 0 |
| 2700 | 4 | 2400 | 0 |
| 2500 | 10 | — | 0 |
| 2450 | 10 | — | 0 |
| 2700 | 4 | — | 0 |
| 2600 | 4 | — | 22 |
| 2500 | 4 | — | 65 |
| 2400 | 4 | — | 124 |
| 2300 | 4 | — | 140 |
| 2200 | 4 | — | 350 |
| 2250 | 10 | 2250 | 340 |
| 2250 | 10 | 2200 | 350 |
| 2250 | 10 | 2150 | 420 |
| 2250 | 10 | 2100 | 650 |
| 2250 | 10 | 2070 | 680 |
| 2250 | 10 | 2050 | 550 |
| 2250 | 10 | 2025 | 550 |
| 2250 | 10 | 1975 | 470 |

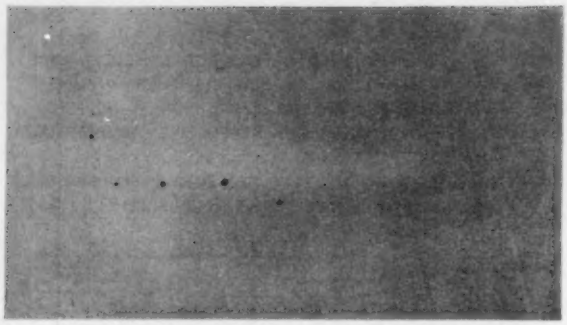


Fig. 5 — Hypereutectic ductile iron (4.00 per cent C, 2.30 per cent Si) quenched from 2250 F, showing presence of graphite nuclei. These nuclei are present from 2450-2500 F and down. Unetched. 1000×.

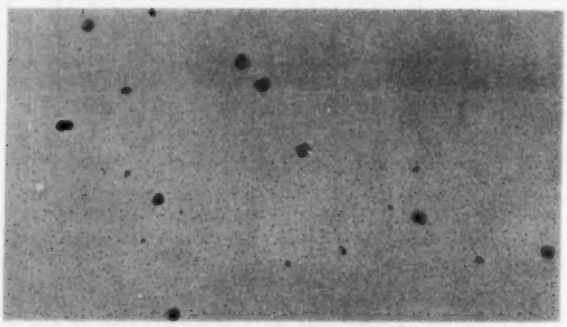


Fig. 7 — Same sample as in Figs. 5 and 6, except it was quenched from 2200 F. Growth of some of the spheroids has already begun over the temperature range of 2250 F down to 2200 F. Small nuclei are still present. Unetched. 250 ×.

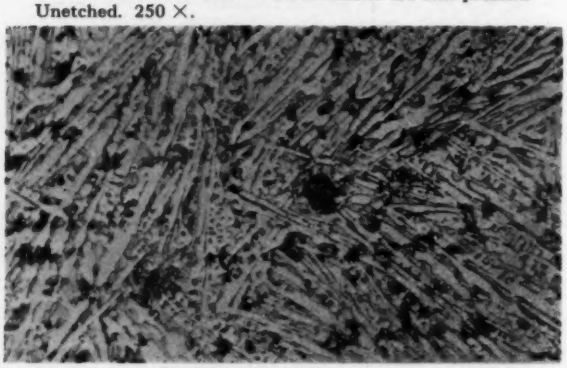


Fig. 8 — Same sample as in Fig. 7, showing austenitic shell which existed around spheroid which becomes martensite on quenching. 2 per cent nital etch. 250 ×.

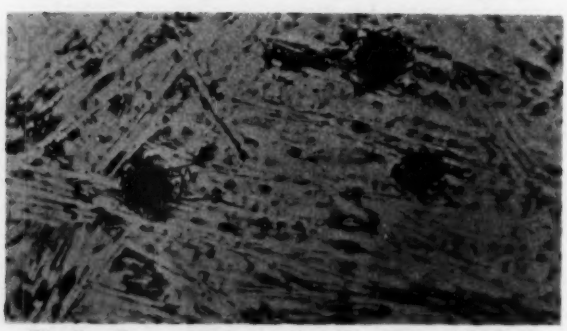


Fig. 9 — Same sample as shown in Figs. 5-8, but quenched from 2150 F. Note well developed spheroids and shell of austenite (quenched to martensite) encasing them. 5 per cent nital etch. 750 ×.

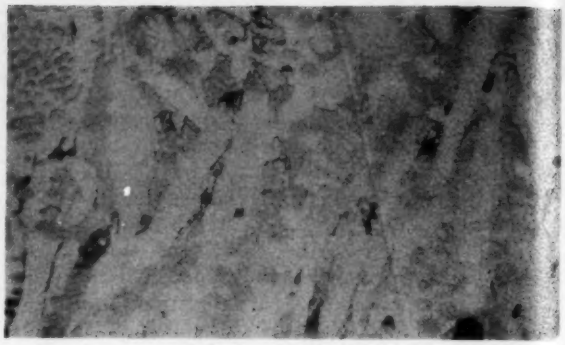


Fig. 6 — Same sample as in Fig. 5, showing location of graphite nuclei in coarse carbides of quenched liquid. 2 per cent nital etch. 1000 ×.

eutectic start temperature is reached at 2100 F. Over this temperature range, however, there is only a small increase in the size of the spheroids formed. This effect of temperature on the number of spheroids formed is shown in Fig. 10. The number of spheroids formed does not increase, and in fact decreases slightly, after the start of eutectic solidfication. After complete solidification of the alloy, the number of spheroids is essentially constant down to room temperature. The number of spheroids, illustrated in Fig. 10, should not be regarded as generally applicable to all ductile iron casting, because of the small sample size used. Rather Fig. 10 illustrates the principle of nucleation for this type of solidification.

All of the graphite spheroids formed were separated from the liquid metal by a thin shell of austenite. The thickness of this austenite shell was observed to be constant until the eutectic starts to solidify. In the temperature range of eutectic solidification some graphite spheroids rapidly grow to larger sizes, and at the same time the austenite shell surrounding them increases in thickness. Other spheroids present in the liquid do not take an active part in eutectic solidification and appear to remain dormant. Some of the smaller spheroids are re-dissolved in the liquid as the temperature decreases through the eutectic range.

The foregoing observations are illustrated by the temperature-spheroid number curve, presented in Fig. 10. The structure of a sample of alloy GR′ quenched from 2025 F, a temperature in the eutectic freezing range, is presented in Fig. 11. This photomicrograph shows both the large spheroids which have already taken part in the eutectic solidification and the dormant spheroids located in the quenched liquid. Subsequently most of the dormant spheroids will participate in the solidification of the remaining eutectic and increase in size as temperature decreases further. This process of eutectic solidification continues until all of the liquid is solidified (1985 F).

### *Thermal Analysis*

The temperatures from which samples of alloy GR′ was quenched, and the thermal analysis cooling curve for this alloy, are shown in Fig. 12. Included in this graph are the metallographically determined eutectic start and end temperatures which are in

agreement with the thermal analysis data. A considerable difference is evident in the shape of the cooling curve for the hypereutectic flake graphite iron (Fig. 3) and the spheroidal graphite iron (Fig. 12). Most notable is the extended temperature range of solidification of the ductile iron. This difference has been noted by other investigators, and is to be expected because of the difference in the mode of formation of the graphite in the two cast irons.

The austenite flake graphite eutectic of the hypereutectic gray cast iron (alloy GR′) forms by the growth of eutectic cells of $\gamma$ + graphite which are nucleated at various points throughout the specimen. These cells grow quite rapidly at first, causing the evolution of heat which prevents the temperature of the sample from falling. In this way the majority of the cellular flake graphite eutectic solidifies at almost constant temperature. The small amount of the eutectic remaining solidifies over the remainder of the eutectic range (60 F).

On the other hand, alloy GR′, the hypereutectic ductile cast iron, enters the eutectic solidification range containing graphite spheroids which are already surrounded by a thin film of austenite. Solidification of the eutectic then takes place by the precipitation of austenite onto the austenite shell surrounding the graphite spheroid, and the diffusion of carbon through the austenite shell to the graphite spheroid. In order for the spheroid to increase in size, iron must diffuse out so that carbon atoms may precipitate on the graphite spheroid.

As the thickness of the austenite shell increases, it becomes increasingly more difficult for sufficient diffusion to occur for growth to proceed. A decrease in temperature is then required to effect an increase in the driving force for the precipitation of graphite. Also, dormant spheroids with thinner shells will begin to grow as temperature decreases. Thus, the eutectic must solidify over a correspondingly greater temperature range (115 F for the ductile cast iron eutectic compared to 60 F for the gray cast iron eutectic).

It is convenient at this time to point out that

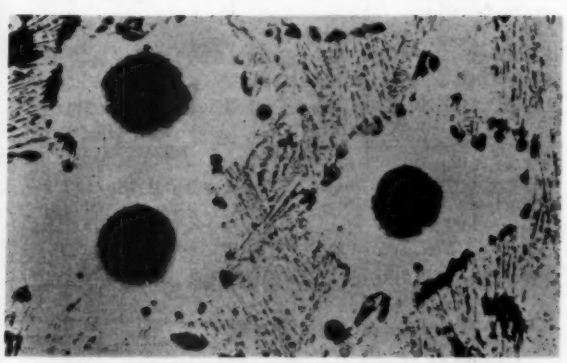


Fig. 11 — Sample of same iron, as shown in Figs. 5-10, except it was quenched from 2025 F. This temperature is in the eutectic solidification range, and the sample shows large spheroids encased in thick shells of austenite growing as part of the eutectic. Also evident are dormant, small spheroids in the quenched liquid. These will grow at temperature decreases. 2 per cent nital etch. 250 ×.

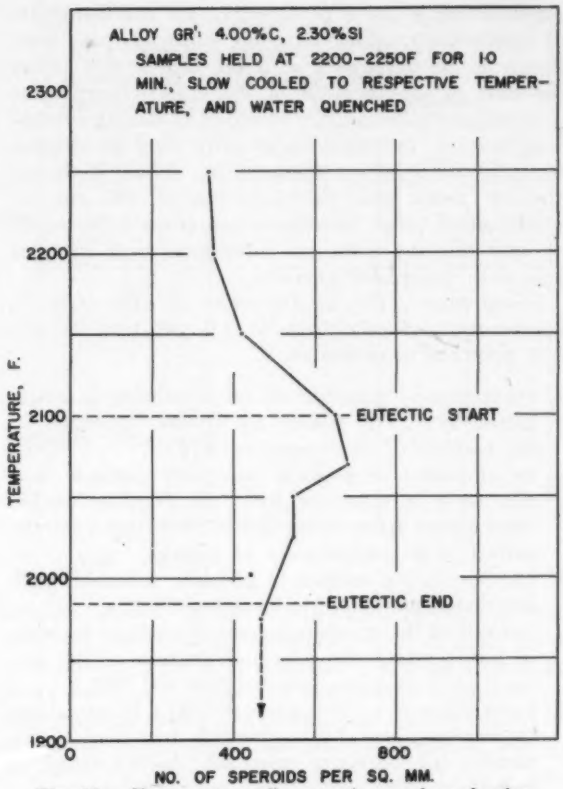



Fig. 10 — Temperature effect on the number of spheroids present on cooling from 2250 F after holding 10 min at temperature.

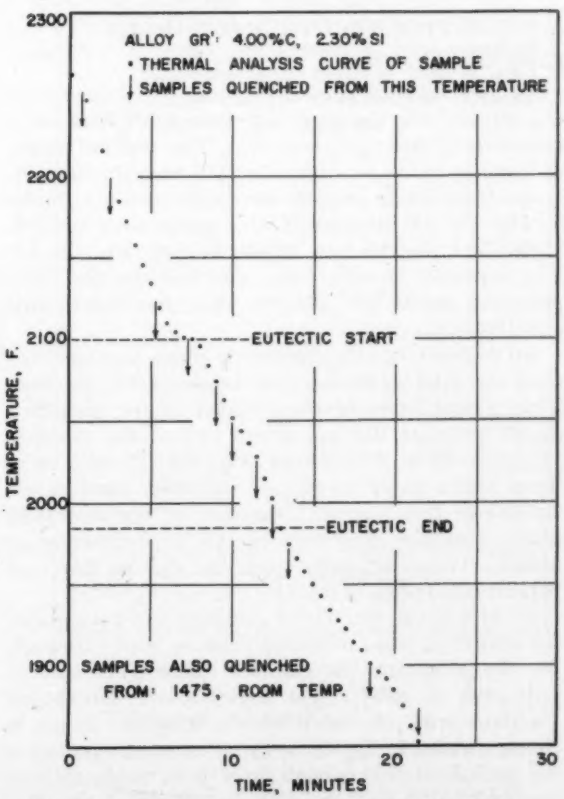



Fig. 12 — Cooling curve for ductile iron alloy GR′

supercooling is not a requirement for the formation of spheroidal graphite. Graphite spheroids have been shown to be developed at temperatures well above the start of solidification of the eutectic, and some of these grow substantially in size early during eutectic solidification. Dormant nuclei grow later in eutectic freezing and at lower temperature. Thus, it should also be noted that the widening of the eutectic solidification range for ductile cast irons is the result of, and not the cause of, solidification of the cast iron with spheroidal graphite.

In summary, the solidification of alloy GR′, a hypereutectic ductile iron of 4.0 per cent C and 2.30 per cent Si occurs as:

1. Formation of graphite nuclei beginning at about 2500-2450 F. The nuclei are found imbedded in the carbide of the quenched liquid.
2. Development of graphite spheroids, encased in a thin shell of austenite, from the graphite nuclei. This occurs from about 2210-2200 F down to the eutectic start temperature of 2100 F.
3. Increase in the number of graphite spheroids with decreasing temperature down to 2100 F.
4. Growth of the graphite spheroids without increase in number and the associated growth of the austenite shell thickness between 2100 and 1985 F.
5. Completion of solidification at 1985 F by the above mechanism, and by a gradual increase in the number of dormant spheroids participating in growth.

Other aspects of the solidification of spheroidal graphite cast irons will be taken up later.

## HYPOEUTECTIC CAST IRONS

### *Flake Graphite Irons*

An alloy of 3.12 per cent C and 1.70 per cent Si was selected for the study of the solidification of a hypoeutectic flake graphite iron. The thermal analysis cooling curve for this alloy (F) and the temperatures from which samples were quenched are shown in Fig. 13. All samples in this series were held at 2700-2750 F for 10 min before cooling was started. The liquidus, eutectic start, and eutectic end temperatures shown for alloy F were also determined metallographically.

No evidence of any proeutectic phase was observed until the samples were cooled below 2315 F. Furthermore, examination of the carbides in the quenched liquid structure did not reveal any of the graphite nuclei noted in ductile iron alloy GR′. Neither were these nuclei noted in any of the other samples examined at this analysis. Dendrites of the first solid phase, austenite, appeared in the microstructure at 2265 F. These dendrites grew in size as the temperature decreased to 2070 F.

At 2070 F, the eutectic of austenite and flake graphite started to form at several random points throughout the specimen. The graphite flakes in the eutectic cells grew at nearly right angles to the solid-liquid interface with the ends of the graphite flakes in contact with the liquid metal. Photomicrographs of the growth of this eutectic have been presented previously.[15,23,24,49] Therefore no additional ones will be shown here. The sample of alloy F was completely solidified at 2010 F.

Thus, the mechanism of the solidification and graphitization of this hypoeutectic gray cast iron, as determined by this investigation, is in accord with the general consensus of opinion expressed in the literature. In summary, alloy F solidifies in the manner:

1. Solidification of austenite dendrites beginning at about 2315 F and continuing down to 2070 F.
2. Start of eutectic cell formation at 2070 F, resulting in the formation of flake graphite-austenite eutectic cells, which are nucleated on the prior austenite dendrites and grow radially with the flakes of graphite in contact with the liquid.
3. Completion of eutectic solidification by overlapping of the cells at 2010 F.

### *Spheroidal Graphite Irons*

Alloy GSW (3.54% C, 2.14% Si) is a slightly hypoeutectic, magnesium treated, commercially produced ductile cast iron. As such, it was necessary to modify the standard procedure used in this investigation by selecting proper combinations of temperature and exposure time at temperature sufficient to dissolve all graphite spheroids present in the sample. At the same time it was necessary to retain an adequate amount of magnesium to insure solidification of the graphite in a spheroidal form. Table 3 presents data which show the effect of temperature and time on the number of graphite spheroids retained.

Two procedures were therefore used in the study of alloy GSW. The first, used for temperatures from 2700 to 2200 F, involved heating the sample up to the respective temperature, holding for 4 min and then quenching in water. The second, used for temperatures from 2200 F to room temperature, involved heating the sample to 2300 F, holding for 10 min, cooling to the respective temperatures and quenching in water. The techniques were duplicated at 2200 F and produced results which were in complete agreement.

No evidence of any proeutectic phase was evident in the samples quenched from above 2600 F. How-

TABLE 3 — EFFECT OF TEMPERATURE AND TIME FOR DISSOLVING GRAPHITE IN A 3.54 PER CENT C, 2.14 PER CENT Si, MAGNESIUM TREATED DUCTILE CAST IRON

| Temperatures at which samples were held, F | Time at Temperature, min | Temperature to which samples were slow cooled, F | No. of Retained Graphite, Spheroids/mm² |
|---|---|---|---|
| 2700 | 4 | — | 0 |
| 2600 | 4 | — | 0 |
| 2500 | 4 | — | 32 |
| 2400 | 4 | — | 50 |
| 2300 | 4 | — | 155 |
| 2200 | 4 | — | 128 |
| 2300 | 10 | 2200 | 125 |
| 2300 | 10 | 2125 | 82 |
| 2300 | 10 | 2100 | 83 |
| 2300 | 10 | 2070 | 66 |
| 2300 | 10 | 2025 | 65 |
| 2300 | 10 | 2000 | 10 |
| 2300 | 10 | 1900 | 10 |

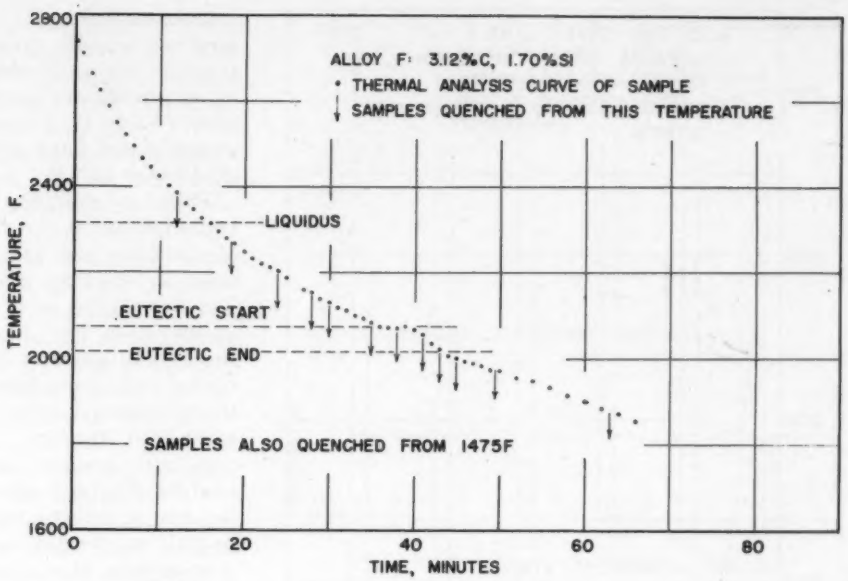



Fig. 13 — Cooling curve for hypoeutectic gray iron alloy F.

ever, at temperatures below 2600 F small graphite nuclei were formed in the carbides of the quenched liquid, identical to those of Figs. 5 and 6 formed in the hypereutectic ductile cast iron. Graphite spheroids were present in all samples obtained at lower temperatures and were surrounded by a shell of austenite, and thus were not in direct contact with the liquid. The effect of temperature on the number of spheroids is shown in Fig. 14. Data using both procedures are presented in this graph which show that the number of spheroids decrease. It may be noted that in alloy GSW, the number of spheroids present continues to decline even during the eutectic freezing range.

The average size of the spheroids increased only slightly with decreasing temperature down to the eutectic temperature, as shown in Fig. 15. During this temperature interval the thickness of the austenite shell surrounding the spheroid remained constant. This austenite envelope was evident even at temperatures where graphite was being re-dissolved into the liquid.

When the eutectic temperature range is approached, solidification may proceed in several ways:

1. By the growth of spheroids according to the mechanism already described in the case of hypereutectic ductile iron.
2. As white iron eutectic.
3. By the nucleation and growth of nonspheroidal forms of graphite such as flakes.
4. By a mixture of the preceding.

Case one, where solidification proceeds with the growth of spheroids, has already been described in connection with a hypereutectic iron. Accordingly, case 3 will be considered, wherein a new flake shaped graphite is nucleated and grows during eutectic solidification. Alloy GSW provides this case.

### *Eutectic Solidification*

Just prior to the beginning of eutectic solidification graphite spheroids are present in alloy GSW, in the number shown at 2115 F in Fig. 14 (about 90 spheroids per sq mm). Solidification of eutectic proceeds both by growth of spheroids and their austenite shells, and by the nucleation and growth of austenite-flake graphite eutectic. The number of spheroids present in the liquid metal was decreased during solidi-

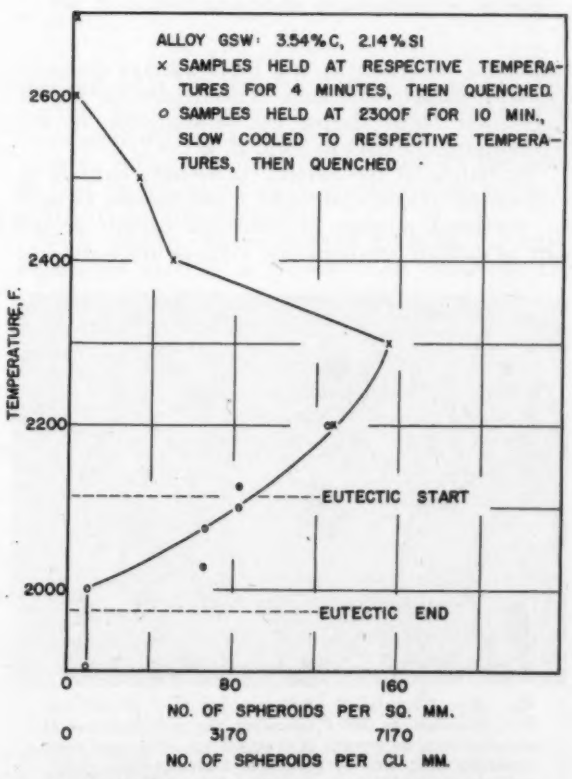



Fig. 14 — Effect of holding melt 4 min at temperature on the number of spheroids retained in the melt. Comparison of number of spheroids is made with samples held 10 min at 2300 F and then cooled.

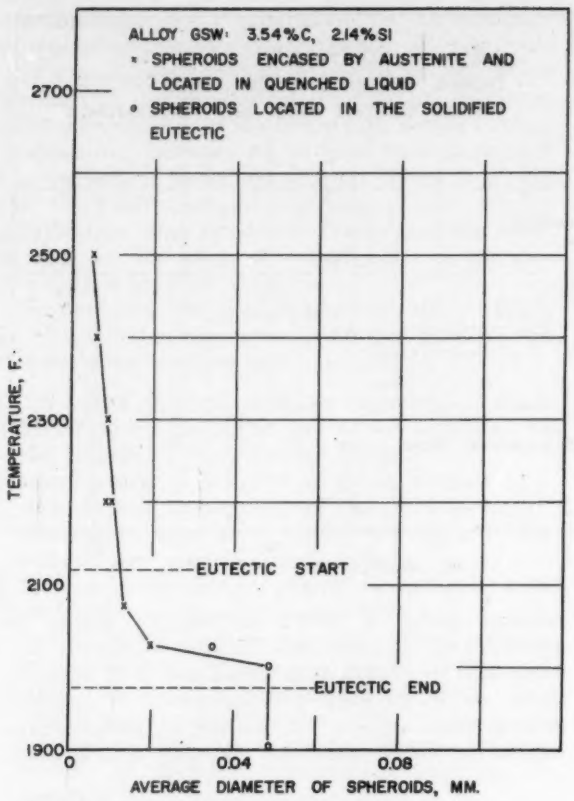



**Fig. 15 — Diameter of spheroids as related to temperature of quenching in alloy GSW.**

fication of the eutectic, and their average diameter was only slightly increased. This is similar to the observation noted during the solidification of the hypereutectic ductile cast iron, alloy GR′.

The failure of the eutectic to solidify entirely as a spheroidal graphite structure is attributable to both the decreased number of spheroids present at the start of eutectic solidification, and the nucleation of eutectic austenite and flake graphite by the austenite dendrites already present. The nucleation of flake graphite begins at about 2100 F, and is permitted by an insufficient percentage of magnesium in the alloy (Table 1). The combined spheroidal graphite-austenite and flake graphite-austenite eutectics were solidified at 1975 F.

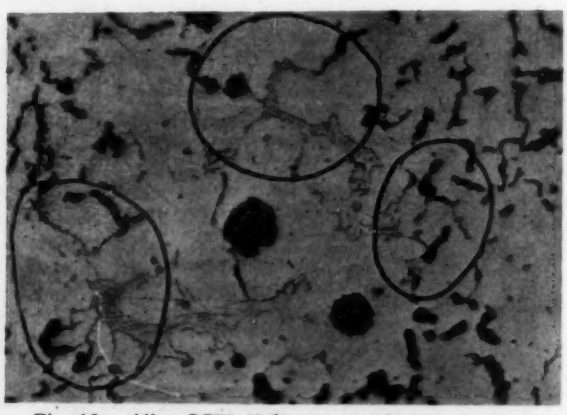


**Fig. 16 — Alloy GSW (3.54 per cent C, 2.14 per cent Si) quenched at 200 F, showing the solidification of eutectic both by growth of graphite spheroids and their austenite shells and by growth of austenite-flake graphite eutectic. Encircled areas show flake-shaped graphite growing into unsolidified melt which appears as quenched white iron structure. 2 per cent nital etch. 100 ×.**

This dual solidification process provides an excellent example of the difference between the growth of spheroids and flakes. The former are always encased in austenite and do not contact the liquid. The flakes grow with tips protruding into the melt, as shown in Fig. 16. Figure 16 shows the microstructure of a sample quenched from 2000 F. The encircled area shows flake-shaped graphite growing into the unsolidified melt, which appears as a quenched white iron structure.

As with previous samples, the results of thermal analysis (Fig. 17) and microstructural analysis were in close agreement regarding the eutectic start and eutectic end temperatures. Figure 17 also presents the temperatures at which several of the samples, which have already been discussed, were obtained.

In summary, alloy GSW, a slightly hypoeutectic ductile cast iron, solidifies as:

1. Formation of graphite nuclei between 2700 and 2600 F.
2. Development of graphite spheroids from the nuclei between 2600 and 2300 F.
3. Resolution of spheroids by the liquid metal at temperatures below 2300 F.
4. Start of eutectic solidification at about 2115 F as both flake graphite-austenite and spheroid graphite-austenite structures.
5 Completion of solidification at about 1975 F resulting in a mixed flaked graphite and spheroidal graphite structure.

## NONEUTECTIC CAST IRONS

### *Compact Graphite Irons*

Two alloys were selected to study for formation of compact graphite. Alloy CS (1.55% C, 2.33% Si) is a commercially produced, as-cast malleable iron in which the graphite is present in the compact shape. Alloy A (0.98% C, 3.52% Si) also contains compact graphite and was reported in a reference.[10] Thermal analysis cooling curves for both alloy A and alloy CS are presented in Figs. 18 and 19, respectively. These graphs also show the temperatures from which samples were obtained following the normal procedure discussed previously and the metallographically determined liquidus and solidus temperatures.

Although the composition of these two alloys are not the same, the mechanism of solidification and graphite formation in them is similar. The two alloys will, therefore, be discussed at the same time. The start of solidification in both of these alloys was accompanied by the appearance of long, thin dendrites in the quenched liquid structure. These dendrites grew in size as the temperature decreased until all of the liquid was consumed. The completely solidified structure of alloy A then consisted of only austenite. Alloy CS contained an extremely small amount of

carbides present as short films in the austenite grain boundaries after solidification was complete. The metallographically determined liquidus and solidus temperatures, respectively, were 2620 and 2120 F for alloy A and 2490 and 2050 F for alloy CS.

No graphite was formed in either of these alloys during the solidification process. During the subsequent cooling after complete solidification the few carbides present in alloy CS decomposed to form small, compact graphite shapes at temperatures below 2000 F. The number and size of these compacts increased as the temperaure was decreased. In alloy A, compact graphite was nucleated at temperatures below 1475 F. This alloy contained less graphite than alloy CS because of its lower carbon content. An example of the type of structure formed in these alloys is shown in Fig. 20. This structure is of alloy CS cooled to room temperature and contains compact graphite in a ferrite plus pearlite matrix.

The presence of compact graphite in cast irons having a carbon content less than the limit soluble in austenite, but which contain at the most only a small amount of carbide after complete solidification, has been discussed previously in a reference,[10] and is shown schematically in Fig. 21. In that work, 4 in. x 4 in. x 8 in. castings were produced of various carbon and silicon contents and were examined metallographically. Compact graphite, as shown in Fig. 22, was noted in samples having carbon and silicon contents just below the maximum solubility of carbon in austenite alloyed with silicon $-\%C + \frac{1}{6}\%Si = 2.0$. Increasing the carbon content caused graphite to appear as long thin films in the microstructure, Fig. 23. Slightly higher carbon contents produced the structure, shown in Fig. 24, where the graphite films have broken down into a transition type of graphite. Eutectic type graphite, Fig. 25, was obtained with slightly higher carbon contents.

### *Graphite Shape Occurrence*

The occurrence of the graphite shapes just mentioned may be discussed in the light of the results

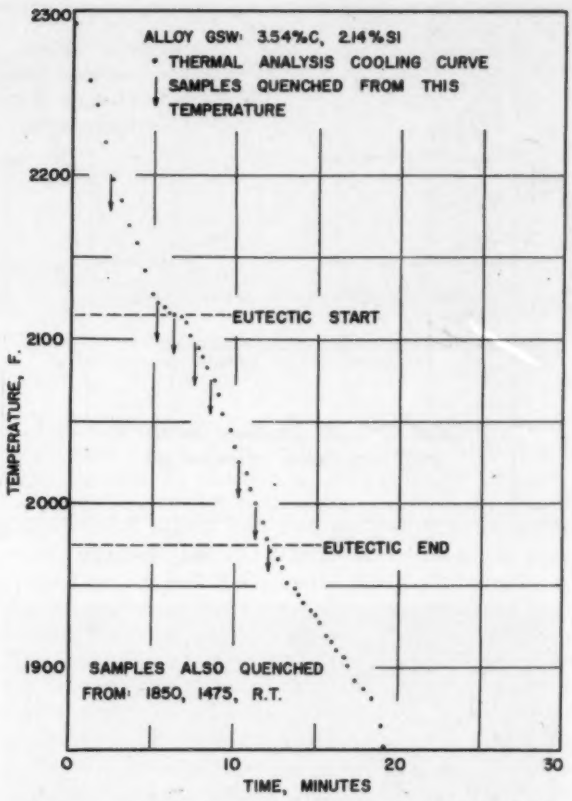



Fig. 17 — Cooling curve for hypoeutectic ductile iron GSW.

obtained from the studies of alloys A and CS. The compact graphite, presented in Fig. 22, was formed in the same manner as that found in alloys A and CS. That is, compact graphite was formed in the solid state by the precipitation of graphite from austenite as the temperature decreased.

With slightly higher carbon contents, a small amount of eutectic is formed. This occurs in the

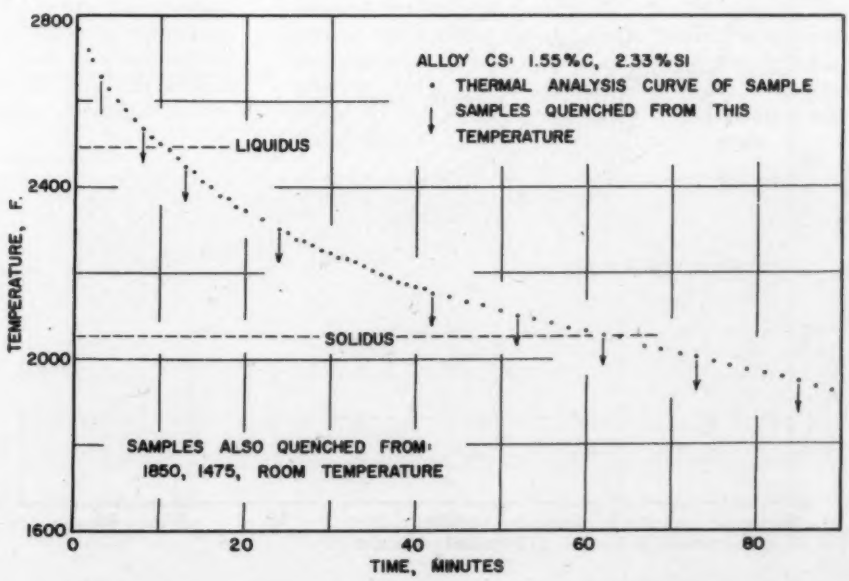



Fig. 18 — Cooling curve for non-eutectic cast iron alloy CS.

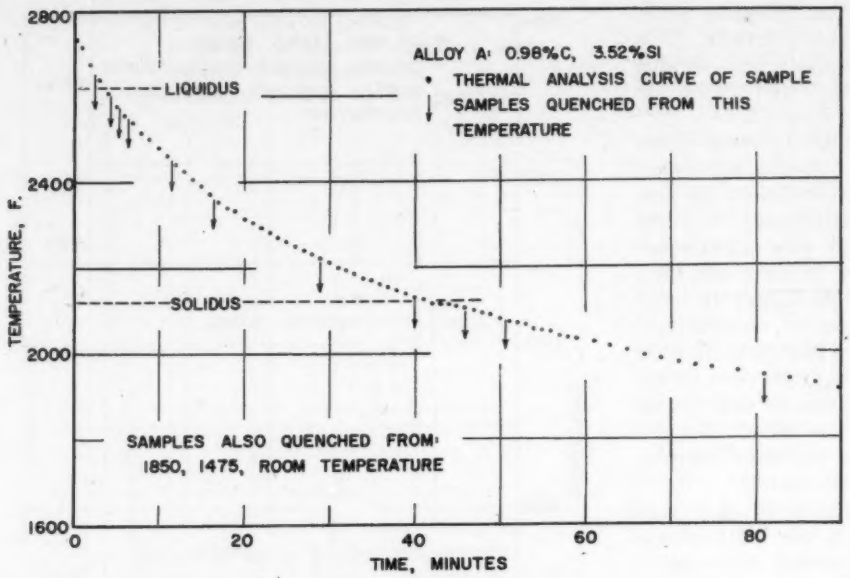



Fig. 19 — Cooling curve for noneutectic cast iron alloy A.

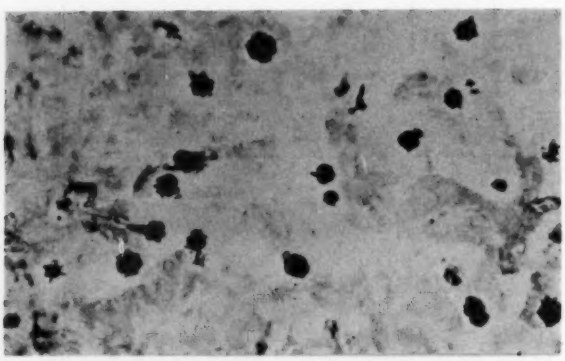


Fig. 20 — Alloy CS (1.55 per cent C, 2.33 per cent Si) containing compact graphite. This sample was furnace cooled to room temperature from 2700 F. The graphite is nucleated and grows in the solid state. 2 per cent nital etch. 100 ×.

cast iron as divorced carbide, and is thus present as films in the austenite grain boundaries. With subsequent cooling of the alloy, these films decompose to graphite, and produce a structure as shown in Fig. 23.

Still higher carbon contents result in the formation of more eutectic in the alloy, and cause the graphite structure, shown in Fig. 24, where the lacy graphite present occurs in areas where the amount of eutectic present between the austenite grains was greater than in other areas, resulting in films. An example of this structure would be at the juncture of austenite grain boundaries. The eutectic type graphite, shown in Fig. 25, is produced at slightly higher carbon contents. This graphite also appears to be the result of the decomposition of carbide.

In summary, the solidification and subsequent graphite formation of alloys A and CS, which contained compact graphite, consisted of the steps:

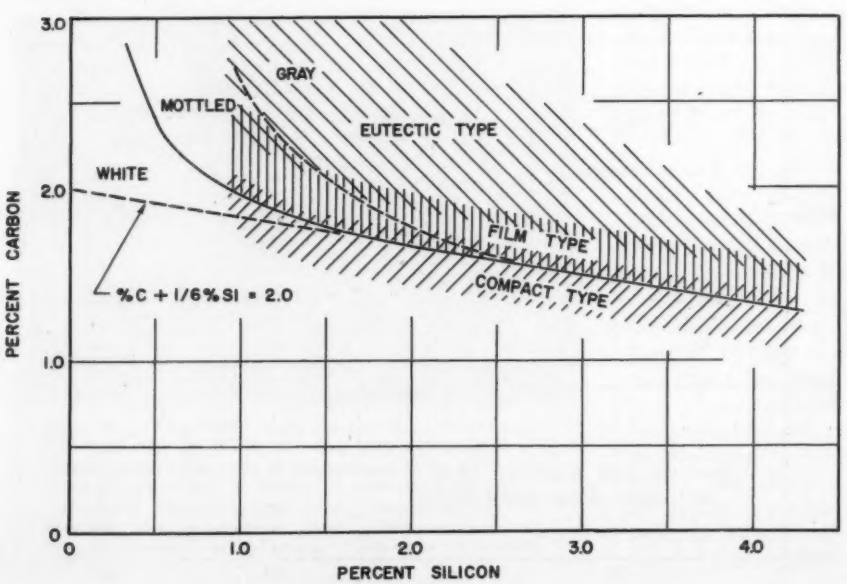



Fig. 21 — Relationship of graphite type to combinations of carbon and silicon percentages approaching the solubility limit of carbon in austenite. The film and compact types of graphite occur in essentially noneutectic compositions.

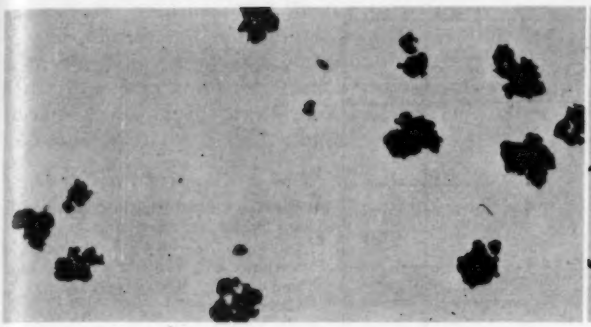


Fig. 22 — Compact graphite of the type occurring at compositions in the area so labelled in Fig. 21.

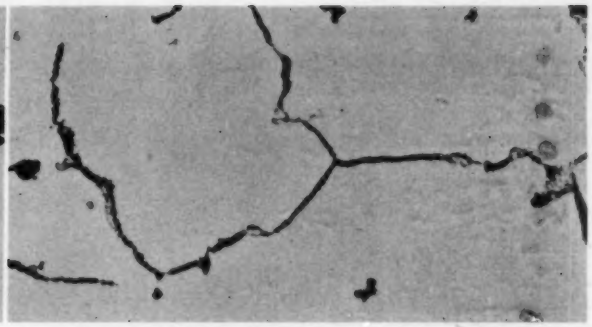


Fig. 23 — Film type graphite occurring in compositions in the area so labelled in Fig. 21.

1. Solidification of the first solid phase at about 2620 F for alloy A and about 2490 F for alloy CS.
2. Completion of solidification at about 2120 F for alloy A and 2050 F for alloy CS. Alloy A was entirely austenite at this temperature while alloy CS contained austenite plus a small amount of divorced carbides.
3. Formation of compact graphite with decreasing temperatures as a result of decomposition of the carbides present and precipitation of graphite from the austenite solid solution.

The formation of films of graphite occurs in cast irons of slightly higher carbon content as a result of the decomposition during cooling of long, thin films of divorced carbides.

Some of the circumstances which prevent the formation of graphite during solidification will now be considered. This is necessary for improved understanding of graphitization during freezing.

Previous work[10] has shown base line curves of limiting carbon and silicon percentages which separate white from mottled and gray fractures. These limiting composition curves hold true under the standard melting conditions described.[10] Figure 26 shows these compositions for a 4 in. x 4 in. x 8 in. block casting. Similar curves for other size castings are given.[10] Further, elements such as tellurium, bismuth, thallium and cerium (and magnesium) are able to shift these curves to higher levels of carbon and silicon when small percentages are added to the iron.

For example, in Fig. 26 the carbon-silicon curve separating white from mottled iron is shown for 0.01 per cent Te addition, as well as for no addition. From these curves it is seen that a cast iron containing 2.60 per cent C and 1.00 per cent Si (also about 0.08% S, 0.05% P and 0.30% Mn) will freeze gray normally, but will freeze white if 0.01 per cent Te is added to the melt. This provides an interesting case where the same base composition may be caused to freeze gray or white without altering cooling rate of the casting. Cooling curves of the casting made in these two cases are shown in Fig. 27. These curves show the expected depression in eutectic temperature range for the white iron as compared with the gray iron of the same composition. The temperatures are 2060 F for the gray iron and 2052 F for the white iron.

Between 0 and 0.01 per cent Te addition, the magnitude of the addition determines whether the casting will freeze gray, mottled or white. From 0 to 0.004 per cent Te addition the casting fracture becomes progressively more white. At 0.004 per cent Te only a few gray mottle specks remain, as illustrated in Fig. 28. These mottle specks are individual eutectic cells (of type A or type E graphite), and their number steadily decreases with increasing Te addition. Finally they disappear at 0.005 per cent addition and the casting is white. Tellurium thus poisons the sites of eutectic cell nucleation. As this occurs, the cooling curve of the casting is progressively altered from that of gray iron to that of white iron, as in Fig. 27. The preceding observations also apply to the effect by bismuth[10] and cerium and magnesium.[48,32]

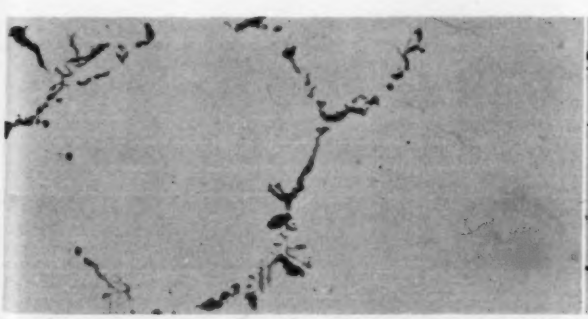


Fig. 24 — Film and eutectic transition graphite in the over lapping areas labelled film and eutectic graphite in Fig. 21.

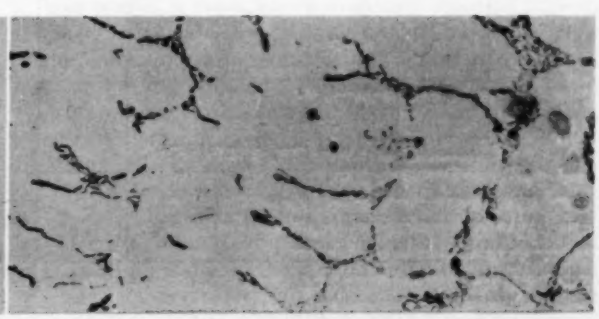


Fig. 25 — Eutectic graphite at carbon percentages slightly above that of Fig. 24, and as related to Fig. 21.

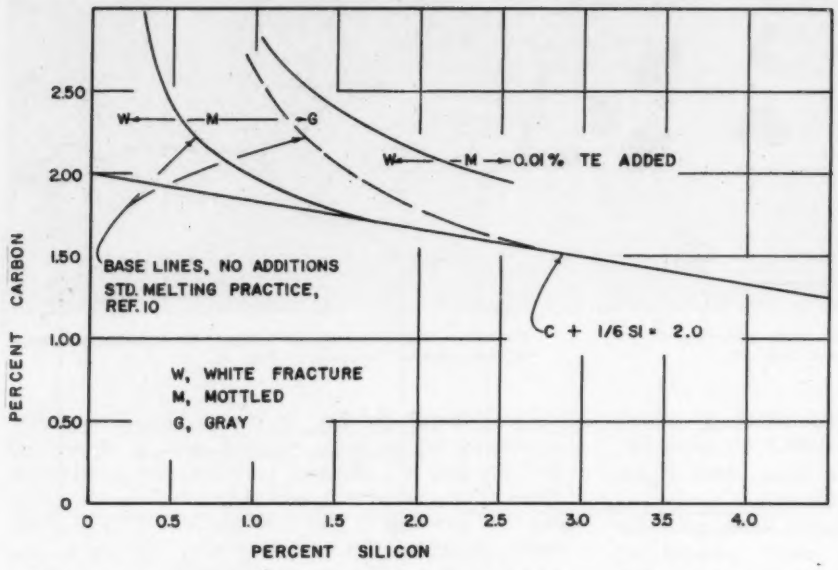



Fig. 26 — Limiting compositions for white, mottled and gray fractures in a 4 x 4 x 8 in. casting without any additions to the melt and with 0.01 per cent Te added.[10]

If the silicon content (or carbon) is now raised above the limit for 0.01 per cent Te addition, a new type of graphitization occurs and a new type of mottling develops (Fig. 29). No eutectic cell specks are visible, and instead the white appearing fracture is flecked with irregular patches of light gray. The cooling curve remains as for white iron. However, the graphite form in the mottle is no longer type A or E, but is as shown in Fig. 30. This graphite is described as lacy, and is frequently intermingled with carbide. Often there is a film of graphite at the carbide interface, as shown in Figs. 30 and 31.

In some areas, type D graphite will be observed, while in other areas what appears to be clusters of flakes exist, as shown in Fig. 31. These types of graphite are the result of poisoning of the normal eutectic cell nuclei, and thus the prevention of graphitization, until the final temperature and moment of solidification or immediately thereafter. This is the reason the cooling curve appears to be the same as that for a white iron. Since the graphitization process appears to be due to the decomposition of carbide, increasing the silicon content of an iron, in which the eutectic cell nuclei are poisoned, is especially effective in producing lacy graphite (and type D) and the irregular mottle of Fig. 29.

As stated previously, bismuth, thallium, cerium and magnesium are elements that have the above effects in varying degree. The general hypothesis for the effectiveness of this class of elements has been advanced.[10] This will now be related to the role of magnesium in ductile irons.

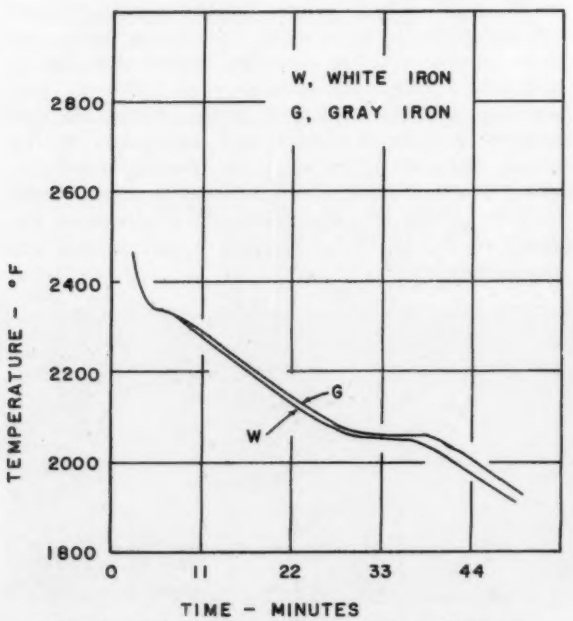



Fig. 27 — Cooling curves for a 4 x 4 x 8 in. block cast iron at 1.0 per cent Si, 2.60 per cent C and cast white at the same analysis but with 0.01 per cent Te added.

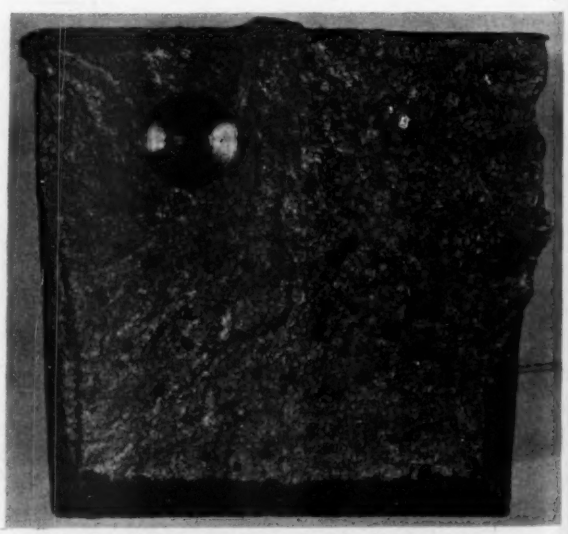


Fig. 28 — Mottle specks in a white fracture when not quite enough Te is added to cause the casting to freeze white.

Since the production of ductile cast iron is based on the addition of magnesium to a low sulfur cast iron, the role of magnesium should be considered with respect to the soldification mechanisms of the various graphite shapes. Magnesium serves a threefold capacity in ductile cast irons:

1. As a desulfurizer.
2. As a means of providing for graphite nucleation and a wide eutectic freezing range.
3. As a poisoner of eutectic cells.

The fact that magnesium forms sulfides and is a desulfurizer is well established.

In the early stages of the solidification of either hypo or hypereutectic ductile cast irons, graphite nuclei are developed which in turn develop into graphite spheroids. These graphite nuclei are not found in untreated cast irons, but are evident in all ductile cast irons at temperatures below about 2500 F and well above the eutectic solidification range. The role of magnesium in causing these graphite nuclei to form is therefore evident, even though the mechanism has not positively been established. One mechanism which has been proposed is that the nuclei generate within the first carbide to freeze,[33] and therefore are not in contact with the melt.

As the carbide decomposes, a shell of austenite develops which continues to protect the graphite from the melt. The graphite nuclei develop into spheroids of graphite encased in austenite as the eutectic start temperature is approached. If a sufficient number is present at the start of eutectic solidification, the eutectic will solidify by the growth of these spheroids and their accompanying austenite shell. However, if an insufficient number of spheroids is stable at the eutectic start temperature, then the balance of the eutectic will either freeze white or flake graphite will be nucleated, and the eutectic will solidify at least in part as a flake graphite-austenite eutectic.

In order to obtain a completely spherulitic graphite cast iron, it is necessary to nucleate a sufficient number of spheroids and to prevent the formation of other graphite types during the freezing of the eutectic. This is less of a problem in hypereutectic cast irons than hypoeutectic cast irons. In the hypereutectic irons many spheroids can be nucleated, because of both the high carbon content and the temperature range available for nucleation from 2500 F down to just above the eutectic start temperature.

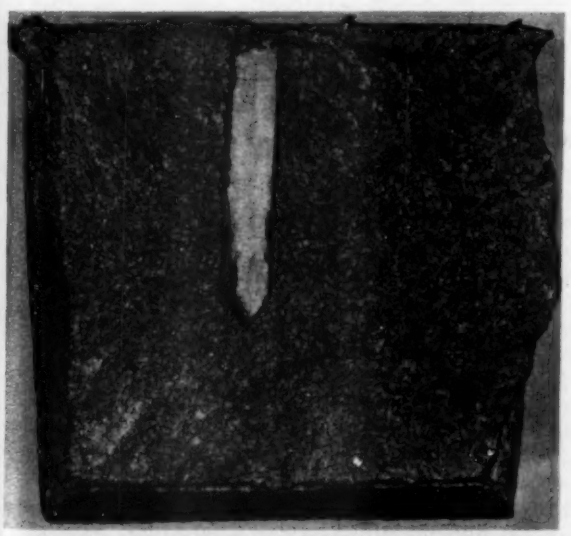


Fig. 29 — A mottled fracture showing irregular patches of gray and white. Compare with Fig. 28.

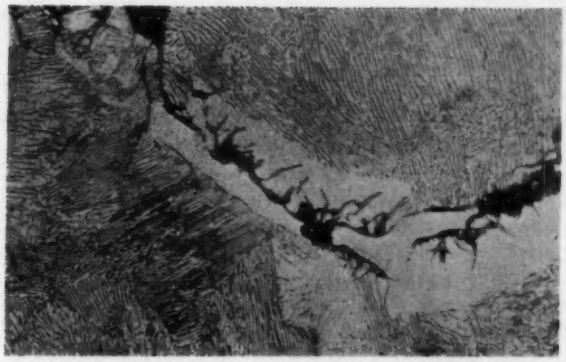


Fig. 30 — Lacy graphite mixed with carbide, produced in mottled iron with the fracture shown in Fig. 28 and composition just above the limit curve for 0.01 per cent Te added shown in Fig. 26. 2 per cent nital etch. 250 ×.

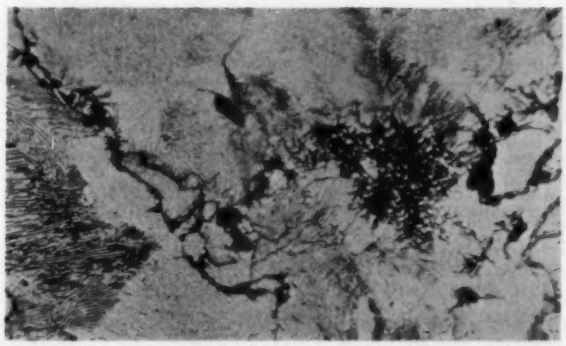


Fig. 31 — Lacy graphite at carbide interfaces and a cluster of flakes. 2 per cent nital etch. 500 ×.

### *Hypoeutectic Cast Irons*

Hypoeutectic cast irons have difficulty providing enough graphite nuclei because the carbon content is lower, and those initially nucleated at temperatures above the eutectic start are not stable and may dissolve above and in the eutectic range. In both carbon ranges, it is necessary in ductile irons to prevent eutectic cell or flake graphite nucleation at or below the eutectic start temperature. This provides the need for the third role of magnesium, which is to prevent the formation of flake graphite in the eutectic freezing range by functioning as a poisoner of eutectic cell or flake graphite nucleation.

In this role, the magnesium functions in the same way as tellurium, bismuth and cerium in the prevention of mottle, as discussed earlier. If insufficient magnesium is present, spheroids may nucleate at the higher temperatures, but eutectic flakes will also be nucleated below 2100 F and mixed graphite types will result in the solidified iron. If the minimum percentage of magnesium is not present, the freezing

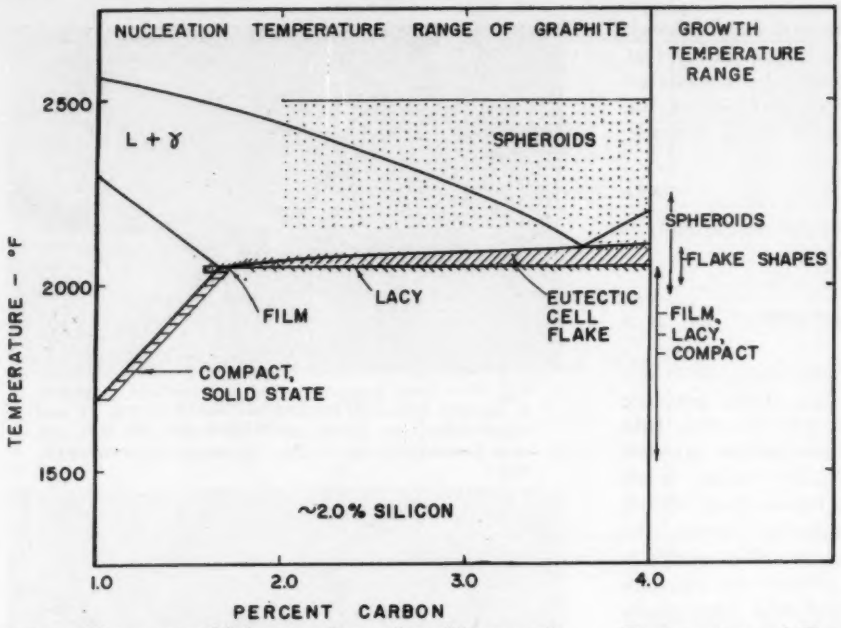



Fig. 32 — Schematic representation of temperature range for nucleation and growth of spheroidal, flake, compact, film and lacy graphite, as related to carbon content at approximately 2.0 per cent Si. Graph is actually semi-quantitative for slowly cooled iron.

range will not start high enough nor be wide enough to nucleate and grow spheroids. The evidence for magnesium raising and extending the eutectic freezing range consists of the thermal analysis data of this report, and the fact that a divorced iron carbide eutectic develops in iron-carbon-magnesium alloys.[27,33]

When an excess of magnesium is present, the tendency is for the final eutectic solidification to proceed as white iron. However, this depends on the silicon percentage present in the iron. It is too high for the magnesium content of the iron, lacy graphite may develop late or at the end of freezing according to the principle discussed previously. Such graphite has been shown to occur in commercial ductile irons, and is reported in the literature a number of times and will therefore not be illustrated here.

In summary, then, magnesium in ductile cast irons functions in at least three ways. First, it reacts with sulfur to permit the second two functions to be fulfilled. Second, it extends the freezing range and permits graphite spheroids to nucleate at high temperatures and grow during cooling. Finally, it inhibits the nucleation of flake graphite in the eutectic solidification range and thereby permits fully spheroidal structures to be obtained.

## DISCUSSION

The forms of graphite studied by the authors include spheroidal, eutectic flake, compact, film and lacy graphite. A general theory for the formation of these graphite types can be offered in the light of this and related research. Of primary significance in this theory is the temperature range of nucleation and growth of the various graphite forms. Secondary to the temperature range for nucleation and growth, are the circumstances of chemical composition of the base iron and additions to it which control or modify this temperature range. A semi-quantitative representation of the relationship of nucleation and growth temperature range to composition and temperature for the various graphite forms is shown in Fig. 32.

### *Temperature of Nucleation and Growth*

With reference to Fig. 32, the nucleation and growth of flake graphite as eutectic cells of ($\gamma$+ graphite) has been shown to occur in the eutectic solidification range (generally 2100 to 2000 F) by many investigators, as pointed out previously. The authors found agreement with this view in their work. However, this study has not been concerned with the details of the number of eutectic cells nucleated or whether the graphite flakes are type A, B or E. Rather, it has been concerned with the mechanism of growth, which is as eutectic cells of simultaneously solidifying $\gamma$+ graphite.

The mechanism involves an interface of solid $\gamma$ and graphite, with the tip of the graphite flake advancing into the melt from the cell nucleus. Again the authors have not been concerned with the question of whether multiplication of the graphite flakes occurs by branching of graphite flakes, or is initially determined by the number of flakes radiating from the cell nucleus. The essential point concerns the fact that the cells and cell flakes are nucleated and grow over the narrow temperature range of 2100 to 2000 F (or 2100 F to 2040 F in the 2.0-2.40 per cent Si range for type A graphite). Also, the flake grows with its tip in contact with the melt. No graphite flakes or flake nuclei are noted in the melt above the upper eutectic temperature, except in the case of kish in hypereutectic irons.

By contrast, graphite spheroids are nucleated, as shown in Fig. 32, at temperatures of about 2450 to 2500 F down to about 2100 F in both hypo and hypereutectic ductile irons. This is reconcilable in hypereutectic irons but not in hypoeutectic irons where only $\gamma$ should be present. Nevertheless, this observation was made experimentally. Even in hypereutectic irons, the spheroids were noted to first nucleate at 2400 to 2500 F at the same composition where kish would nucleate at 2200-2250 F. Because of this, it appears a fundamental principle that spheroids

must be nucleated at high temperatures where they are initially stable whether in contact with the melt or not.

A second principle appears to be that graphite nuclei in contact with the melt will grow as flakes, once a temperature between 2200 to 2100 F and lower is reached during cooling. Since spheroids are encased in austenite at these temperatures (or in carbide) they are not in contact with the melt, and may continue to grow as spheroids. Kish nuclei in hypereutectic irons are not encased, and may therefore grow as flakes.

### *Spheroid Growth*

Actual growth of spheroids and the surrounding $\gamma$ shell begins above the eutectic temperature, is indicated in Fig. 32, and continues rapidly early in the eutectic range, and gradually diminishes as the end of the eutectic freezing is reached at about 1985 F. Growth occurs by adding $\gamma$ to the shell, and by carbon diffusing through the shell to the graphite sphere. This mechanism of growth is quite different from eutectic cell growth. It is in this mechanism that the added element, such as magnesium or cerium, is helpful by its effect of extending the freezing range of the eutectic. The $\gamma$ shell is caused to start growing at higher temperatures by this extended freezing range. In addition, the freezing of the eutectic is extended to lower temperatures because the carbide (or graphite dissolved in the melt) is divorced from the $\gamma$, depositing on the $\gamma$ shell surrounding the spheroids.

According to the theory under discussion, these graphite spheroids are nucleated and grow at high temperatures, and are prevented from becoming flakes at lower temperatures by encasement in $\gamma$. The boundary temperature is about 2100 F. Below this temperature, graphite nuclei in contact with melt grow as flakes. It follows that if the $\gamma$ shell encasing the spheroids is ruptured below this temperature, graphite growth then continues by the flake process. This is observed in the exploded graphite emanating from spheroids in commercial ductile irons.

If graphite spheroids are nucleated as a general condition at high temperature, it would also be expected that irons not treated with magnesium or cerium would under some circumstances (quenching, for example) show graphite spheroids retained in the quenched structure. This has also been reported.[15] Also, if spheroids are nucleated at high temperatures in hypoeutectic irons, the graphite should appear in the hypoeutectic dendrites rather than in the eutectic carbide (or flakes). This is also reported.[15]

Further, in ductile irons, it should be necessary to have present an element that will poison eutectic cell nucleation even after graphite spheroids are nucleated. Magnesium performs this function, as does cerium. Bismuth and tellurium can also be used in ductile iron for this purpose. But those elements, such as Te and Bi, which poison eutectic cell nucleation may permit some graphite spheroids nucleated at higher temperature to grow and appear in the solidified iron even without magnesium present. This has been reported,[32] and has been observed by the authors in irons where sulfur has not been below 0.06 per cent and no magnesium was added.

In summary, it is seen that the experimental data plus related observations in a wide variety of cast irons, support the principle that graphite spheroids nucleate and initially grow at higher temperatures than flake graphite, i.e., it is a temperature-dependent phenomenon. This fits the same principle of crystallization of other crystals, such as ice or snow crystals whose shape and growth pattern varies with temperature. With reference to graphite growth, above a temperature of about 2100 F nucleation occurs, and growth is uniform in rate on all axes of the nucleus resulting in a spheroidal form. Below this temperature, nucleation occurs and growth proceeds rapidly on one crystal axis, resulting in flake shapes.

### *Nucleation and Growth Temperature*

Relative to temperature of nucleation and growth, compact graphite has been shown to occur in noneutectic cast irons where no eutectic is present during solidification, because of the low percentage of carbon in the iron. These compositions are located in Fig. 32. The nucleation of graphite occurs when the metal is in the solid state by precipitation from $\gamma$. Growth occurs by diffusion of carbon to the nucleus and precipitation thereon. This requires a gradually decreasing temperature between the solidus and the $A_1$ transformation temperature of the iron. The term compact graphite is used to describe this form of graphite, and to indicate that it is formed during cooling of the iron from the casting operation. Temper carbon in malleable iron requires reheating after cooling for its formation, and this differs in sequence of formation from compact graphite.

Film and lacy graphite form under conditions which have been amply discussed. However, in connection with temperature of nucleation and growth, as shown on Fig. 32, it should be recognized that this occurs in the lowest temperature of eutectic solidification or by decomposition of carbide just after solidification. This, then, is also a low temperature form of graphite nucleation and growth.

## SUMMARY

By combining accurate thermal analysis techniques with quenching of small samples, the authors have studied the freezing mechanism of cast irons containing spheroidal, flake, compact, film and lacy graphite. The work by earlier investigators was confirmed, regarding the mechanism of flake graphite formation by eutectic cell nucleation and solidification in the range of 2100 F and down. Graphite spheroids were shown to nucleate at temperatures of 2400 to 2500 F and to begin growth most significantly in the range of 2200 and down. The mechanism of growth involves solidification of austenite from the eutectic liquid into the austenite shell encasing the graphite nucleus.

The nucleus grows by diffusion of carbon through the $\gamma$ shell. The mechanism was described for hypoeutectic and hypereutectic compositions. The formation of compact graphite was shown to be a solid state graphitization process in low carbon non-eutec-

tic compositions. Film and lacy graphite were shown to be special cases of low temperature graphitization at temperatures of the end of eutectic solidification or just below, and of special compositions such that no eutectic cells are nucleated. Finally, a general hypothesis has been advanced and supported to the effect that the form of graphite nucleated and grown is primarily temperature dependent with other factors of iron composition and additional elements, functioning primarily through their ability to influence the temperature of nucleation and circumstance of growth.

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# NEW HIGH STRENGTH DEZINCIFICATION RESISTANT BRONZE CASTING ALLOY

by A. H. Hesse


## ABSTRACT

**Foundry characteristics including pouring temperature, sand, melting practice, gating and risering of a new dezincification resistant alloy are presented. Also included are data on the alloy's chemistry, mechanical properties, machinability, corrosion resistance and electrical conductivity. Applications are suggested. A method for production of internally sound value stem castings is also presented.**


## OBJECT

Specification 8959 for Double Disc Gate Valves in paragraph 2.7 Bronze Catalogs material for valve stems states:

> "a. Valve stems shall be cast or forged from bronze having a tensile strength of not less than 60,000 pounds per square inch, a yield point of not less than 30,000 pounds per square inch, and an elongation of not less than 12 per cent in two inches.
>
> "c. Bronze for all interior parts of valves shall contain not more than 2 per cent aluminum nor more than 7 per cent zinc. Heat treatment may be used to develop the desired strength characteristics."

The Department of Water and Power of the City of Los Angeles in these two paragraphs definitely limits the bronzes which can be used for valve stems to only a few.

Of the few qualified alloys, each has an undesirable characteristic which most consumers would prefer that the material did not possess. For instance, manganese bronze which foundrymen have learned to handle is not desired because it dezincifies; aluminum bronze which foundrymen have learned to tolerate dealuminumizes; silicon bronze does not have enough as-cast strength; silicon brass has too much zinc, and not enough yield strength; wrought silicon bronze must be work hardened to meet mechanical properties; and nickel-tin bronze must have practically no lead in order to meet mechanical properties by heat treatment.

Although manganese bronze, aluminum bronze, silicon bronze and silicon brass do not satisfy in one way or another the City of Los Angeles, these alloys have not been completely outlawed as a valve stem material in other locations. In fact, their use is still widespread in many other parts of the country, where they apparently are putting up an above average performance.

However, the City of Los Angeles has seemingly set a precedent for a valve stem material which must be more than satisfactory. This precedent is being adopted by others, one of which that can be mentioned at this writing is the City of Oklahoma City, Okla.

## INTRODUCTION

Manganese bronze is a high strength casting alloy with suitable machining qualities that dezincifies under certain corrosive conditions. A noteworthy instance is found when this material is used as a valve stem. Depletion of zinc leaves the stem as a spongy coppery mass which sooner or later fails in service because of its weakened condition. Aside from its resistance to corrosion, manganese bronzes (A.S.T.M. B-132-52, Alloy A or B) are selected because they combine high tensile strengths (60,000 and 80,000 psi, min.), good yield strengths (20,000 and 32,000 psi, min.) and good elongation (15 per cent elongation in 2 in., minimum for both alloys). Both alloys have good machining qualities.

These alloys are still satisfactory valve stem materials. Certain valve consumers have written them out of their specifications because of their behavior in waters with higher pH than 9, or with a specific conductance or more than 350K x $10^6$ micromhos per centimeter as determined in A.S.T.M. D1125-50T, as such water will remove zinc from bronze.[3] To retard dezincification in manganese bronze certain additives as 0.5 per cent tin have been found effective, though presumably not to a satisfactory degree in view of the pressure brought to bear by some consumers for a better material. For a time it was felt that aluminum bronze could be an alternative, or more suitable material. Reasoning behind this assumption was that mechanical properties approximately equal to manganese bronze were available in aluminum bronze in the as-cast condition. Moreover, it was felt that there would be no dezincification since no zinc was present in aluminum bronze.

This proved to be true except that a new corrosion phenomenon presented itself. It promptly was termed dealiminumization, and rightly so. This type of corrosion apparently parallels dezincification in

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manganese bronze, in that aluminum is depleted from the alloy. The corrosive media work their way through the material, gradually lowering its strength until failure results. Bearing in mind the chemical and mechanical requirements of the Department of Water and Power of the City of Los Angeles, such commercially available alloys as silicon bronze, silicon brass and a relatively new proprietary alloy are eliminated because of inability to meet all requirements.

For instance, silicon bronze (A.S.T.M. B-198-52, 12-A) meets chemical specifications but falls short on tensile and yield strength. Silicon brass (A.S.T.M. B-198-52, 13-B) fails on two counts. It contains more than 7 per cent zinc and falls short on yield strength, yet it is used successfully as a stem material in many valves.

The relatively new and patented alloy, also fails on one count—chemical composition. It contains more than 2 per cent aluminum and is restricted to 0.03 per cent lead in castings. The latter presents a problem of contamination to most foundries. However, it more than meets the minimum mechanical requirements.

### *Acceptable Materials*

The materials which have been successful in meeting the City of Los Angeles requirements on all counts are wrought silicon bronze (A.S.T.M. B-98-58; Alloy D) and heat-treated nickel-tin bronze (A.S.T.M. B-292-56; Alloy A). One exception to the latter alloy must be mentioned. According to A.S.T.M. the heat treated nickel-tin bronze has 5 per cent elongation in 2 in. minimum. It is assumed that in actual practice the heat treated material has higher elongation, because the City of Los Angeles, where this alloy is accepted, requires 12 per cent elongation in 2 in. minimum.

Silicon bronze meets chemical specifications without qualifications, but must be upset and finished half-hard in order to satisfy mechanical properties. Upset collars are used to reduce machining time on valve stems. Some prefer to purchase half-hard rod with a diameter equal to or slightly greater than the O.D. of the collar. Machine shop scrap and machining costs are quite high for such stems considering the amount of stock required to meet O.D. of the collar. Nickel-tin bronze has been the only known commercial casting alloy which seems to satisfy the City of Los Angeles. It meets the chemical requirements and with a solution treatment followed by a precipitation heat treatment, meets the mechanical properties with the possible exception of elongation. Principal drawbacks to this alloy stem from the foundryman who shies away from heat treatment because he is usually not so equipped, and because it increases the cost of his product substantially. Moreover, the lead in the alloy must be less than 0.01 per cent in order to obtain a satisfactory response to heat treatment.

This also plagues the foundrymen who must procure sufficiently pure base metals and carefully segregate returns to the melting room to avoid contamination. Production volume on this type of material is usually small, therefore chances for contamination with lead are always present, since most brass foundries handle bronzes containing substantial quantities of lead. More than 0.01 per cent lead can be picked up by melting nickel-tin bronze in a crucible in which 85-5-5-5 had been melted previously.

In reviewing the characteristics of the alloys which satisfied the City of Los Angeles, it appeared that a market for a new alloy would be readily available, providing some of the shortcomings of the aforementioned materials could be overcome. Such improvements would incorporate:

1. Develop a casting alloy which would satisfy the City of Los Angeles in all areas without heat treatment.
2. Develop a casting alloy which would not be sensitive to contamination by reasonable amounts of lead, e.g., 0.25 per cent.

These objectives have been accomplished by the development of an iron-nickel-aluminum-silicon-zinc bronze. When prepared within the preferred chemical limits, the alloy meets the minimum mechancial requirements of the City of Los Angeles for stems in double disc gate valves. The chemistry and mechanical properties satisfy the City of Los Angeles without the need for heat treament. The alloy is not extremely sensitive to lead contamination, and therefore is a candidate for use in the majority of nonferrous foundries since abnormal policing is not mandatory.

## MELTING PRACTICE

The alloy has been satisfactorily melted in oil and gas fired crucible furnaces and in the rocking arc type and high frequency type electric furnaces. This does not imply that other types of furnaces are unsatisfactory. It means only that melting has not been undertaken in the other types of melting equipment normally employed for copper casting alloys. Oxidizing melting conditions have been employed throughout the research with the exception of one test (Table 1). The results did not show any great difference in mechanical properties between oxidizing and reducing melting atmospheres. However, a series of tests may prove otherwise.

TABLE 1 — EFFECT OF POURING TEMPERATURE ON MECHANICAL PROPERTIES OF NEW BRONZE

| HEAT NO L44 | | | |
|---|---|---|---|
| POURING TEMP., °F | TENSILE STRENGTH, PSI | YIELD STRENGTH, PSI | ELONGATION, % IN 2" |
| OXIDIZING CONDITIONS | | | |
| 2150 | 59,500 | 26,000 | 18.4 |
| 2050 | 62,800 | 31,100 | 22.1 |
| 1950 | 69,100 | 32,300 | 28.1 |
| 1950 | 68,900 | 39,300 | 29.4 |
| | REDUCING CONDITIONS * | | |
| 2000 | 63,400 | 30,300 | 21.8 |

*AVERAGE OF TWO TESTS

A subsequent test on another heat (101-1, Table 4), in which the chemistry was satisfied, but possibly because of a gassed metal condition, the tensile strength and elongation failed to meet the minimum require-

ments. A substantial difference then exists between, say a heat prepared under oxidizing conditions (Table 1, P.T. 1950 F) and one presumably gassed (Table 4, 101-1). The beneficial effect of a dry nitrogen treatment is also demonstrated in Table 4. The nitrogen treatment amounted to a 10 min purge in a melt superheated to 2400 F. Treatment was accomplished outside the furnace in a crucible. At the completion of the treatment the temperature had dropped to 2050 F. It was then allowed to cool naturally to 1970 F at which temperature the keel block was poured.

In order to eliminate the possibility that superheating rather than nitrogen was the beneficial factor, Heat 101-1 was prepared in the exact manner as Heat 101-3, excluding only the nitrogen treatment. The use of returns, e.g., gates, risers and once and twice melted pigs was employed in amounts to and including 50 per cent of the charge. No deleterious effect on either casting quality or mechanical properties was noted. In fact, a heat prepared in the foundry of a valve manufacturer containing only gates, runners, etc., produced the following mechanical properties after a 9 min nitrogen treatment beginning at 2375 F. The keel block test bars were poured at 1950 F.

| | |
|---|---|
| Tensile Strength, psi | 68,300 |
| Yield Strength, psi | 32,600 (0.2% offset) |
| Elongation, % in 2 in. | 27.5 |
| Reduction of Area, % | 29.1 |

The precautions necessary for melting manganese bronze, aluminum bronze or silicon bronze are to be followed for this iron-nickel-aluminum-silicon-zinc bronze. There is evidence that lead can be tolerated without adversely affecting mechanical properties. Experimental work has been conducted in which 0.90 per cent lead produced these as-cast mechanical properties from a keel block test bar:

| Heat 104-41 | |
|---|---|
| Tensile Strength, psi | 64,000 |
| Yield Strength, psi* | 31,500 |
| Elongation, % in 2 in. | 24.8 |
| Brinell hardness (500 Kg load) | 93 |

*Divider Method

However, other experiments with lead additions were not so successful. Heat 102-5 containing 0.49 per cent lead had the mechanical properties as-cast of:

| | |
|---|---|
| Tensile Strength, psi | 64,500 |
| Yield Strength, psi* | 43,900 |
| Elongation, % in 2 in. | 9.4 |
| Brinell hardness (500 Kg load) | 106 |

*Divider Method

In this instance, elongation is below the 12 per cent minimum of the City of Los Angeles requirements. Therefore, as a precautionary measure, and until more is learned about the alloy, lead is being held at 0.25 per cent maximum. Research has shown that no zinc replacement or other additions are necessary to achieve quality castings even though 2400 F is attained, so long as at least a 50/50 ingot-scrap return balance is maintained for each charge.

Although there may be other ways of degassing, best results for optimum mechanical properties have been obtained to date by the use of dry nitrogen. Quality commercial castings (i.e., valve stems) have been produced without a nitrogen treatment. Commercial castings should be heated to a temperature which will provide sufficient superheat to pour without misruns. No covers or fluxes have been employed. The melt is merely skimmed prior to pouring. It has been found that the amount of dross can be considerably reduced by avoiding unnecessary skimming and/or other agitation.

## POURING TEMPERATURE

The pouring range for the alloy was determined by studying the effect of pouring temperature on mechanical properties and the minimum temperature at which certain commercial castings can be poured. The study relating to pouring temperature versus mechanical properties was conducted under oxidizing conditions in which 4 oz of cuprous oxide were placed on the bottom of the crucible, ahead of the 100 lb charge. Ten min prior to pouring, another 4 oz of cuprous oxide were added to the charge which had been melted under the oxidizing atmosphere in a gas fired furnace. The charge was allowed to cool from 2200 F without further additions to the desired pouring temperatures.

Mechanical properties are obviously affected by pouring temperature, Table 1. Optimum properties were obtained by pouring the keel block at 1950 F. Mechanical properties decreased as the pouring temperature increased above 1950 F. The bottom temperature level for keel blocks was indicated as having been reached when a keel block had to be poured from the top at 1850-1900 F in order to avoid a misrun. Properties of test specimens from the later keel blocks were 63,000 psi, tensile strength; 34,000 psi, yield strength; and 15.8 per cent elongation in 2 in.

The two 6 in. valve stems mounted to fit a 14 x 26 in. steel flask were poured through a common runner with a 1¼-in. diameter sprue and $\frac{3}{32}$ x 1 in. choke without difficulty as low as 2020 F. Three such molds were poured from one crucible beginning at 2150 F and ending at 2020 F. The total distance traveled by the metal before it finally came to rest was approximately 30 in. Aside from the fact that the choke slowed the pouring a little too much, no dfficulty was experienced in filling all corners of the mold cavity.

Eight, 4¼ x 12¼ x $\frac{7}{16}$-in. plates, were poured without misruns through a $\frac{3}{32}$ x 1 in. choke beginning at 2300 F. The total distance the metal traveled before coming to rest was approximately 24 in. In general, the pouring temperature range for this alloy seems to parallel closely the aluminum bronze family of alloys.

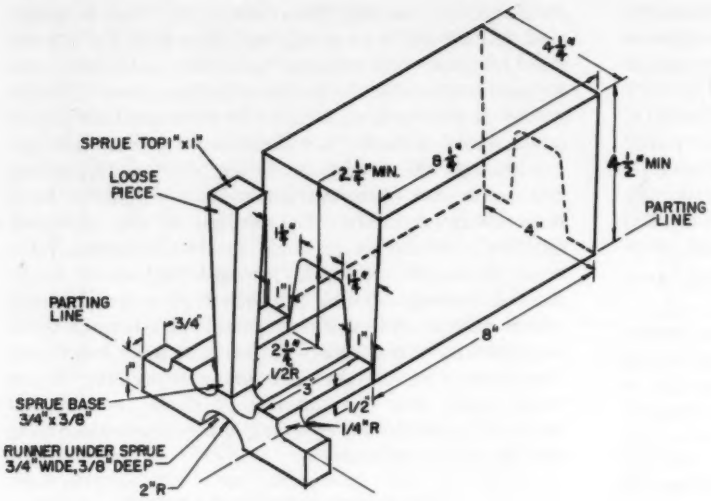



Fig. 1 — Double keel block test bar gated at the bottom.

## GATING AND RISERING

Practices employed for manganese bronze and aluminum bronze apply to the new dezincification resistant alloy. This means large risers are required to accommodate shrinkage, and agitation in mold is to be avoided by inverted horn gate, or comparable techniques, bottom feeding the casting. About 50 per cent of a mold is gates and risers. It is good practice to gate into riser attached to heaviest section of casting so that hottest metal will be in riser, thus promoting directional solidification.

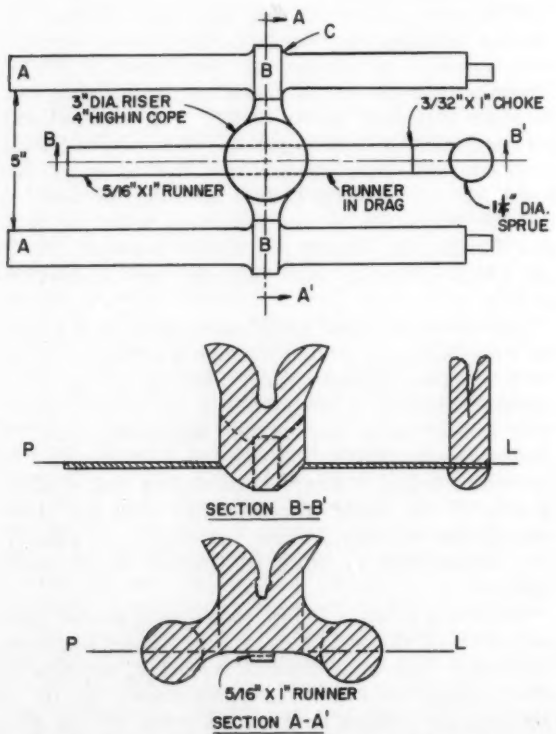



Fig. 2 — Rigging for 6 in. valve stem casting (heat 124).

Sections which cannot be reached by a riser may be chilled, if practicable. Chokes or screens may be used effectively in retaining dross in the sprue and a bypass in the runner (Fig. 2) serves to collect the dross which is formed during the first moments of the pour. Altogether the alloy presents no new problems in gating and risering. They have been encountered previously in production of manganese bronze and aluminum bronze castings.

## SAND

The alloy performs well in all types of sand, e.g., natural bonded green sand, synthetic sands, synthetic oil sands, etc. Because the alloy contains aluminum and silicon, the castings shake out clean and bright. Best results have been obtained with sand having at least a 25 permeability, while with adequate venting sands with a 15 permeability can be tolerated.

## PRODUCTION OF INTERNALLY SOUND 6 IN. VALVE STEMS

Most foundries prefer to gate valve stems at the threaded end. This is done to facilitate cutoff and grinding and to simplify gating to the extent that greater casting yields per mold are achieved. It is agreed that this is not the best procedure for maximum internal soundness because the laws of directional solidification are not followed. As shown in Fig. 7, nonrising cast valve stems have a collar about $\frac{2}{3}$ the distance from the threaded end. The heavy collar section requires some other means than the riser at the threaded end to feed it. Depending upon the size of the stem, a chill around the collar and equal to the collar section eliminates centerline shrinkage at this point. However, as the length of the stem increases, the feeding distance of the riser at the threaded end becomes less effective and a secondary pipe in the stem at some intermediate distance is encountered.

In Tables 2 and 3, typical results of properties are given which can be expected from threaded end gating and risering without feeding or chilling the collar section, as shown in Fig. 7. The low elongation

TABLE 2 — CHEMICAL COMPOSITION OF COMMERCIAL VALVE STEMS

| DESIGNATION | | | CHEMICAL COMPOSITION, % | | | | | | | | | | |
|---|---|---|---|---|---|---|---|---|---|---|---|---|---|
| LAB. NO. | COMMERCIAL | A.S.T.M. | COPPER | TIN | LEAD | IRON | NICKEL | ZINC | PHOS. | MN | SILICON | ALUM. |
| 119 | NICKEL TIN BRONZE | B292-56,A | 88.3 | 5.5 | .27 | .03 | 5.1 | .85 | .009 | | | |
| 120 | SILICON BRASS | B198-52,13B | 83.8 | | .05 | .36 | .05 | 11.9 | | .04 | 3.7 | |
| 122 | SILICON BRASS | B98-58, D | 95.2 | .02 | .27 | .08 | .05 | .20 | | 1.4 | 3.1 | |
| 123 | MN BRONZE | B147-52,8A | 59.6 | .24 | .23 | .94 | .12 | 37.4 | | .68 | .03 | .77 |

and tensile strength of the test bar (B) cut from the center of the collar section indicates lack of internal soundness in that area. While the test bar (A) cut from the threaded end has about 50 per cent greater tensile strength and yield strength and five times the elongation of test bar (B). This is undoubtedly attributed to the difference between a well fed and poorly fed section.

In either case, it is to be noted that the results are below A.S.T.M. minimums. This is not an alarming fact, since it has been proved many times that test bars cut from castings rarely, if ever, meet specifications, while from the same heat, test bars machined from accepted test bar designs will meet requirements. Apparently lack of strength in the core of the stem does not impair its serviceability. Design factors have compensated for the discrepancy between test bars and cast stems.

In spite of these accepted differences which have been lived with for years, and which have not been the cause for casting rejections, it was decided to see what had to be done to 6 in. valve stem castings to produce the same mechanical properties in the castings that are obtained in the test bar. As a starting point, four valve stems were taken from stock valves. From each stem a 0.505 in. tensile specimen was machined at the threaded end (A) and the collar section (B). The chemical compositions of these stems are shown in Table 2. The mechanical properties including A.S.T.M. minimum specifications are given in Table 3.

### *Sand Cast Valve Stems*

Two of the commercial stems were sand cast. The upset bar stock stem (Lab. 122) was produced from half hard silicon bronze rod with the collar formed by an upsetting process. This produces a stem whose mechanical properties are enhanced by cold work. The chill cast stem (Lab. 123) was produced by pouring directly into the riser of a solid iron mold held vertically with the threaded end up. Properties of such castings are enhanced by the chilling effect of the iron mold.

Only the chemical requirements of Specification 9859 were met by Lab. 119 and 122. Only the end section (A) of Lab. 122 met mechanical property minimums. Only the end section (A) of Lab. 123 which did not meet chemicals passed the minumum mechanical requirements. The collar section (B) of Lab. 123 had a large shrink, and therefore no tensile test could be made.

Keeping in mind the requirements of the City of Los Angeles, none of these stems are acceptable, mechanically. Yet, these same specifications do not ask for said properties in the stem. They merely require that test bars be made for each heat of metal. Therefore, there is every reason to believe that had test bars been supplied with the four stems, all but one would have met the A.S.T.M. requirements for the alloy. Laboratory 119 specimen would not have responded to heat treatment because the lead content exceeds 0.01 per cent. However, as previously mentioned, none of the test bars except Lab. 122 would

TABLE 3 — MECHANICAL PROPERTIES OF COMMERCIAL VALVE STEMS

| LAB. NO. | REMARKS | TEST BAR LOCATION IN STEM * | TENSILE STRENGTH, PSI | YIELD ** STRENGTH, PSI | ELONGATION % IN 2" | REMARKS |
|---|---|---|---|---|---|---|
| 119 | SAND CAST | A | 34,100 | 26,700 | 7.0 | (NI VEE) |
| | | B | 33,900 | 26,800 | 6.0 | |
| | | A.S.T.M. MINIMUMS | 75,000 | 50,000 | 5.0(H.T.) | *** |
| 120 | SAND CAST | A | 46,300 | 34800 | 10.0 | (TOMBASIL) |
| | | B | 31,100 | 24,200 | 2.0 | |
| | | AS.T.M.MINIMUMS | 60,000 | 24,000 | 16.0(A.C.) | |
| 122 | UPSET BAR STOCK | A | 92,700 | 72,200 | 18.0 | (EVERDUR) |
| | | B | 49,200 | 19,300 | 29.5 | |
| | | A.S.T.M.MINIMUMS | 70,000 | 38,000 | 17.0(H.H.) | |
| 123 | CHILL CAST | A | 65,900 | | 14.5 | (MN BRONZE) |
| | | B | NO TEST BECAUSE OF LARGE SHRINK AT CENTER OF REDUCED SECTION OF TEST BAR | | | |
| | | A.S.T.M.MINIMUMS | 65,000 | 25,000 | 20.0(A.C.) | |

* SEE FIG.-2
** .005 IN./IN. EXTENSION UNDERLOAD
*** A.S.T.M. MINIMUMS BASED ON STANDARD PROCEDURES FOR TEST BARS

TABLE 4 — EFFECT OF DRY NITROGEN ON MECHANICAL PROPERTIES OF NEW ALLOY

| HEAT NO. | POURING TEMP. °F | CONDITION | MECHANICAL PROPERTIES | | | |
|---|---|---|---|---|---|---|
| | | | TENSILE STRENGTH, PSI | YIELD STRENGTH, PSI * | ELONGATION, % IN 2" | BRINELL (500 Kg) |
| 101-1 | 1970 | WITHOUT DRY NITROGEN | 55,200 | 36,200 | 11.0 | 78 |
| 101-3 | 1970 | WITH DRY NITROGEN | 69,500 | 37,700 | 29.6 | 90 |

* DIVIDER METHOD

have satisfied the City of Los Angeles on all mechanical requirements.

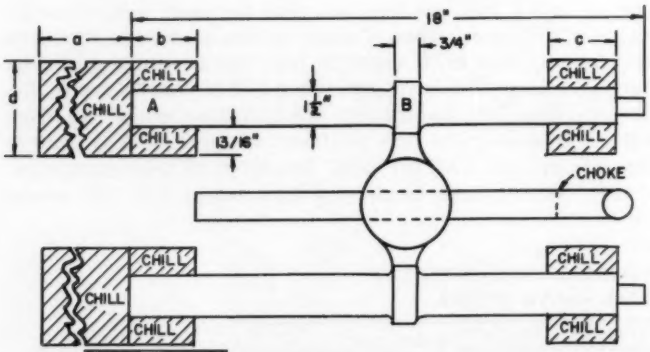
A study of the stem design indicated that a logical approach to directional solidification and consequently, internal soundness, would be to runner into a riser gated at the collar (Fig. 2). Foundries avoid gating into the collar because it makes the cutoff, grinding and machining operations difficult. This fact was recognized at the start of the program, but the gating practice had to be adopted because no other practical way of securing complete internal soundness could be devised. The first run with the matchplate gated, as in Fig. 2, was unsuccessful because a crack or shrink draw occurred at the junction of collar and stem (C, Fig. 2). The reason for this was that the junction was a sharp 90 degree angle. A fillet (3/8-in. radius) cured this, and results of subsequent tests are presented in Fig. 3.

*Chills vs. No Chills*

Heat 124 was prepared without the use of chills, while Heat 118 with variations in graphite chills shown in Fig. 3. The control specimen for each heat was poured to determine melt quality. Neither Heat 118-3 with 3 in. diameter x 6 in. solid end chills and chills 3 in. O.D. x $1\frac{3}{16}$-in. wall x 2 in. placed at both ends of the stem (Fig. 3), nor Heat 118-4 with 3 in. diameter x 3 in. solid end chills and chills 3 in. O.D. x $1\frac{3}{16}$-in. wall x $1\frac{1}{2}$-in. were superior to Heat 118-5 which did away with the end chills and used only the 3 in. O.D. x $1\frac{3}{16}$-in. wall x 2 in. chills at both ends of the stem.

In other words, by using the gating system detailed in Figs. 2 and 3 in. O.D. x $1\frac{3}{16}$-in. wall x 2 in. graphite chills placed at (B) and (C) in Fig. 3, a 6 in. valve stem can be produced in cast this bronze which will meet the minimum requirements of Specification 8959 without heat treatment. As shown in the table (Fig. 3), properties such as those obtained in stems of this bronze exceed to a considerable extent the properties found in similar stems prepared with other commercial alloys (Table 3).

One pertinent point—the new dezincification resistant alloy was used in the gating system study. This does not imply that other materials would not have performed in a manner equal to the capabilities of their chemical composition. There is little doubt that had these materials been used with the gating technique in Fig. 2, values considerably higher than those reported to Table 3 for Lab. 119 and 120 would have been obtained.




| HEAT NO. | CHILL DIMENSIONS a | b | c | d | T.S. psi | Y.S.** psi | EL. %in2" | REMARKS | POURING TEMP. °F |
|---|---|---|---|---|---|---|---|---|---|
| 118-1* | | | | | 66,000 | 34,000 | 30.6 | CONTROL SPECIMEN | 2,000 |
| 118-3A | 6" | 2" | 2" | 3" | 68,000 | 35,000 | 26.8 | THREADED END | 2,150 |
| 118-3B | | | | | 66,700 | 36,000 | 26.6 | COLLAR SECTION | |
| 118-4A | 3" | 1½" | 1½" | 3" | 65,000 | 38,000 | 24.8 | THREADED END | 2,100 |
| 118-4B | | | | | 65,000 | 30,000 | 22.0 | COLLAR SECTION | |
| 118-5A | NO | 2" | 2" | NO | 67,000 | 38,000 | 24.6 | THREADED END | 2,050 |
| 118-5B | CHILL | | | CHILL | 65,700 | 36,000 | 20.5 | COLLAR SECTION | |
| 124-1* | | | | | 65,000 | 37,000 | 32.0 | CONTROL SPECIMEN | 2,000*** |
| 124-2A | NO CHILLS | | | | 53,000 | 42,000 | 7.7 | THREADED END | 2,050*** |
| 124-2B | NO CHILLS | | | | 65,000 | 36,000 | 19.0 | COLLAR SECTION | |

* STANDARD KEEL BLOCK (FIG. 1) ** DIVIDER METHOD *** ESTIMATED

Fig. 3 — Graphite chills effect on mechanical properties of new alloy 6 in. valve stem castings gated as in Fig. 2.

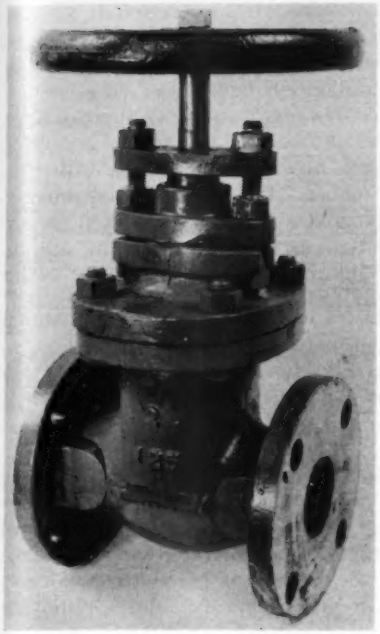


Fig. 4 — A 2 in. flanged valve with new alloy stem.



Fig. 5 — A stem of new alloy for 2 in. valve.

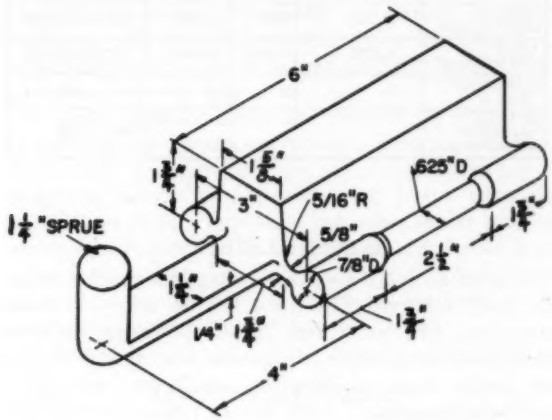



Fig. 6 — Modified Web-Webbert test bar.

## CHEMICAL COMPOSITION

As previously mentioned, the alloy was developed to meet the chemical and mechanical requirements set forth by the City of Los Angeles in their Department of Water and Power Specification No. 8959 for Double Disc Gate Valves. This specification obviously was written around the work of Tabor[2] and Streicher,[1] with a view toward eliminating dezincification and dealuminumization. Bearing in mind the chemical limitations of the specification, the following composition was devised which would produce an alloy with minimum mechanical requirements of said specification. The chemical composition of the new alloy and the chemical limitations of Specification No. 8959 are:

| Material | New Alloy, % | Specification 8959, % |
|---|---|---|
| Copper | 80.00 to 88.00 | |
| Nickel | 1.00 to 5.50 | |
| Iron | 1.00 to 5.50 | |
| Aluminum | 0.50 to 2.00 | 2.00 (Max.) |
| Silicon | 0.00 to 2.00 | |
| Zinc | 0.00 to 7.00 | 7.00 (Max.) |
| Others | 0.00 to 2.00 | |

Others include such elements as tin, lead, manganese, phosphorous, antimony and magnesium. Judicious use of tin, lead and manganese impart slight improvements to corrosion resistance, machinability and yield strength, respectively, without adversely affecting tensile strength, elongation and hardness.

## MECHANICAL PROPERTIES

The mechanical properties were established to satisfy structural and service demands of valve stems. It is assumed that the requirements were based on minimum mechanical properties of previously but unsuccessfully used materials because of lack of resistance to dezincification. Such alloys as silicon brass and manganese bronze come to mind as satisfying the tensile and elongation requirements (Table 3) of No. 8959 but fail to reach minimum yield strength, which has apparently been increased by the city of Los Angeles.

All tests were conducted with the standard keel block type of test bar (Fig. 1). Early in the development program a comparison was made between keel block and modified Web Webbert (Fig. 6) test bar properties. The results from tensile specimens cut from keel blocks were consistently superior, so this design was chosen as standard for the program. Since shrinkage characteristics of the alloy are similar to manganese bronze and aluminum bronze, it seems logical that the test bar design which suits these alloys would also suit an alloy with like characteristics.

Table 5 shows mechanical properties to which the alloy was developed in column No. 8959 as well as minimum, typical and range of this alloy's properties. The conclusion that all mechanical requirements are satisfied is obvious.

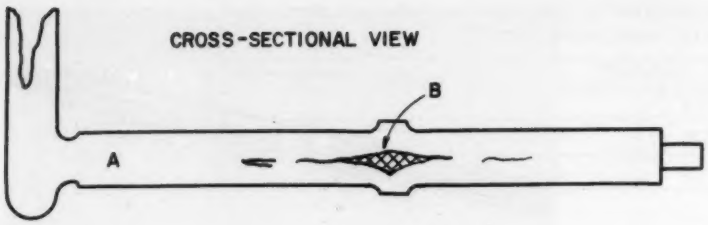



Fig. 7 — Valve stem gated at threaded end.

TABLE 5 — MECHANICAL PROPERTIES OF NEW BRONZE ALLOY

| MECHANICAL PROPERTIES | SPECIFICATION NO.8959, MIN. | N D Z BRONZE | | |
|---|---|---|---|---|
| | | MIN. | TYPICAL | RANGE |
| TENSILE STRENGTH, p si | 60,000 | 60,000 | 65,000 | 60,000-73,000 |
| YIELD STRENGTH, psi | 30,000 | 30,000 | 33,000 | 30,000-46,000 |
| ELONGATION, % IN 2" | 12 | 12 | 25 | 12 -40 |
| BRINELL HARDNESS NO.(500Kg) | | | | 78 -132 |

## MACHINABILITY

It is an accepted fact that ratings for machinability are only relative. This new bronze alloy machines about as well as silicon brass. Although 80-10-10, 65,000 tensile strength manganese bronze and 85-5-5-5 are much more machinable, satisfactory machining can be accomplished with proper tooling, lubrication and reduced speeds.

## CORROSION

As a result of the work of Tabor,[2] it was concluded that dezincification and dealuminumization could be prevented by limiting aluminum and zinc in bronze to 2 and 7 per cent, respectively. Therefore, it may also be concluded, with respect to the new dezincification resistant alloy, that it will not dezincify and/or dealuminumize since it conforms to Tabor's conclusions.

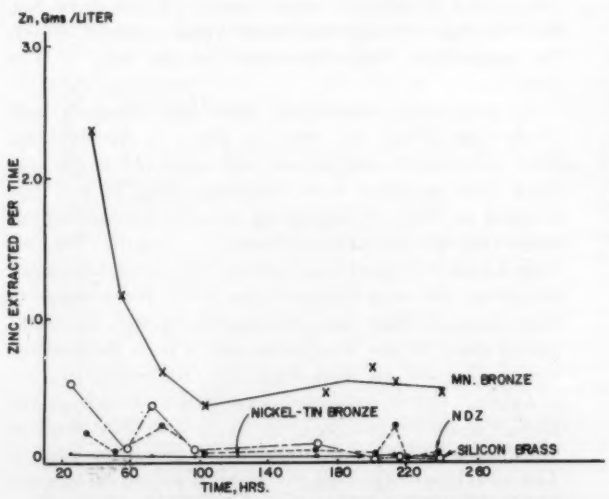



FIG.8- RATE OF DEZINCIFICATION EFFECT OF FILM ON RATE G/L VS. TIME

Acquisition of corrosion data at its best is most generally difficult, time consuming and frequently inconclusive. To get an idea of how the new material compared with the more widely used stem alloys, an accelerated dezincification test was devised by the author's company's chemical laboratory. The purpose of the accelerated test was to show the dezincification characteristics of established alloys versus experimental compositions. An essential prerequisite was to avoid specifically any influences that might be generated by coupling with any other metal. It is known that dezincification can be enhanced or suppressed, depending upon the variety of metal to metal contacts established in placing a valve into a plumbing installation.

The synthetic dezincification solution employed consisted of cupric chloride (5 per cent by weight) and of hydrochloric acid (one per cent by volume). Each specimen was suspended in 1000 ml of the above solution for a total of 240 hr. The temperature was thermostatically controlled between 95-100 C. The solution was vented to the atmosphere through a 32 in. air condenser. The specimen was suspended in the center of the flask with a $\frac{1}{16}$-in. diameter line.

Every 24 hr the specimen was removed, washed with hot water, dipped in acetone and dried at 105 C, weighed and its resistance determined with a Kelvin double bridge. Simultaneously, 25 ml of the synthetic dezincification solution was removed for analysis of zinc, copper and nickel. After completing the resistance measurements, the specimen and 25 ml fresh synthetic dezincification solution were returned to the flask.

By knowing the exact composition of the solution initially, the subsequent weight change of the specimen and a shift in the composition of the zinc, copper and nickel content of the solution, it was possible to determine what per cent of zinc, copper and nickel preferentially went into the solution above and beyond the incremental weight changes of the specimen. Thus, it became possible to predict expediently the tendency of any alloy to show dezincification. Also, because of the acute quantitive nature of this test procedure, it became possible to pin point

Fig. 8 — Corrosion curve.

minute differences in the tendency to dezincify in alloys that had previously been presumed to be comparable in their resistance to dezincification.

Previously established dezincification test methods have a tendency to be insensitive quantitatively and exceedingly prolonged in yielding reliable data. Thus, with this accelerated synthetic dezincification test it became possible to quickly evaluate comparative dezincification tendencies in a large number of experimental alloys that yielded desired mechanical properties. Ultimately, the compositions yielding optimum mechanical properties and resistance to dezincification were isolated.

Figure 8 shows how the new alloy lines up. Manganese bronze appears to corrode the fastest. Nickel-tin bronze, silicon brass and the new bronze alloy vary in their corrosion rates until about 220 hr has elapsed, at which time all alloys level off at about the same point.

There is, of course, no better evidence than that obtained by practical test. To this end the following is in progress:

1. Stems of the new alloy are in service in southwestern United States.
2. Immersion specimens and a 2 in. flanged valve are undergoing tests at Wrightsville Beach, North Carolina, by The International Nickel Co.
3. Corrosion cells are installed at Curtis Publishing Co., Philadelphia. This research will yield information on manganese bronze, Navy M, silicon brass, nickel-tin bronze and the new alloy.
4. Specimens are being subjected to tests described in the work of Tabor.[2]

## ELECTRICAL CONDUCTIVITY

Per cent electrical conductivity of five materials for corrosion cell studies at Curtis Publishing Co. was determined prior to the corrosion test. The reason for this is that periodic determination of electrical conductivity as the cell test progresses are expected to indicate corrosion rate. At present, however, an interesting comparison of the electrical characteristics may be made (Table 6).

## APPLICATIONS

The new bronze alloy was developed primarily to be used as a valve stem material (Figs. 4 and 5). Valve bodies and bonnets have also been produced. In view of the corrosion resistance and strength of the material, it would seem that it would be useful in other marine and corrosive applications, such as propeller wheels.

Electrical conductivity is surprisingly good. Coupled with high yield strength many applications come to mind requiring the combination of good electrical and mechanical properties.

## CONCLUSIONS

A copper-base alloy which resists dezincification and dealuminumization has been developed. Typical mechanical properties are 65,000 psi tensile strength, 33,000 psi yield strength and 25 per cent elongation. The alloy, in addition to copper, contains iron, nickel, aluminum, silicon and zinc.

Foundry characteristics are similar to manganese bronze and/or aluminum bronze. Tensile specimens taken from commercial valve stems have mechanical properties which are, as expected, lower than the capabilities of the individual materials under test bar conditions. A properly gated, risered and chilled 6 in. valve stems will have same mechanical properties in the critical sections as those which are available in the alloy prepared under test bar conditions.

TABLE 6 — PERCENTAGE ELECTRICAL CONDUCTIVITY OF FIVE CASTING ALLOYS USED FOR VALVE STEMS

| MATERIAL | %ELECTRICAL CONDUCTIVITY |
|---|---|
| MN BRONZE (65,000 psi) | 17.76 |
| NAVY "M" | 14.92 |
| SILICON BRASS | 6.42 |
| NICKEL-TIN BRONZE | 12.08 |
| NDZ | 16.82 |

## ACKNOWLEDGMENTS

The author wishes to thank the management of R. Lavin & Sons, Inc. for its interest in this project. Were it not for this, the development could not have been accomplished. A special mention is due Mr. Carl Morken, Vice Pres. & Works Mgr. of The Kennedy Valve Mfg. Co., Elmira, N.Y., who was instrumental in undertaking such a project, and who has extended full cooperation and contributed practical ideas as the research progressed. Mr. Floyd Keller, Head of Laboratories, and members of his staff are to be recognized for the accelerated corrosion testing and the many chemical and mechanical determinations required for the culmination of the alloy development program.

The Standard Casting Corp. of Chicago cooperated in preparing many of the heats necessary to determine the composition of the new alloy, and all the heats required to find proper gating and risering for internal soundness in 6 in. valve stem castings. The Iowa Valve Co. of Oskaloosa, and The Darling Valve & Mfg. Co. of Williamsport, Pa., who have subjected the alloy to foundry trials, and have been instrumental in getting corrosion tests performed, respectively.

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# DIE AND PERMANENT MOLD CASTING ALUMINUM ALLOY MINOR ELEMENTS

by D. L. Colwell and R. J. Kissling


## ABSTRACT

**The aluminum alloys used in die casting and permanent mold processes usually contain silicon or silicon and copper as alloying elements. There are other elements, beneficial and not, which are usually present in larger or smaller amounts. This report gives results of work done on different grades of these alloys, and the effects of some of these minor elements — namely, zinc, magnesium, iron and manganese.**


## INTRODUCTION

The common aluminum alloys used in the die casting and permanent mold processes usually contain silicon or silicon and copper as the most common alloying elements. There are several other elements however which are usually present in greater or lesser amount, some of which are beneficial, and others not beneficial. The purpose of this paper is to present some results of work done on various grades of these alloys and the effects of varying amounts of the four elements iron, zinc, magnesium and manganese.

In his Dudley Award paper for the American Society for Testing Materials, W. Bonsack has covered in great detail the effects not only of these four elements but of others less likely to be encountered.[1] Work done since this publication confirms the conclusions he drew, and it seems desirable at this time to summarize some of these confirmations.

## IRON

Iron is always present in aluminum. In permanent mold castings superior mechanical properties, particularly elongation and impact strength, can be obtained when the iron content is low. In two popular permanent mold casting alloys, SG70A and SC51A, the American Society for Testing Materials specifies iron at maxima of 0.50 per cent and 0.6 per cent, respectively.[2] With the advent of higher purity aluminum, it has been found that with an iron content of about 0.10 to 0.15 per cent the elongation of these alloys is greatly improved.

By modification in heat treatment, higher yield strengths can be secured without losing elongation.[3] Consequently the high purity grades of these two alloys are being used in many instances. These high purity compositions have been recognized by A.S.T.M. in the addition of SG70B and SC51B showing iron maxima at 0.15 per cent.[2] Because of the higher purity of aluminum required these grades are somewhat more expensive, and care should be taken in their manufacture to use silicon of the lowest possible iron content.

In die casting alloys an iron content of around one per cent is highly desirable. When the pot metal alloys direct from the reduction cell were introduced to the die casting industry, they carried iron contents of 0.50 per cent or below, and an epidemic of soldering to dies and cores developed. Iron is now added to these pot metals, and the soldering difficulties are not nearly as prevalent as they were. Residual cryolite contents however are often continuing the soldering difficulty.

The principal effect of iron on the properties of the die casting is to increase hardness and decrease elongation and impact strength. This is illustrated in Fig. 1. The tensile strengths, elongation and Charpy impact strengths on four common die casting alloys with iron contents ranging from 0.75 to 2.05 per cent are plotted. S5C has a lower tensile strength and higher elongation than the other three, S12A, SC84A and SG100A, but the effect is the same.

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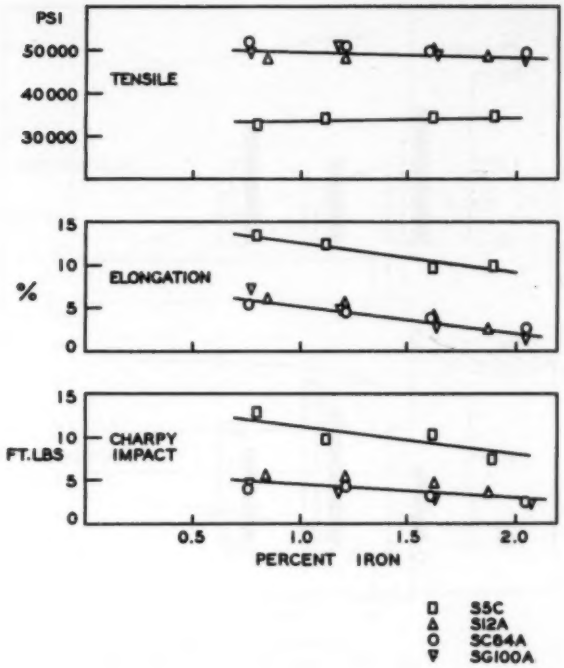



Fig. 1 — Iron effect on four die casting alloys aged one year at room temperature.

The values for the last three are plotted with one curve, but the individual points indicate that iron is somewhat more embrittling to SG100A and that SC84A has somewhat more tolerance for iron. These tests were made after one year of room temperature aging.

## ZINC

The principal effect of zinc up to about 3 per cent in aluminum alloys containing copper and silicon is to improve machineability. Even in aluminum-silicon alloys such as S5C and S12A, a zinc content of about 0.5 per cent also helps machineability and has no harmful effects. A summary of the effect of zinc on several properties was presented before AFS in 1952[4] and similar conclusions were drawn by the British Non-Ferrous Metals Research Association.[5] The effect of zinc ranging from 0.4 to 2.0 per cent on CS43A alloy is shown in Table 1.

After one year of room temperature aging, the yield strength is somewhat higher with 2.0 per cent zinc, and there is almost no loss of elongation. In the T6 condition the yield strength is somewhat higher, and neither the tensile strength nor elongation is affected. The recent decision of the A.S.T.M. to increase the maximum zinc content of SC84A from one to 3 per cent was made after years of exhaustive testing. Under the control of Committee B6, test bars of this alloy with various zinc contents were exposed at two sites; one on the roof of the Port Authority Building in New York City, and the other 800 yards from the sea at Kure Beach, North Carolina. The results after 6 years exposure at these sites have just been released, and are summarized in Fig. 2.[6]

Each point on these curves is the average value of the tests of three laboratories on the bars made by

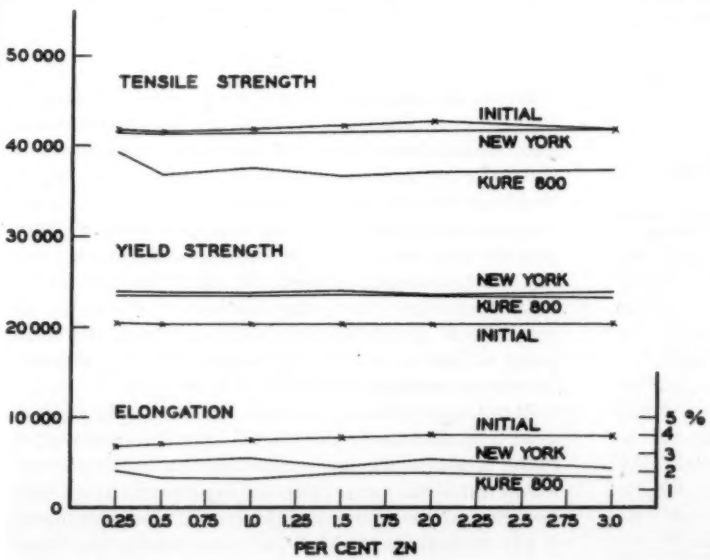
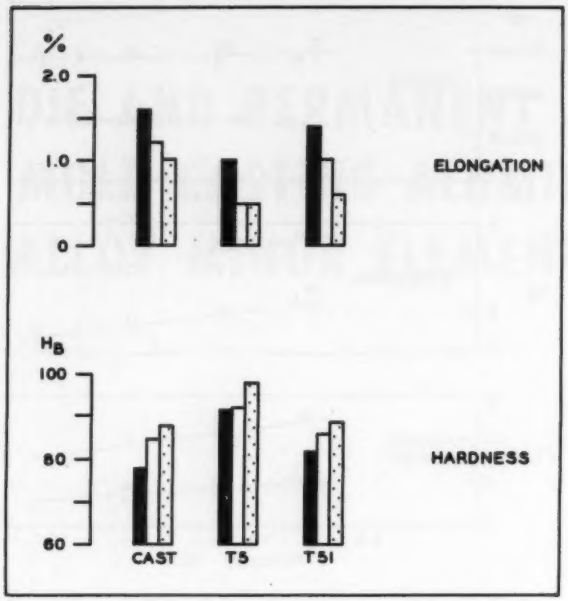



Fig. 2 — Die cast SC84A alloy with increasing zinc content after 6 years outdoor exposure at New York and at Kure Beach.




| | MG | ZN | MN | SI | CU | FE | NI |
|---|---|---|---|---|---|---|---|
| ■ | 0.09 | 0.62 | 0.28 | 8.73 | 3.90 | 0.77 | 0.07 |
| □ | 0.59 | 0.84 | 0.27 | 9.03 | 3.76 | 0.75 | 0.08 |
| ⁘ | 0.56 | 0.87 | 0.64 | 9.10 | 3.72 | 0.99 | 0.41 |

CAST 1 DAY AT ROOMTEMPERATURE
T5 400F / 8 HRS
T51 450F / 8 HRS

three producers. There was some loss in tensile strength at the Kure Beach location, but the New York exposure indicated no effect on tensile strength. At both sites the yield strengths were higher and the elongations lower due to the aging of the alloy. The zinc contents, however, did not affect any changes due to the 6 years exposure, as the slope of the curves in every case parallels the curves of initial properties, and the 3 per cent zinc properties were equal to the 0.5 per cent zinc properties after 6 years. Even the expected severe corrosion of the Kure Beach location was no worse on the high zinc samples than on those with less zinc.

TABLE 1 — ZINC EFFECT ON PERMANENT MOLD CAST CS43A

| | Zn, % | Yield Str., psi | Tensile Str., psi | Elong., % |
|---|---|---|---|---|
| Room | | | | |
| One Year | 0.4 | 22,500 | 33,100 | 1.8 |
| | 1.0 | 23,500 | 32,900 | 1.5 |
| | 1.5 | 22,900 | 35,400 | 1.6 |
| | 1.75 | 26,400 | 34,200 | 1.5 |
| | 2.0 | 25,600 | 32,400 | 1.4 |
| T6 | 0.4 | 25,200 | 40,500 | 4.0 |
| | 1.0 | 25,900 | 39,900 | 3.7 |
| | 1.5 | 26,700 | 40,300 | 3.6 |
| | 1.75 | 27,000 | 40,200 | 3.7 |
| | 2.0 | 27,000 | 41,600 | 4.0 |

Fig. 3 — High magnesium and high magnesium plus manganese content effect on permanent mold cast SC84A alloy.

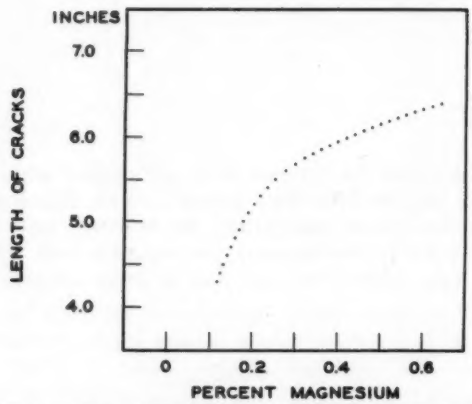



Fig. 4 — Magnesium effect on hot shortness of permanent mold cast SC64D alloy.[13]

TABLE 2 — COMPOSITION OF DIE CAST SC84A TEST BARS FOR EFFECT OF MAGNESIUM

| Composition, % | | | | | | |
|---|---|---|---|---|---|---|
| Mg | Zn | Mn | Si | Cu | Fe | Ni |
| 0.10 | 0.93 | 0.32 | 9.65 | 3.32 | 0.91 | 0.18 |
| 0.24 | 0.92 | 0.32 | 9.55 | 3.24 | 0.86 | 0.17 |
| 0.46 | 0.99 | 0.34 | 9.66 | 3.37 | 0.89 | 0.17 |
| 0.57 | 0.98 | 0.32 | 9.77 | 3.33 | 0.85 | 0.17 |

## MAGNESIUM

Where magnesium is not required as an alloying element it is limited by most U. S. aluminum casting alloy specifications to 0.10 per cent maximum. In the two popular permanent mold casting alloys SC51A and B and SC70A and B, silicon and magnesium form magnesium silicide, and the solution and precipitation of this compound by heat treatments improves the mechanical properties. In the common silicon-copper casting alloys, the hardening effects of the magnesium silicide are harmful both from the standpoint of cold shortness (brittleness) and hot shortness (tendency to crack in the die or mold).

Magnesium should be held to low limits in these alloys. Fortunately magnesium can be removed so there is no excuse for requesting higher limits than about 0.15 per cent. The hardening effects of magnesium in a permanent mold casting alloy containing 4 per cent copper and 9 per cent silicon are shown

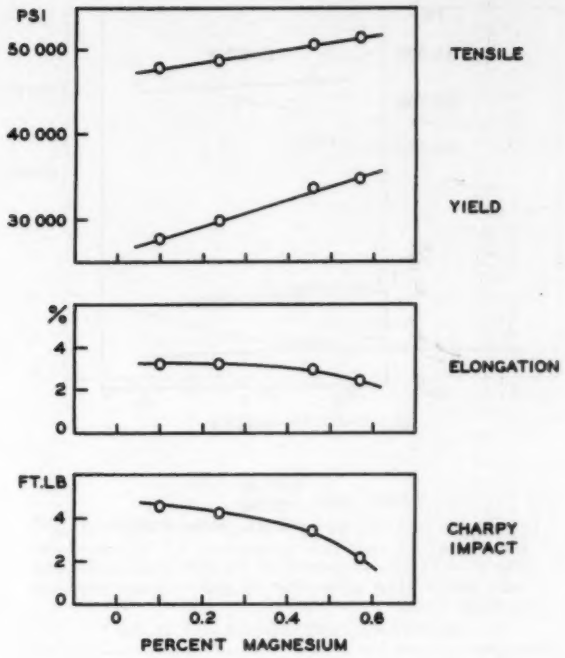



Fig. 5 — Magnesium effect on die cast SC84A alloy aged 6 months at room temperature.

in Fig. 3. The T5 and T51 properties are approximately those that would be typical after a few years service. The 50 per cent loss of elongation in the T5 condition is particularly important where this property is already at a low value.

Similar results are reported by Martin,[7] Quadt and Davis,[8] Scheuer[9] and B.N.F.M.R.A.[10] British specification DTD424[11] covering an alloy similar to the A.S.T.M. SC64D has a magnesium maximum of 0.15 per cent but recommends that the magnesium be kept below 0.10 per cent if optimum results are to be obtained. Gittins and Mew[12] have studied the effect of magnesium in combination with copper on LM4 alloy (5.8 per cent silicon) cast in sand. With no copper the magnesium can be as high as 0.40 per cent and with 4 per cent copper there should be no magnesium at all to obtain a minimum of 2 per cent elongation.

Bertram et al[13] found that 0.6 per cent magnesium reduced the elongation of chill cast SC64D alloy from 2 to one per cent. They also present an interesting study of the effect of magnesium on the hot shortness of this alloy. An increase in magnesium from about 0.1 to 0.6 per cent increases the hot shortness cracks in their test approximately one half. This is illustrated in Fig. 4. In the die casting alloy SC84A, magnesium up to 0.6 per cent gradually increases the tensile strength and yield strength and reduces the elongation and Charpy impact strength.

The compositions tested are shown in Table 2, and the relative values after 6 months room temperature aging are shown in Fig. 5. The drop in Charpy impact strength from 4.5 ft-lb at 0.10 per cent magnesium to 2.2 ft-lb at 0.57 per cent magnesium illustrates the harmful embrittling effect of magnesium. Conversely it should be pointed out that in alloys such as ZG32A and GM70B where magnesium is used for its strengthening properties, silicon is a harmful impurity and should be kept below 0.20 per cent. These represent the high strength casting alloys where heat treatment is unnecessary, and silicon in these alloys has the same harmful effect as magnesium does in the aluminum-silicon-copper-alloys.

## MANGANESE

Manganese in small quantities is usually considered a desirable addition to both die casting and permanent mold aluminum alloys, whereas in larger quantities it has a hardening and embrittling effect. In the permanent mold casting alloy SC51 for example, the A.S.T.M. recommends a manganese content in an amount equal to one half the iron if the iron content exceeds 0.45 per cent.[2] It is a desirable addition to the high strength alloys ZG32A and ZG42A, and is used to improve the high temperature strength in the piston alloys SC122A and CS66A. In most other alloys the acceptable maximum is 0.50 per cent.

Above this amount it increases hardness and decreases ductility and is likely to give trouble in casting. The dotted right hand bars in Fig. 3 indicate how a manganese content of 0.64 per cent lowers the elongation and increases the hardness of a chill cast alloy which is already hard and brittle.

The smaller quantities of manganese are of value in changing the iron-aluminum complex from a harmful needle or plate type structure into a complex iron-manganese-aluminum which minimizes the harmful effect of the iron. The reduction in elongation and impact strength with increasing iron content shown in Fig. 1 would probably have been much greater had it not been for the presence of manganese.

Manganese is almost a necessity in aluminum die casting alloys up to about 0.40 per cent as the manganese-iron combination greatly reduces the tendency of an aluminum alloy to solder. One of the great difficulties with the pot metal type of alloy being offered to the die casting industry is in soldering, and one of the reasons for soldering is the low iron and manganese usually in these alloys.

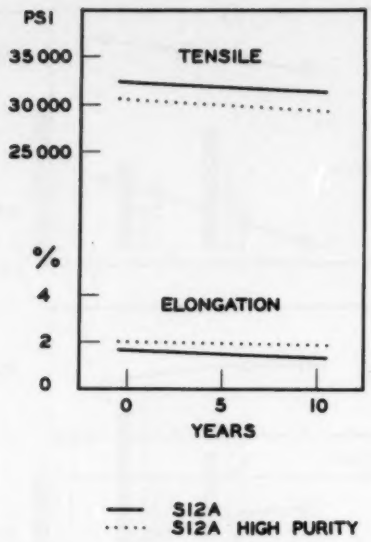



Fig. 6 — Tensile properties of die cast commercial and high purity S12A alloy after 10 years outdoor exposure at three sites.[15]

## HIGH PURITY ALLOYS

Although the heat treated properties of the high purity grades SC51B and SG70B discussed above are due to low iron content, die casting alloys seem to be improved by the impurities allowed by the specifications of the A.S.T.M. Committee B6 of the

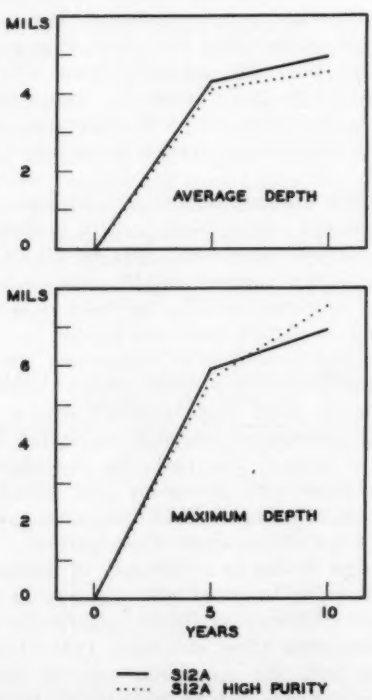



Fig. 7 — Depth of corrosion attack on die cast commercial and high purity S12A alloy after 10 years outdoor exposure at three sites.[15]

A.S.T.M.[14] initiated a cooperative exposure test on S12A back in 1934. Bars made by several producers were exposed at Sandy Hook, New York City and Altoona, Pa., and the results after 10 years exposure were summarized by the author in 1950.[15] The two compositions tested are shown in Table 3, and the average of the changes in mechanical properties at all three sites is shown in Fig. 6. The depth of corrosion attack after 10 years is shown in Fig. 7. Committee B6 drew the conclusion that after 10 years exposure both alloys exhibited "a high resistance to corrosion" and that "there were no really significant differences."[16]

TABLE 3 — COMPOSITION OF COMMERCIAL AND HIGH PURITY GRADES OF S12A TEST BARS FOR A.S.T.M. TEN YEARS OUTDOOR EXPOSURE TEST[15]

| Composition, % | Commercial Purity | High Purity |
|---|---|---|
| Mg | 0.055 | 0.005 |
| Zn | 0.52 | 0.00 |
| Mn | 0.26 | 0.01 |
| Si | 12.14 | 11.94 |
| Cu | 0.42 | 0.09 |
| Fe | 1.44 | 1.20 |
| Ni | 0.32 | 0.01 |

A similar test on alloy SG100A was initiated by Committee B6 in 1943 with similar results.[17] The two compositions tested are shown in Table 4, the effect of outdoor exposure for 10 years at the three sites in Fig. 8 and the effects of one year exposure in 20 per cent salt spray are shown in Fig. 9. The high purity grade has a slightly lower tensile strength and a higher elongation than the commercial grade. The nominal compositions of the alloys, and A.S.T.M. designations, are given in Table 5.

The atmospheric exposure shows a degree of age hardening for the first two years and then a flattening out of properties. The salt spray exposure shows a little faster deterioration of tensile strength in the commercial grade, and a little faster deterioration of elongation in the high purity grade. Generally speaking the properties of the commercial grade are equal to those of the high purity grades, and certainly the castability of the commercial grade is superior.

TABLE 4 — COMPOSITION OF COMMERCIAL AND HIGH PURITY GRADES OF SG100A TEST BARS FOR A.S.T.M. TEN YEAR OUTDOOR EXPOSURE TEST[17]

| Composition, % | Commercial Purity | High Purity |
|---|---|---|
| Mg | 0.54 | 0.54 |
| Zn | 0.006 | 0.00 |
| Mn | 0.01 | 0.005 |
| Si | 9.72 | 9.61 |
| Cu | 0.20 | 0.04 |
| Fe | 1.11 | 0.30 |
| Ni | 0.01 | 0.01 |

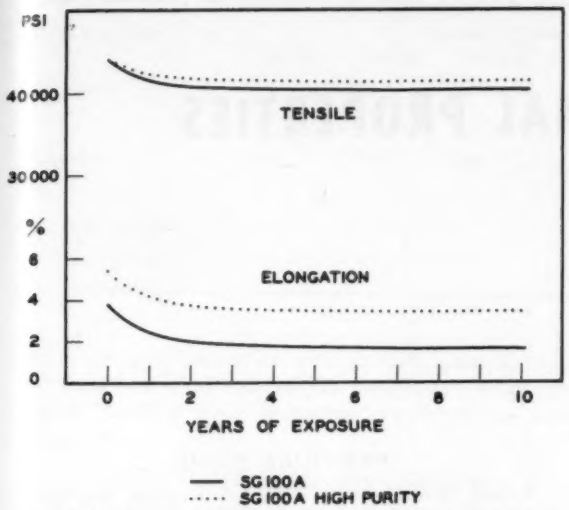



Fig. 8 — Ten years outdoor exposure at two sites effect on die cast commercial and high purity SG100A alloy.[17]

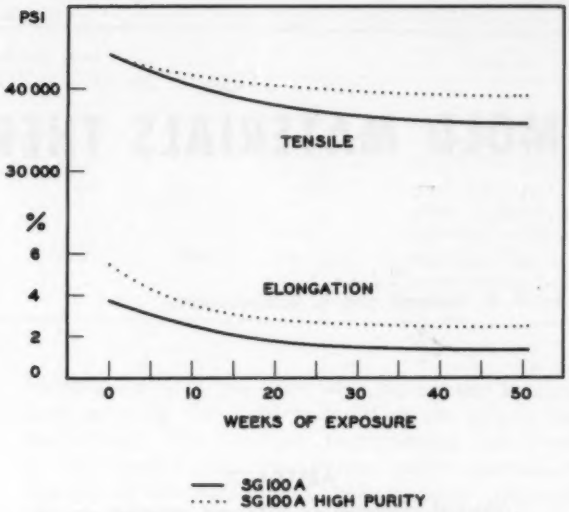



Fig. 9 — One year exposure to 20 per cent salt spray effect on die cast commercial and high purity SG100A alloy.[17]

## CONCLUSIONS

It is believed that the following conclusions are justified:

1. Except in exceptional applications the limits on minor elements specified by the A.S.T.M. will prove entirely satisfactory.
2. Iron has a hardening effect decreasing elongations, and particularly in certain permanent mold alloys it should be held low. In die casting alloys castability is improved up to one per cent or a little more, and beyond this embrittlement, particularly in SG100A, occurs.
3. Zinc up to the limits specified by the A.S.T.M. is valuable for improved machineability and has no harmful effects.
4. Magnesium in amounts greater than about 0.10 per cent is harmful both to castability and to the mechanical properties obtained. This element can be removed, and there is no sound reason for higher allowances.
5. Manganese is usually a desirable element up to about ½ of the iron content. It breaks up the iron needles and helps both castability and mechanical properties, particularly at high temperatures. Its use above the A.S.T.M. limits of 0.50 per cent cannot be justified, except in special cases such as in piston alloys where the effect on the coefficient of expansion is favorable.

TABLE 5 — ALLOY DESIGNATIONS

| A.S.T.M. Designation | Nominal Composition, % | | | | | |
|---|---|---|---|---|---|---|
| | Cu | Si | Mn | Mg | Zn | Cr |
| CS 43A | 4.0 | 5.0 | | | | |
| CS 66A | 6.5 | 5.5 | | 0.4 | | |
| GM70B | | | 0.2 | 7.0 | | |
| S5C | | 5.0 | | | | |
| S12A | | 12.0 | | | | |
| SC51A | 1.2 | 5.0 | | 0.5 | | |
| SC51B | 1.2 | 5.0 | | 0.5 | | |
| SC64D | 4.0 | 6.0 | | | | |
| SC84D | 3.5 | 8.5 | | | | |
| SC122A | 1.5 | 12.0 | 0.7 | 0.7 | | |
| SG70A | | 7.0 | | 0.3 | | |
| SG70B | | 7.0 | | 0.3 | | |
| SG100A | | 9.5 | | 0.5 | | |
| ZG 32A | | | 0.5 | 1.7 | 3.0 | 0.3 |
| ZG 42A | | | 0.5 | 2.1 | 4.2 | 0.3 |

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# MOLD MATERIALS THERMAL PROPERTIES

by M. R. Seshadri and A. Ramachandran


## ABSTRACT

**Thermal properties of new mold materials can be determined with ease by an unsteady state method involving melting and casting. Results of the present investigation reveal that (1) in general heat diffusivity values of mold materials bonded by $CO_2$ process will be higher compared to bentonite bonded, (2) among nonmetallic mold materials investigated, silicon carbide ($CO_2$) has the maximum heat diffusivity and synthetic sand the minimum and (3) solidification time of test castings calculated from the thermal properties of mold materials evaluated by the unsteady state method compares favorably with those of experimentally determined values.**


## INTRODUCTION

In recent years attention has been directed towards the scientific aspects of the production of quality castings. The casting quality is dependent upon a number of variables, such as the nature of metal or alloy cast, properties of the mold material employed and the casting technique. However, the thermal properties of the mold materials, as reported by earlier investigators, will determine to a large extent the rate of heat extraction for a given casting and thus influence its quality.

There have been several investigations in the past few years on the evaluation of thermal properties of mold materials by different methods. Of the several methods available for the determination of the thermal characteristics of the mold materials, the unsteady state method of evaluating thermal properties, involving melting and casting has attracted much attention. However, a literature survey indicates that limited data are available on the thermal characteristics of various mold materials bonded with sodium silicate, i.e., $CO_2$ process. Moreover, no data are available on indigenous mold materials bonded with bentonite. Hence experiments were conducted to determine the thermal properties of mold materials bonded with bentonite and sodium silicate, i.e., $CO_2$ process.

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## PREVIOUS WORK

Rapid developments have taken place recently in measurement of thermal properties of mold materials. The several methods adopted can be classified:

1) Steady state method for conductivity and temperature diffusivity.[1,2,3,4]
2) Unsteady state methods:
   a. Line or plane heat source for conductivity and temperature diffusivity.[5,6]
   b. Involving making of a casting for temperature diffusivity, conductivity and heat diffusivity.[7,8,9,10,11,12,13,14]
3) Methods of calculation for conductivity, temperature diffusivity and heat diffusivity.[14]

Measurements of the thermal conductivity of clay free sand, molding sand and several other molding materials including zircon, plaster and fired exothermic materials over a wide range of temperatures have been made by Atterton[1] using a steady state method. Effects of temperature, grain size and ramming density on thermal conductivity of mold materials, were also investigated by him. Care must however be taken that a correct mean value is used in applying Atterton's data to solidification problems.

Evaluation of thermal constants by unsteady state methods involving melting and casting have been reported by many workers. Ruddle,[8] Ohira,[9] Halbart,[10] Abcouver,[11] Briggs and Locke,[12] Adams and Taylor,[13] and Berry, Kondic and Martin[14] have determined the thermal properties of various mold materials by the unsteady state method, involving melting and pouring of metal or alloy into molds. Ruddle,[8] who has determined the mean temperature diffusivity for the following mold materials—(a) fine silica sand (Erith) dry state, (b) naturally bonded sand (Mansfield) in dry state, (c) silicon carbide grit (bonded) and (d) magnesite bonded (dead burnt powder)—has adopted Chvorinov's[7] method for evaluating the mean temperature diffusivity. Taylor[13] in his paper has stated that the thermal properties of mold materials must be determined under casting conditions in order to be useful for solidification studies.

The previous work, however, indicates that not much of data are available about the thermal prop-



61-100

TABLE 1A

| Mold Material | Grading, % retained on B.S. Sieve No. | | | | | | | Remarks |
|---|---|---|---|---|---|---|---|---|
| | −30+44 | −44+60 | −60+100 | −100+150 | −150+200 | −200+300 | Pan | |
| Silica Sand | 11.4 | 39.0 | 40.2 | 9.4 | — | — | — | Pale white in color composed mostly of sub-angular grains. |
| Magnesite | 11.00 | 30.50 | 39.60 | 11.20 | 7.40 | — | — | Brownish red in color. Angular and sub-angular grains. |
| Silicon Carbide | — | — | — | 6.8 | 23.0 | 48.6 | 21.6 | Slate gray in color mostly angular particles. |

erties of mold materials used in $CO_2$ process. Also, no data are available about indigenous mold materials bonded with bentonite employed in local foundries.

## EXPERIMENTAL PROCEDURE

High purity aluminum and aluminum-12 per cent silicon alloys (unmodified) were used as test casting for evaluating the thermal properties of nonmetallic mold materials by the unsteady state method involving melting and casting. The test casting, cylindrical in shape was 8 cm in diameter and 16 cm long.

### *Molding*

Nonmetallic mold materials, such as silica sand, magnesite and silicon carbide, the gradings of which are shown in Table 1A, were used for preparing bentonite bonded and $CO_2$ molds. Properties of the molds thus prepared and the treatments given to them are listed in Table 1B.

TABLE 1B — MOLD DETAILS

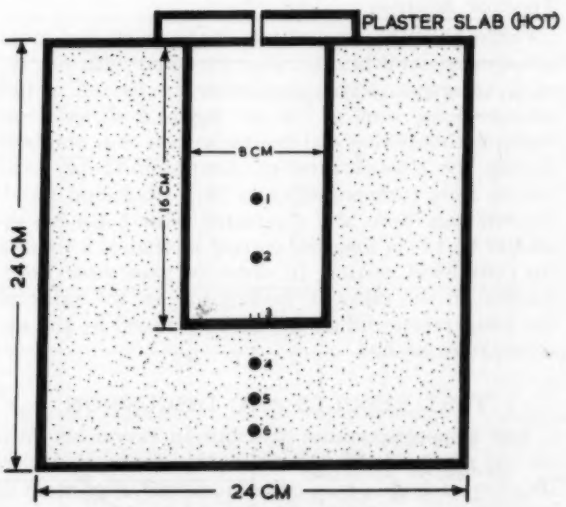
| Mold Material | Bond, % | Moisture content, % | Baking treatment hr | Baking treatment C (F) | Density, gm/cc (dry) |
|---|---|---|---|---|---|
| Silica Sand | 5 (Bentonite) | 4 | 8 | 200 (392) | 1.46 |
| Silica Sand | 3 (Sodium Silicate)† | 2 | 8 | 200 (392) | 1.54 |
| Magnesite | 5 (Bentonite) | 4 | 8 | 200 (392) | 1.60 |
| Magnesite | 3 (Sodium Silicate) | 2 | 8 | 200 (392) | 1.69 |
| Silicon Carbide | 5 (Bentonite) | 4 | 12 | 200 (392) | 1.65 |
| Silicon Carbide | 6 (Sodium Silicate)† | 2 | 12 | 200 (392) | 1.74 |

† Molds hardened by $CO_2$ process.

Molds were prepared using cylindrical molding boxes 24 cm in diameter and 24 cm long, having a number of vent holes on the periphery. In all the experiments, molds were hand rammed with necessary precautions by the same operator. Chromel-alumel thermocouples were placed in the mold at predetermined locations, as shown in Fig. 1. Due care was taken to minimize the shifting of the thermocouple during ramming. The actual positions of the thermocouples in the mold were determined by carefully scraping the mold material at the end of the experiment. The interface thermocouple was fixed in the manner used by Atterton and Houseman.[15] Twenty-six chromel-alumel thermocouples, sheathed in twin bore oval section 3 x 1.5 mm unglazed porcelain sleeves, were inserted from the top through the central hole of the plaster slab into the mold cavity for recording the metal temperature.

### *Melting and Pouring*

Normal melting and degassing procedures were adopted for melting high purity aluminum and aluminum-12 per cent silicon alloy (unmodified). Melting was carried out in a forced draught coke fired pit furnace. Casting temperature was 700 C (1292 F), and in all the experiments the metal was not allowed to exceed a temperature of 770-780 C (1418-1436 F) in the furnace.




THERMOCOUPLE POSITIONS

1. 80 MM FROM CASTING INTERFACE.
2. 40 MM FROM CASTING INTERFACE.
3. CASTING INTERFACE (ATTERTON TECHNIQUE)
4. 8 MM FROM MOULD INTERFACE.
5. 20 MM FROM MOULD INTERFACE.
6. 33 MM FROM MOULD INTERFACE.

Fig. 1 — Thermocouple position in mold.

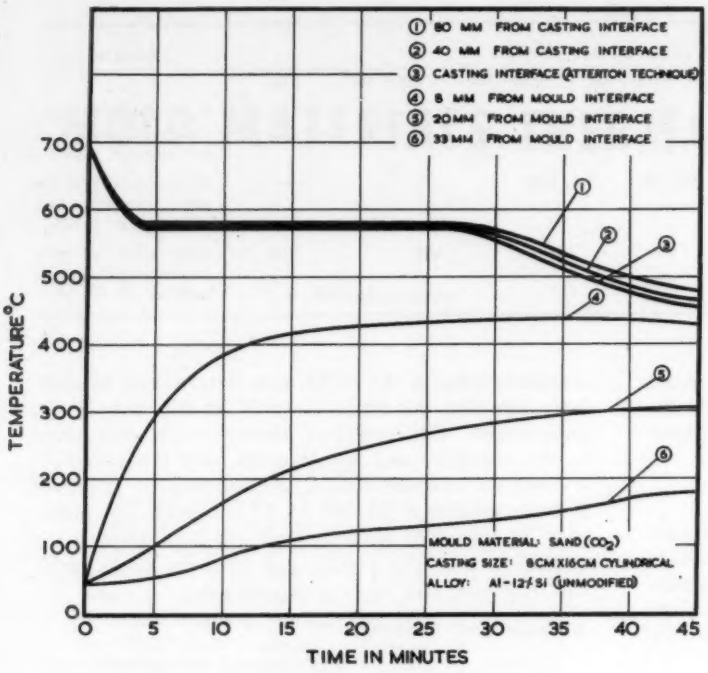



**Fig. 2 — Cooling and heating curves on the casting and mold side during solidification of Al-12 per cent Si alloy (unmodified).**

While pouring the metal into the molds due care was taken to minimize turbulence. As soon as the mold cavity was filled up with molten metal, a plaster slab was placed covering the top surface of the metal and the mold to minimize the heat loss from the top surface.

### *Thermal Analysis*

Cooling and heating curves on the metal and mold side during solidification were recorded with the aid of a six point high speed recorder having a full scale response time of 2.0 sec. Figure 2 shows as an example the cooling and heating curves thus obtained during the solidification of Aluminum-12 per cent silicon alloy (unmodified) cast in a $CO_2$ sand mold. Experiments were also conducted with a 6.5 cm diameter by 13 cm long test casting instead of 8 cm x 16 cm cylindrical casting, in order to gain some information on the effect of casting size on the value of the temperature diffusivity, $\alpha_2$ determined by the unsteady state method.

## TEST RESULTS AND DISCUSSION

The time-temperature distribution curves obtained on the mold side in various molds, during the solidification of high purity aluminum and aluminum-12 per cent silicon test castings, were analysed for the determination of the temperature diffusivity.

### *Temperature Diffusivity ($\alpha_2$)*

If the plane boundary of a semi-infinite solid body initially at a uniform temperature $\theta_o$ is instantaneously raised to a temperature $\theta_1$ at time $t = 0$, then after a time $t$ the temperature $\theta_m$ at any point whose perpendicular distance from the boundary is $x$, is given by

$$\theta_m = \theta_o + (\theta_1 - \theta_o)\,\mathrm{erfc}\,\frac{x}{2\sqrt{\alpha_2 t}} \qquad (1)$$

Where

$\theta_m$ = the temperature in C at a point in the solid whose perpendicular distance from the interface is $x$.
$\theta_o$ = Ambient temperature C.
$\theta_1$ = Interface temperature in C.
x = Distance from interface, in cm.
$\alpha_2$ = Temperature diffusivity of mold material.
t = time lapsed, in sec.

$\mathrm{erfc}\,\frac{x}{2\sqrt{\alpha_2 t}} = 1 - \mathrm{erf}\,\frac{x}{2\sqrt{\alpha_2 t}}$ and $\mathrm{erf}\,\frac{x}{2\sqrt{\alpha_2 t}}$ is the error function defined by

$$\mathrm{erf}\frac{x}{2\sqrt{\alpha_2 t}} = \frac{2}{\sqrt{\pi}} \int_0^{\frac{x}{2\sqrt{\alpha_2 t}}} e^{-\beta^2}\, d\beta$$

This equation can be applied to mold walls of low conductivity with little error,[15] despite the assumption that the mold wall is semi-infinite and thus permits evaluation of thermal properties of molding materials. However, for calculating the value of $\alpha_2$ one should know the correct value of $\theta_1$, i.e., the mold metal interface temperature. Several investigators have tried to measure $\theta_1$ experimentally or calculate the interface temperature from other data available. One can easily measure the casting skin temperature by the Atterton and Houseman technique whereas the measurement of mold-metal interface temperature offers considerable practical difficulties.

Reimann[16] and Halbart[10] have suggested a simple method of calculating the interface temperature provided we know the thermal diffusivities $b_1$ and $b_2$

of the metal and the mold material. Recently Berry, Kondic and Martin[14] have suggested a modification in this method of calculating the interface temperature, and this has been shown to be nearer the experimentally determined values.

According to Reimann[16] interface temperature $\theta_i$ is given by

$$\theta_i = \frac{b_1 \theta_1 + b_2 \theta_2}{b_1 + b_2} \qquad (2)$$

Berry et al[14] have suggested a modification of equation (2). The interface temperature $\overline{\theta_i}$ according to them is given by

$$\overline{\theta_i} = \frac{b_1' \overline{\theta_i} + b_2 \theta_2}{b_1' + b_2} \qquad (3)$$

where

$\overline{\theta_i}$ = weighted mean value of temperature.

$b_1' = k_1 \overline{\rho_1} \left( C_1 + \frac{L_1}{\theta_s} \right)$ where $\theta_s$ = solidification temperature.

In the present investigation interface temperature were calculated using equations (2) and (3), and are shown in Table 2 for ease of comparison. From this table one can conclude that the calculation of interface temperature by Berry et al[14] method is nearer to the experimentally determined values. Hence in this investigation for evaluation of temperature diffusivity from equation (1), interface temperature, i.e., $\overline{\theta_i}$ was used.

In Fig. 3 is shown the plot of $\theta_m$ against $\frac{x}{\sqrt{1.33t}}$ for $CO_2$ molding sand. A close agreement between the theoretical curve represented by an equation of the form $\theta_m = 30 + (571 - 30)\ \text{erfc}\ \frac{x}{\sqrt{3 \times 0.0009}}$ and the experimental points can be seen in Fig. 3.

In Table 3 are listed the temperature diffusivities thus determined for various mold materials with aluminum and aluminum-12 per cent silicon unmodified as test castings.

### *Apparent Thermal Conductivity ($\kappa_2$)*

From the specific heat data available in the literature, and the density and temperature diffusivity values experimentally determined, apparent thermal conductivity of various molding materials examined

TABLE 2 — TEST CASTING — HIGH PURITY ALUMINUM

| Poured at 700 C (1292 F) | Solidification range, 659-650 C (1218-1202 F) | |
|---|---|---|
| | Interface temperature (Calculated) C (F) | |
| Mold Material | Reimann Eq. (2) | Berry, Kondic and Martin Eq. (3) |
| Synthetic sand, dry | 667 (1233) | 647 (1197) |
| Sand, $CO_2$ | 658 (1216) | 644 (1191) |
| Magnesite, bonded | 655 (1211) | 640 (1184) |
| Magnesite, $CO_2$ | 654 (1209) | 639 (1182) |
| Silicon Carbide, bonded | 650 (1202) | 636 (1177) |
| Silicon Carbide, $CO_2$ | 647 (1977) | 634 (1173) |

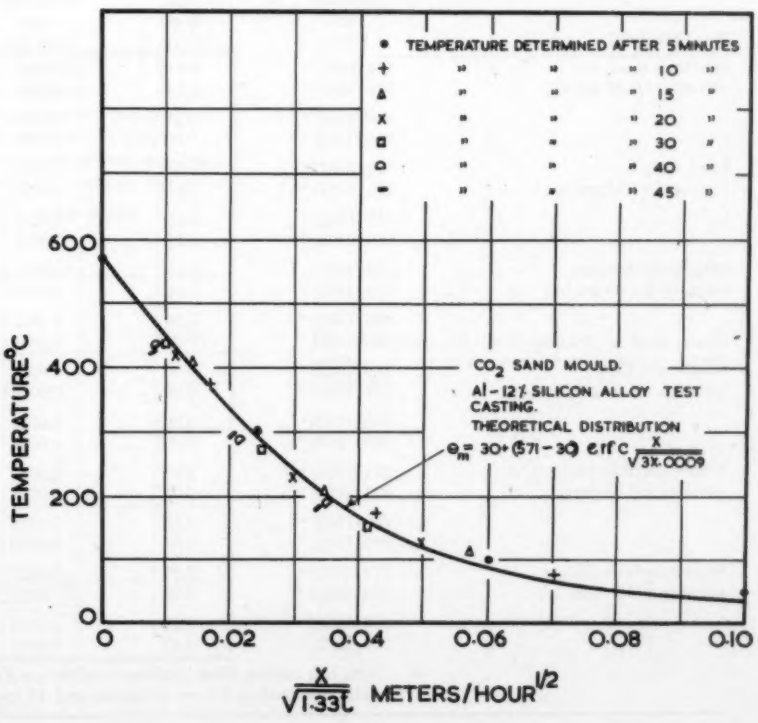



Fig. 3 — $\theta_m$ vs. $\frac{x}{(1.33\,t)^{0.5}}$.

were calculated. In column 5 of Table 3 are listed the apparent thermal conductivity values of various molding materials thus calculated corresponding to the temperature range 30 C (86 F) to 574 C (1065 F) and 30 C (86 F) to 650 C (1202 F).

### *Heat Diffusivity* ($b_2$)

In column 6 of Table 3 are listed the heat diffusivity values calculated for various molding materials at two interface temperatures namely 574 C (1065 F) and 650 C (1202 F). It can be seen from this table that the heat diffusivity values of mold materials prepared by $CO_2$ process are higher compared to those bonded with bentonite. Silicon carbide ($CO_2$) has the maximum heat diffusivity among the nonmetallic mold materials studied.

### *Effect of Temperature on* $K_2$, $\alpha_2$ *and* $b_2$

As reported by earlier investigators,[15] sand and silicon carbide either bonded with bentonite or by $CO_2$ process show an increase in thermal conductivity or heat diffusivity with the increase in interface temperature. Although in the literature magnesite is reported to have higher heat diffusivity compared to silicon carbide at aluminum and its alloy casting temperatures, it is found to have a lower value since the magnesite employed in this investigation is of a lower grade, as indicated by physical and chemical tests carried out in the laboratory. However, the decrease in thermal conductivity of magnesite with the increase in interface temperature as observed by some investigators in this field was found to hold good partly in this magnesite sample also.

### *Test Casting Size Effect*

From Table 3 one can also infer that on reducing the size of the test casting from 8 cm diameter x 16 cm long to 6.5 cm diameter x 13 cm long, the value of $\alpha_2$ obtained for a given mold material will remain unaltered. Probably further decreasing the test casting dimensions may result in a different value of $\alpha_2$ for the same mold material due to corner effects.

### *Mold Constant* ($\xi$)

The quantity $\xi$, often referred to as the mold constant, controls the rate at which the mold absorbs heat. It is a direct measure of the chilling power of the mold. Mold constant values for various molding materials calculated using the equation $Q = 1.128\, b_2 (\theta_1 - \theta_o)\sqrt{t}$ are listed in the last column of Table 3.

### *Total Heat Extracted and the Rate of Heat Extraction*

In Figs. 4, 5, 6 and 7 are shown the curves obtained by plotting the total heat extracted, $Q$ and the rate of heat extraction, $dQ/dt$ from mold surface of various mold materials against time for two interface temperatures. From Figs. 4 and 6 it can be seen that the heat extracted by the silicon carbide or magnesite

TABLE 3

| Mold Material | Temperature range 30 C (86 F) room temperature to C (F) | Mean specific heat (cgs) Cal/gm C | Experimentally determined mean temp. diffusivity (cgs) cm²/sec | Calculated mean apparent thermal conductivity (cgs) cal/cm sec C | Calculated mean heat diffusivity (cgs) | Calculated mold constant cal/cm²/min½ |
|---|---|---|---|---|---|---|
| Synthetic sand, dry | 574 (1065) [a] | 0.248 | 0.0019 | 0.0007 | 0.016 | 76.3 |
| (density = 1.46 gm/cc) | 574 (1065) [b] | 0.248 | 0.0019 | 0.0007 | 0.016 | 76.3 |
| | 650 (1202) [a] | 0.25 | 0.0022 | 0.0008 | 0.017 | 92.2 |
| | 650 (1202) [b] | 0.25 | 0.0025 | 0.0009 | 0.018 | 97.5 |
| Sand, $CO_2$ | 574 (1065) [a] | 0.248 | 0.0025 | 0.00095 | 0.019 | 90.5 |
| (density = 1.54 gm/cc) | 574 (1065) [b] | 0.248 | 0.0025 | 0.00095 | 0.019 | 90.5 |
| | 650 (1202) [a] | 0.25 | 0.0028 | 0.0010 | 0.020 | 108.5 |
| | 650 (1202) [b] | 0.25 | 0.0028 | 0.0010 | 0.020 | 108.5 |
| Magnesite, bonded | 574 (1065) [a] | 0.26 | 0.0034 | 0.0014 | 0.024 | 114.2 |
| (density = 1.60 gm/cc) | 574 (1065) [b] | 0.26 | 0.0035 | 0.00145 | 0.024 | 117.4 |
| | 650 (1202) [a] | 0.263 | 0.0031 | 0.0013 | 0.023 | 124.9 |
| | 650 (1202) [b] | 0.263 | 0.0034 | 0.0014 | 0.024 | 130.1 |
| Magnesite, $CO_2$ | 574 (1065) [a] | 0.26 | 0.0036 | 0.0016 | 0.0253 | 123.0 |
| (density = 1.69 gm/cc) | 574 (1065) [b] | 0.26 | 0.00364 | 0.0016 | 0.0258 | 123.0 |
| | 650 (1202) [a] | 0.263 | 0.0031 | 0.0014 | 0.0248 | 134.2 |
| | 650 (1202) [b] | 0.263 | 0.0034 | 0.0015 | 0.0259 | 140.5 |
| Silicon carbide, bonded | 574 (1065) [a] | 0.25 | 0.0036 | 0.00148 | 0.0247 | 117.8 |
| (density = 1.65 gm/cc) | 574 (1065) [b] | 0.25 | 0.0039 | 0.00161 | 0.0258 | 123.0 |
| | 650 (1202) [a] | 0.25 | 0.0039 | 0.0016 | 0.0258 | 139.8 |
| | 650 (1202) [b] | 0.25 | 0.0040 | 0.00165 | 0.0261 | 141.4 |
| Silicon carbide, $CO_2$ | 574 (1065) [a] | 0.25 | 0.0039 | 0.0017 | 0.027 | 128.5 |
| (density = 1.74 gm/cc) | 574 (1065) [b] | 0.25 | 0.0039 | 0.0017 | 0.027 | 128.5 |
| | 650 (1202) [a] | 0.25 | 0.0039 | 0.0017 | 0.027 | 146.0 |
| | 650 (1202) [b] | 0.25 | 0.0042 | 0.0018 | 0.028 | 152.0 |

[a] with test casting 8 cm diameter and 16 cm long.
[b] with test casting 6.5 cm diameter and 13 cm long.

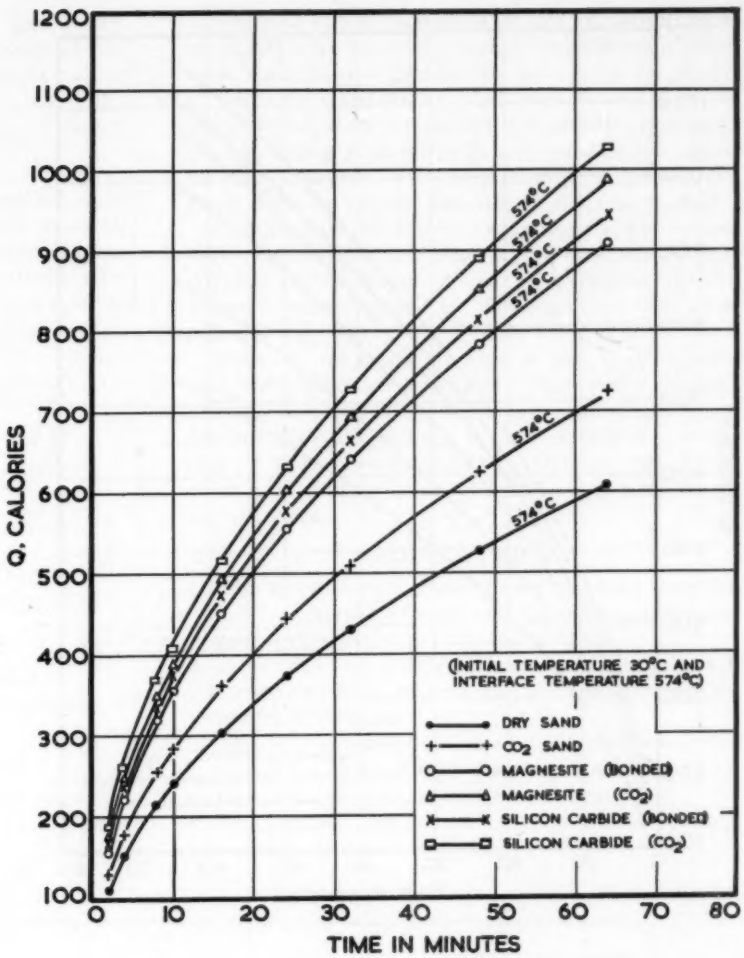



Fig. 4 — Total heat extracted by various mold materials at an interface temperature of 574 C (1065 F).

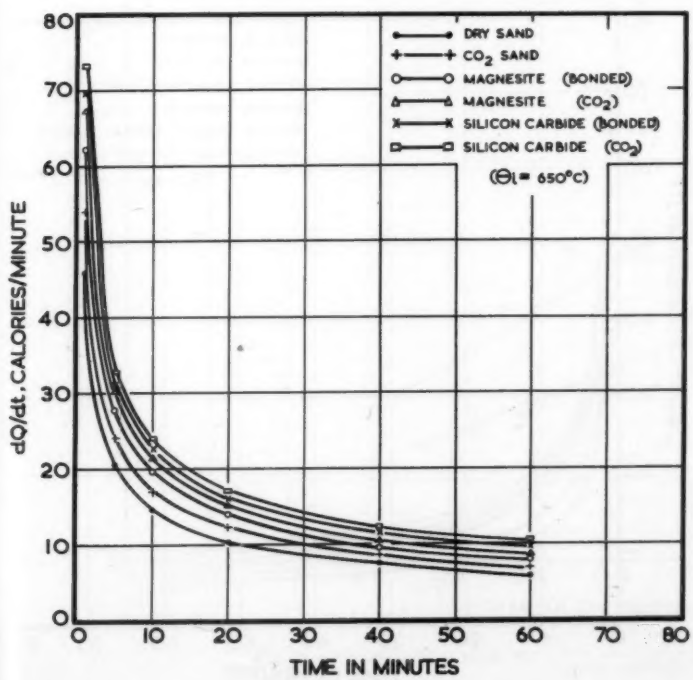



Fig. 5 — Rate of heat extraction from mold surface at an interface temperature of 574 C (1065 F).

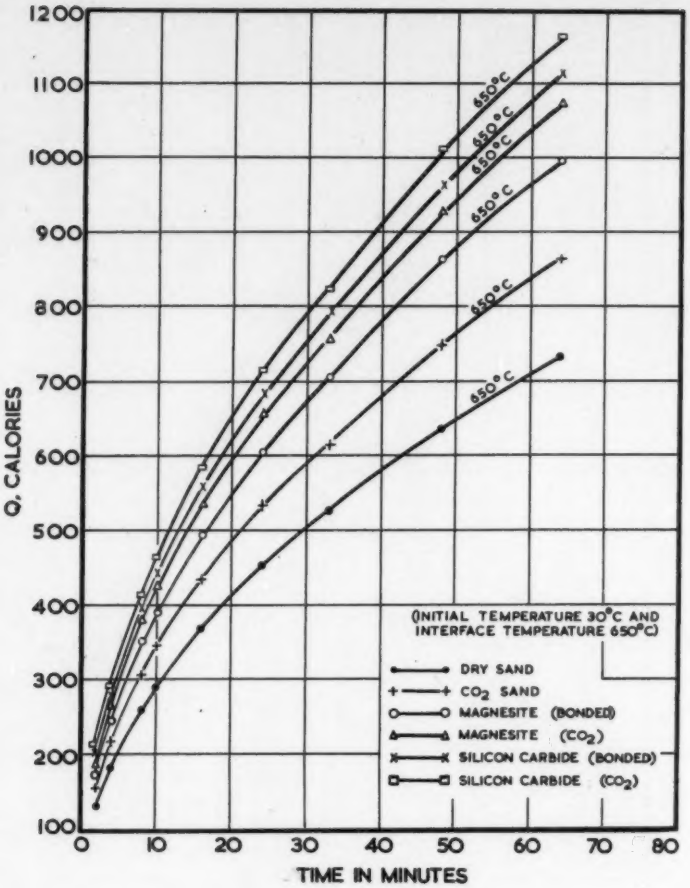



Fig. 6 — Total heat extracted by various mold materials at an interface temperature of 650 C (1202 F).

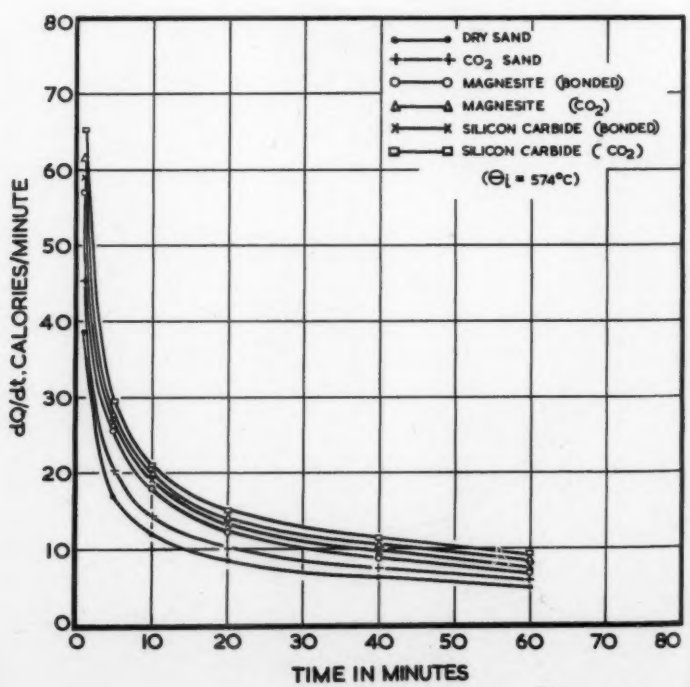



Fig. 7 — Rate of heat extraction from mold surface at an interface temperature of 650 C (1202 F).

molds is quite appreciable compared to sand molds. At an interface temperature of 574 C (1065 F), the total heat extracted and the rate of heat extraction per $cm^2$ of mold surface is more for magnesite ($CO_2$) and less for silicon carbide (bonded), Figs. 4 and 5.

### *Theoretical Solidification Time*

From the thermal properties of nonmetallic mold materials listed in Table 2, and the data available in literature on the thermal characteristics of pure aluminum and aluminum-12 per cent silicon alloy unmodified, the theoretical solidification time ($t_s$)[14] of test castings were determined using the equation

$$t_s = \frac{R^2}{\left[\frac{\sigma b_2 v}{2\rho_L \overline{q_L}} + \sqrt{\frac{v}{\rho_L \overline{q_L}}\left(\frac{\sigma^2 b_2^2 v}{4\rho_L \overline{q_L}} + \frac{k_2}{c_1 c_2}\right)}\right]^2} \qquad (5)$$

where

$\sigma = 1.13$.
$v = \theta_i - \theta_o$.
$\overline{q_L} = L_L + C_L(\theta_L - \theta_s)$.
$\rho_L$ = density of molten metal or alloy.
$\theta_L$ = pouring temperature of metal or alloy.
$\theta_s$ = solidification temperature.
$L_L$ = latent heat of fusion.
$b_2$ = heat diffusivity of mold material.
$k_2$ = thermal conductivity of mold material.
$c_1 = 2$ and $c_2 = 2$ for a cylinder.
R = volume/surface area ratio.

The solidification times calculated from equation (5) for the test castings of aluminum and aluminum-12 per cent silicon (unmodified) cast in various types of molds and the experimental solidification times were then plotted against the inverse of the square of the mold constant, i.e., $1/\xi^2$ Figs. 8 and 9. It can be seen from these figures that the solidification times of test castings is inversely proportional to the square of mold constant. This relationship found to hold good even on varying the size of the test casting.

Ruddle and Mincher's[8] experiments also reveal that the solidification time of test castings of pure copper and Aluminum-30 per cent copper alloy is inversely proportional to the square of the mold constant, Fig. 10.

However according to Figs. 8 and 9, it appears that the difference in experimental and calculated solidification times of Al-12 per cent silicon alloy (unmodified) test castings is more and is less in the case of pure aluminum. This variation may be attributed to the mode of solidification of pure metals and eutectics.

## CONCLUSIONS

1. The thermal properties of new molding materials can be determined with ease by the unsteady state method involving melting and casting.
2. In general heat diffusivity values of mold materials bonded by $CO_2$ process will be higher compared to bentonite bonded.
3. Among the nonmetallic mold materials investigated silicon carbide ($CO_2$) has maximum heat diffusivity and synthetic sand the minimum.
4. Solidification times calculated from the thermal

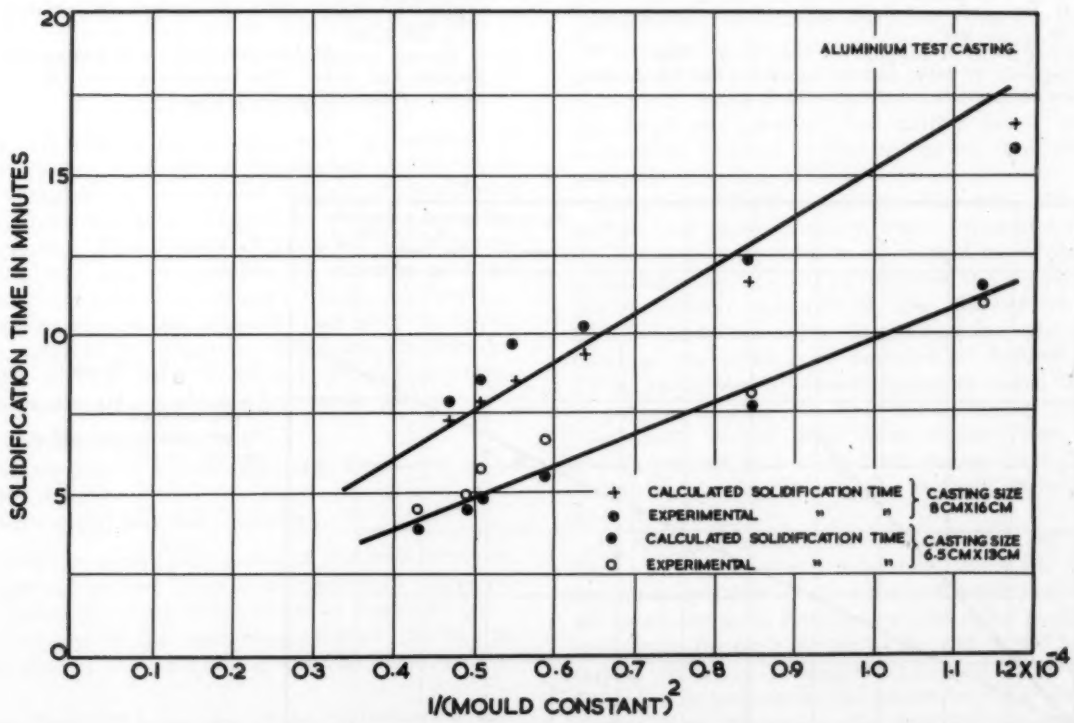



Fig. 8 — Solidification time vs. $\left(\frac{1}{\text{mold constant}}\right)^2$ for aluminum test casting.

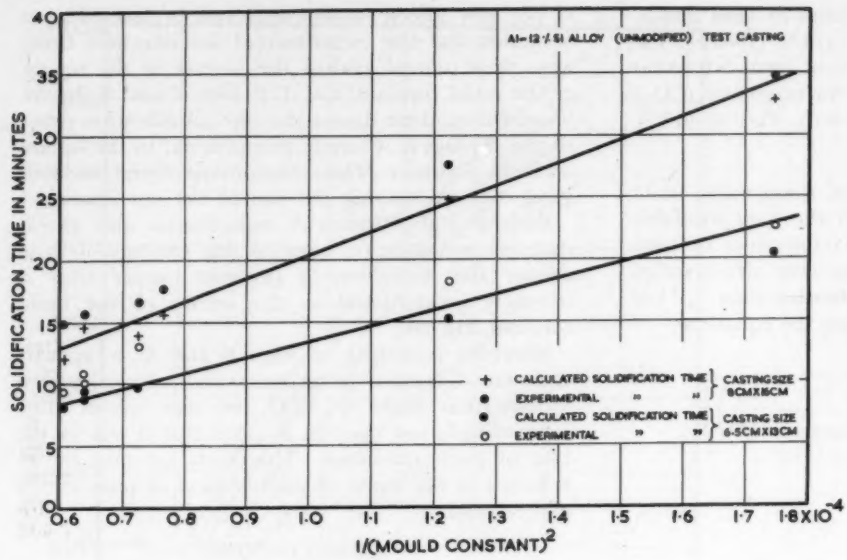



Fig. 9 — Solidification time vs. $\left(\frac{1}{\text{mold constant}}\right)^2$ for aluminum — 12 per cent silicon alloy.

properties of mold materials evaluated by the unsteady state method compare favorably with those of experimentally determined values.

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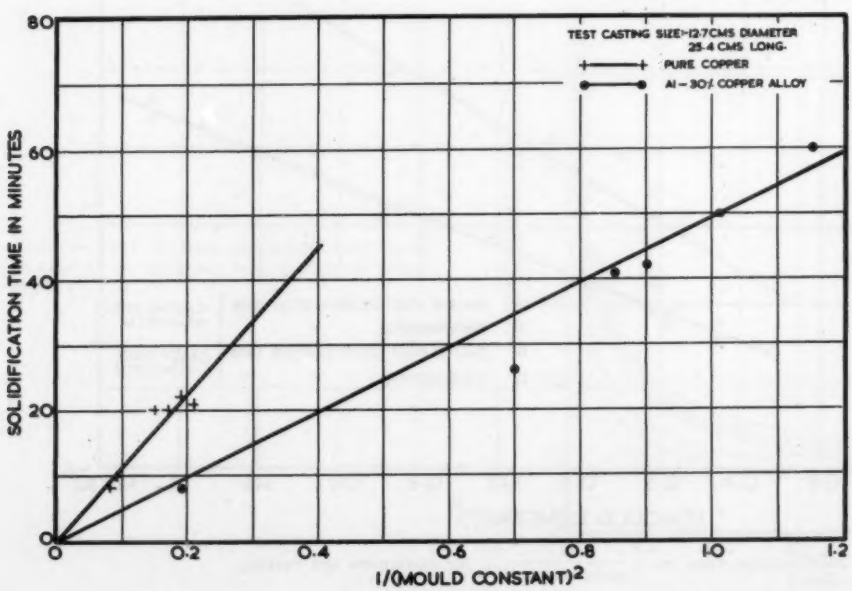




Fig. 10 — Solidification time vs. $\left(\frac{1}{\text{mold constant}}\right)^2$ for aluminum — 30 per cent copper and pure copper test castings.

# FLUIDITY OF ALUMINUM ALLOYS

## *An experimental and quantitative evaluation*

by M. C. Flemings, E. Niiyama and H. F. Taylor


### ABSTRACT

**Fluidity tests were conducted on aluminum-4.5 per cent copper alloy to determine the effects of small additions of a third element on fluidity. Elements studied included titanium, iron, manganese, cobalt, chromium, beryllium, silicon, magnesium, calcium and copper. Results observed were interpreted on the basis of the effects of the elements on (a) liquidus temperature, (b) solidification range and (c) nature of the primary crystals formed.**

**An equation was developed, based on heat and fluid flow considerations and on solidification theory to express fluidity of an alloy in terms of metal and mold variables. The equation describes fluidity behavior of mushy freezing alloys such as aluminum-4.5 per cent copper, qualitatively and quantitatively.**


### INTRODUCTION

In July 1957, a research program was initiated at Massachusetts Institute of Technology by the U.S. Army Ordnance Department through Pitman-Dunn Laboratory, Frankford Arsenal, to obtain a basic understanding of the factors affecting fluidity in aluminum alloys and, if possible, to develop practical methods for increasing fluidity. During the first two years' research on this program, tests were developed and employed to isolate the important variables affecting fluidity, and theoretical and experimental analyses were used to relate metal fluidity to:

1) Solidification mechanism.
2) Surface tension and surface oxide films.
3) Superheat.
4) Other metal and mold variables.

In addition, completely practical methods were developed to increase fluidity of aluminum alloys (in production) by as much as a factor of three.[1-4]

A portion of the research conducted during the third year of this program is described herein. It is a continuation of the earlier work, and was directed at (1) a study of effects on fluidity of adding third elements to aluminum-4.5 per cent copper alloy, and (2) developing a quantitative analysis of factors affecting fluidity of aluminum alloys.

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### ADDED ELEMENTS EFFECT ON FLUIDITY OF ALUMINUM-4.5 PER CENT COPPER ALLOY

#### *Test Apparatus and Procedure*

Fluidity tests were conducted in the vacuum fluidity tester (Fig. 1) in essentially similar fashion to that described earlier.[2] In each test, metal of known composition and temperature was drawn into the fluidity test channnel under predetermined pressures. The distance the metal flowed into the tube was taken as the measure of fluidity. In some cases, the stream of flowing metal was photographed with a movie camera to provide accurate determination of flow velocity and fluid life (time of flow).

Glass tubes ($0.200 \pm 0.006$ in. diameter) were used for the test channel, except where otherwise noted. To facilitate immersion in the liquid bath, one end of the tube was bent with a radius of about 4 in., with the horizontal portion of the fluidity channel $2.75 \pm 0.25$ in. above the entrance of the tube. After bending, the tubes were annealed at 800-900 F for 2-3 hr, and cooled to room temperature in the furnace to prevent breakage by heat shock during the test. The inside of the glass tubes cleaned with glass cleaner, rinsed and then dried immediately before testing by passing dry nitrogen through them for at least 20 min.

In testing, the tubes were immersed in the metal bath to a depth of approximately 0.6-0.8 in., so the horizontal portion of the tube was approximately 2 in. above the melt. Fluidity was measured as the distance from the melt surface to the end of the fluidity casting. When a photographic record of flow was desired, a 16 mm camera (64 frames/sec) was used. An electric clock was used so time and flow distance were recorded on the film simultaneously.

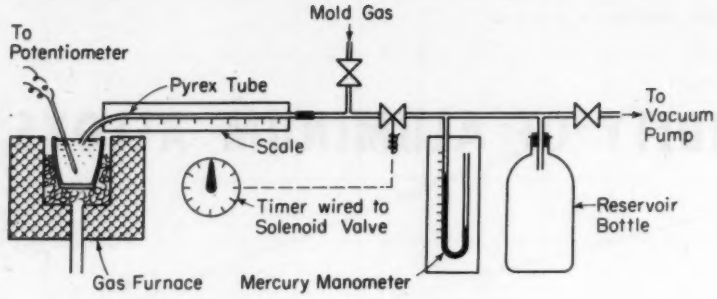



Fig. 1 — Vacuum fluidity test apparatus.

Melting was conducted in a clay-graphite crucible placed inside a gas-fired furnace. For some tests, a salt flux and/or nitrogen degassing was employed. However, small amounts of gas were found to have no significant effect on fluidity, and solidification and remelting (in the crucible) between runs kept the gas content adequately low. Melt temperature was measured using a chromel-alumel thermocouple with a protection tube of quartz or clay. The melt temperature at the instant of the test run was determined with an accuracy of $\pm$ 2 to $\pm$ 4 F.

Pressure was measured with a mercury manometer to within $\pm$ 1 mm Hg. This pressure was found to be unchanged when measured before and after testing. Pressures, in mm Hg, were converted to metal head (in. of aluminum) for subsequent calculations. Density of liquid 195 alloys (including the alloy with third additives) was taken as 2.42 grams/cc.

For tests in which a photographic record was taken, the procedure was—when the melt approached the testing temperature a fluidity tube was fixed in position for testing. The timer (Fig. 1) was started and the melt surface skimmed. At the precise melt temperature desired, the camera was started and the tube was dipped into the melt. At a predetermined time (0.3-0.7 sec), the vacuum was applied by opening a solenoid valve. After the test was complete, the timer and camera were stopped and the test tube removed.

### *Alloy Preparation*

Metal for most of the tests reported was melted from pre-alloyed ingots made from a single master heat. This pre-alloyed ingot was prepared from electrode copper and high purity aluminum (99.99 per cent) to a composition of 4.30 per cent-balance aluminum, and for continuity is frequently referred to herein as aluminum-4.5 per cent copper alloy, or as 195 alloy. The difference in fluidity between actual analysis (4.3 per cent) and the nominal analysis for 195 alloy (4.5 per cent) is essentially negligible, estimated to be at about 2 per cent.

Third elements added included titanium, magnesium, silicon, iron, beryllium, manganese, cobalt, chromium, calcium and lead. To assure accurate and homogeneous additons of these elements, master alloys were first made from these metals and pure aluminum or aluminum-4.5 per cent copper alloy. The master alloys were then analyzed and added to the base alloys to obtain the desired compositions for testing. Except where otherwise noted, copper contents of the final alloys were adjusted to 4.3-4.5 per cent.

### *Precision of Tests*

The precision of the vacuum fluidity test depends on the degree of control exercised over such factors as melt temperature, pressure head, tube diameter, radius of the bend in the tube, dip time, dip depth and cleanliness of the tube. Of these, tube cleanliness was found to have an unexpectedly large effect, apparently because thin films of impurities on the inside of the tubes affect heat transfer at the metal-glass interface. For most of the tests reported herein, the standard deviation of fluidity values was 3 per cent. In other words, 95 per cent of all the results fell in the range of $\pm$ 6 per cent of the average value, which was quite satisfactory for the present work.

### *Preliminary Tests*

Before studying effects of alloy additions on fluidity, a series of standard tests were conducted with pure aluminum-4.5 per cent copper alloy. In these tests only temperature and pressure head were varied (Fig. 2). A linear relation between fluidity and melt temperature was obtained, as would be expected from previous researches[1-3, 6, 7] and from simple theory (outlined later).

Motion picture records were made of each of the standard tests to determine fluid life* accurately. Fluid life decreases with decreasing melt temperature and with increasing metal head (Fig. 3). This agrees with previous experimental findings.[2] Theoretical explanation for the foregoing has been given in general terms,[2, 7] and is presented quantitatively later. Briefly stated, the explanation is—in an alloy with a wide range of solidification (such as aluminum-4.5 per cent copper), fluidity is limited by choking of flow at the leading tip of the stream, due to accumulation of solid crystallites.

As pressure head is increased, flow velocity increases and the flow tip is exposed to more chilling by coming into contact with a greater area of the cold mold walls. The increased pressure head therefore results in increased rate of solidification at the flow tip, and hence decreased fluid life (increased super-heat increases fluid life because more heat must be withdrawn at the tip to effect sufficient solidification to stop flow).

*Fluid life is the length of time metal flows in the fluidity test channel before stopping.

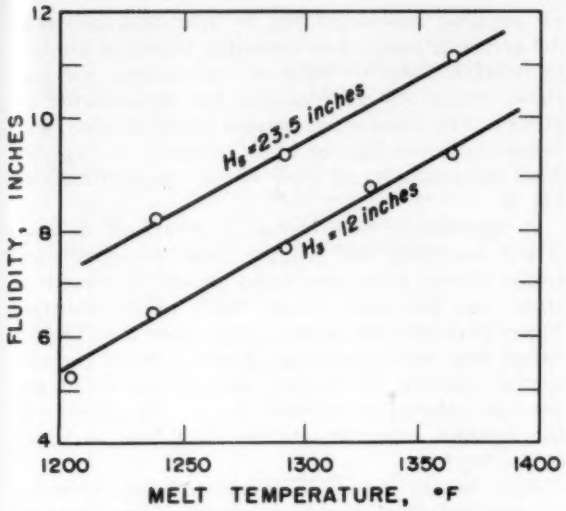



Fig. 2 — Melt temperature effect on fluidity for two different metal heads (tubes cleaned).

As additional evidence that choking occurs at the flow tip, it was observed several times in this research that when a tube was broken by the thermal stress at the instant of flow stoppage, almost all the metal in the tube flowed out, leaving only the extreme leading tip of the metal in the glass tube as a semi-solid. The length of this semi-solid part was sometimes less than ¼-in.

### *Grain Refinement Effect on Fluidity*

In order to determine the effect of commercial grain refiners on the fluidity of 195 alloy, tests were made in which 0.17 per cent titanium was added to the aluminum-4.5 per cent copper alloy. Results of these tests (Fig. 4) show that titanium reduces fluidity approximately 13 per cent at the lower pouring temperature and 10 per cent at the higher temperatures.

The reason why titanium reduces fluidity is not clear. It may well be due to the results of the grain refinement process, since fine particles have been shown to be much more effective in stopping a flowing stream than an equivalent percentage by weight of coarse particles.[8] However, titanium also has a slight effect on the solidification behavior of aluminum alloys in addition to its grain refining action. This may also affect fluidity. The ways in which various alloys affect fluidity are discussed in detail in the following section.

### *Alloying Agents Effects*

Data from many workers are in general agreement that alloying additions to pure metals nearly always result in reduced fluidity. However, the situation is quite different for alloys. For example, eutectic compositions in binary and ternary systems usually possess better fluidity than compositions which are richer or poorer in solute.[9-10] Also, various researches have shown that relatively small amounts of elements added to foundry alloys can bring about significant improvements in fluidity.[11-12]

Ways by which alloying elements can have beneficial effects on fluidity include (1) increase in superheat due to lowering of liquidus temperature, (2) change in temperature range over which solidification occurs, (3) change in the nature of primary crystals which form and (4) changes in the nature of surface films or oxides.

Items (1), (2) and (3) all suggest that alloys might increase fluidity of aluminum-4.5 per cent copper alloy if they form a "liquidus valley" in the ternary phase diagram composed of the alloy, aluminum and copper (if this valley is near the composition of aluminum-4.5 per cent copper). An example of what is meant by a liquidus valley is shown in Fig. 5, a portion of the pseudo-binary diagram for aluminum-5 per cent copper with iron added up to 3.0 per cent. At approximately 1.5 per cent iron a peritectic exists, and in this region some fluidity increase might be expected for one or more of these reasons:

1) The liquidus temperature is a minimum at 1.5 per cent iron. At this temperature it is 1188 F (642 C), compared with 1197 F (647 C) at 0 per cent iron and 1340 F (727 C) at 3.0 per cent iron. Hence, if fluidity is measured at constant pouring temperature, alloys of about 1.5 per cent iron will possess somewhat greater superheat due to depression of the liquidus.
2) The equilibrium temperature range of solidification is a minimum in the neighborhood of the peritectic. It is 140 F (60 C) at 0 per cent iron,

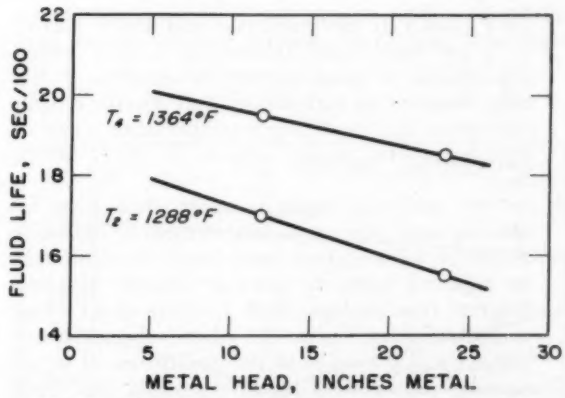



Fig. 3 — Metal head effect on fluid life for two different melt temperatures (tubes cleaned).

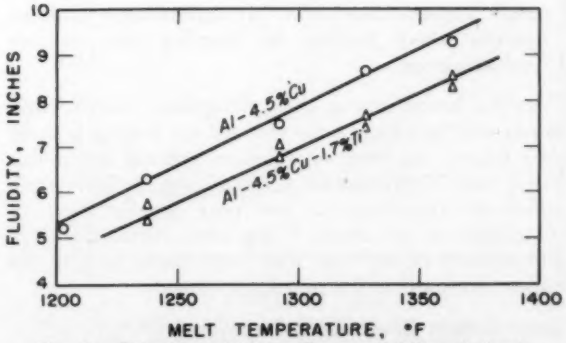



Fig. 4 — Titanium addition effect on fluidity of aluminum-4.5 per cent copper alloy as a function of melt temperature.

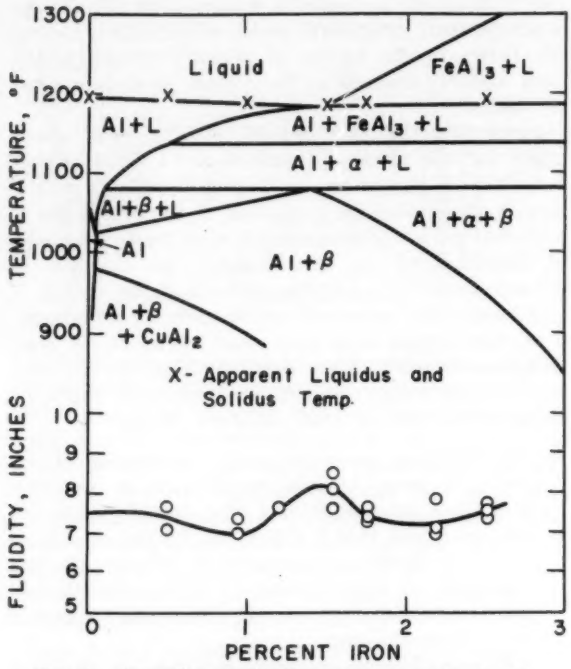



**Fig. 5 — Iron effect on liquidus temperature (top) and fluidity (bottom) of aluminum-4.5 per cent copper alloy. Fluidity tests at 1288 F, 12 in. metal head. Phase diagram from references.[20,21]**

110 F (49 C) at the peritectic and 260 F (127 C) at 3 per cent iron. While this is not always a guarantee of good fluidity, nonetheless, it has been observed in both binary and ternary systems that when the equilibrium temperature range of solidification is small, fluidity is likely to be high.[6, 9, 13, 14]

3) As the peritectic point is approached with increasing iron content, gradual changes in the mechanism of solidification (and hence fluidity) may be expected from the gradual changes in phase diagram relationships (such as temperature range of solidification, item 2). Above the peritectic point a major change in the mechanism of solidification occurs. The primary crystals that form are no longer a solid solution of aluminum and copper but are now the intermetallic compound $FeAl_3$. The heat of fusion of $FeAl_3$ is considerably higher than that of aluminum,[3] and so might affect fluidity by slowing the rate of solidification.

In this investigation, iron, manganese, cobalt, chromium and beryllium were selected for testing because they possess (or were expected to possess) a liquidus valley near aluminum-4.5 per cent copper. They were added to aluminum-4.5 per cent copper alloy in quantities up to about 3 per cent. Adjustments in composition of the base alloy were made to hold the copper constant at 4.3 to 4.5 per cent.

### *Iron, Cobalt, Chromium, Beryllium*

Iron was found to decrease fluidity of aluminum-4.5 per cent copper alloy slightly until approximately 1.0 per cent was added (Fig. 5). Iron additions near the peritectic resulted in somewhat increased fluidity, probably for one or more of the reasons discussed above. Above 1.6 per cent iron had no influence on fluidity. The effect of manganese on fluidity was generally similar to that of iron, as would be expected from the similarity of their ternary phase diagrams, Fig. 6.

In the case of cobalt (Fig. 7) a ternary diagram is not available, but judging from the aluminum-cobalt binary, a liquidus valley should be present at about one per cent cobalt. With cobalt additions higher than this, the primary phase should be $Co_2Al_9$ rather than pure aluminum. Figure 7 shows the substantial increase in fluidity obtained at about one per cent cobalt, undoubtedly due to the presence of the liquidus valley and to the high heat of fusion of $Co_2Al_9$.[3]

From binary phase diagram relationships, it would be expected that chromium would also result in a liquidus valley (at about 0.5 per cent chromium). However, chromium has only a slight effect on the fluidity of 195 alloy, Fig. 8. Beryllium, in quantities of one to 2 per cent significantly improves fluidity of 195 alloy (Fig. 9). It appears likely that this is due to the large decrease in liquidus temperature (30 F at 1.5 per cent Be), although beryllium may also exert other effects on the mechanism of solidification.

### *Silicon, Magnesium, Calcium*

Elements other than those discussed above which were studied herein included silicon, magnesium and calcium. Silicon was added (Fig. 10) because it is generally thought to promote fluidity. However, in

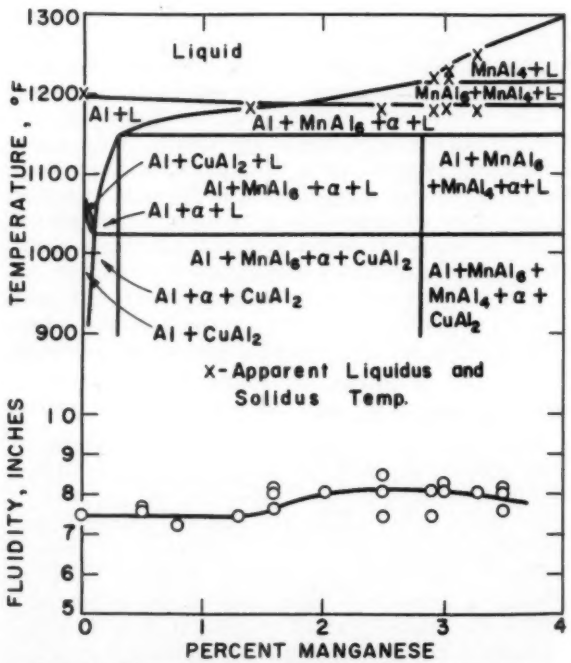



**Fig. 6 — Manganese effect on liquidus temperature (top) and fluidity (bottom) of aluminum-4.5 per cent copper alloy. Fluidity tests at 1288 F, 12 in. metal head. Phase diagram from references.[20,23]**

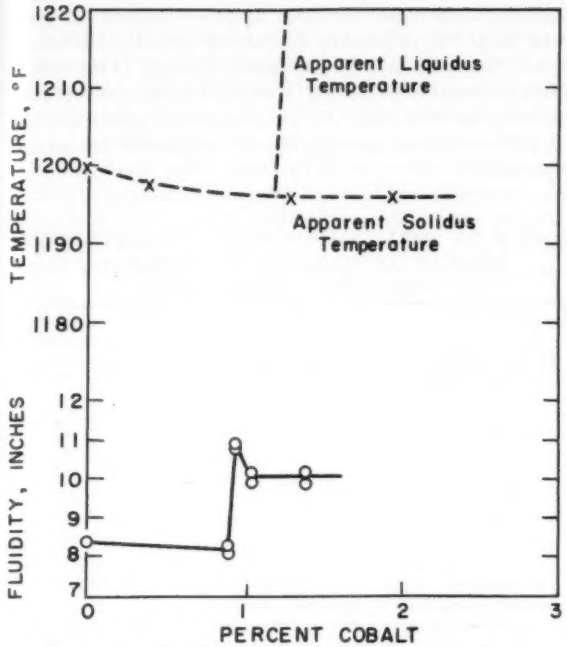



Fig. 7 — Cobalt effect on liquidus temperature (top) and fluidity (bottom) or aluminum-4.5 per cent copper alloy. Fluidity tests at 1328 F, 12 in. metal head.

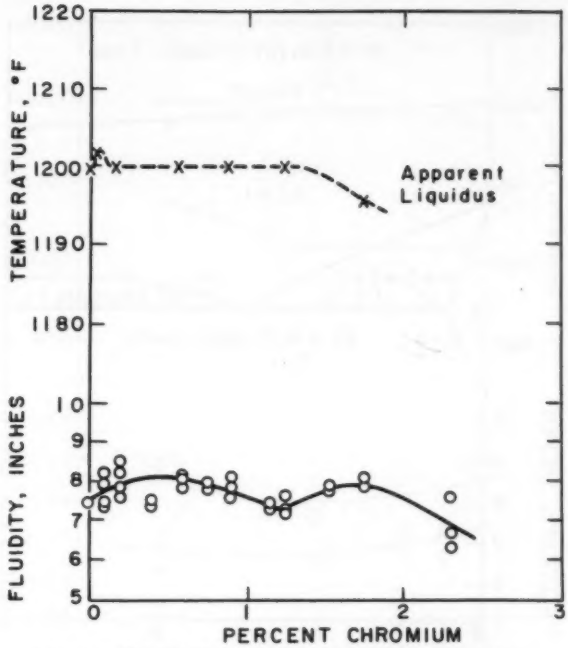



Fig. 8 — Chromium effect on liquidus temperature (top) and fluidity (bottom) of aluminum-4.5 per cent copper alloy. Fluidity tests at 1288 F, 12 in. metal head.

spite of the fact that silicon significantly lowers the liquidus temperature of aluminum-4.5 per cent copper alloy, no improvement in fluidity was found for amounts as high as 2 per cent.

Magnesium and calcium were selected for test because they are highly reactive and should affect fluidity of aluminum by changing the surface characteristics of the flowing stream. Beryllium, mentioned

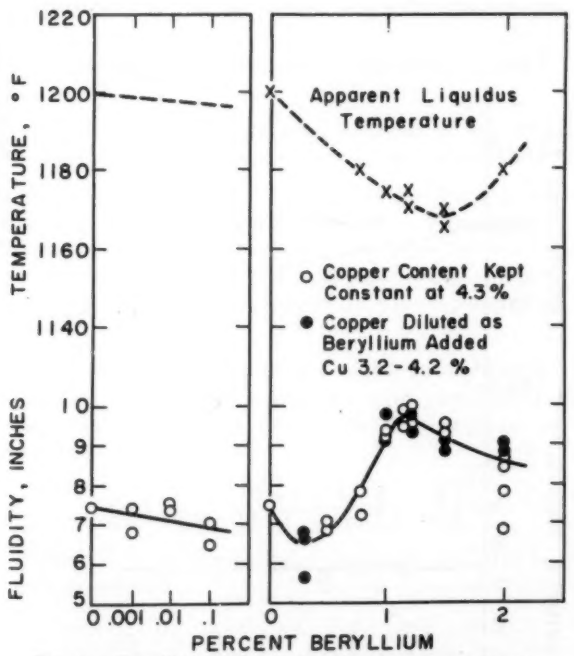



Fig. 9 — Beryllium effect on liquidus temperature (top) and fluidity (bottom) of aluminum-4.5 per cent copper alloy. Fluidity tests at 1288 F, 12 in. metal head. Magnesium was present as an impurity in amounts approximately equal to one-half the beryllium addition (Table 1).

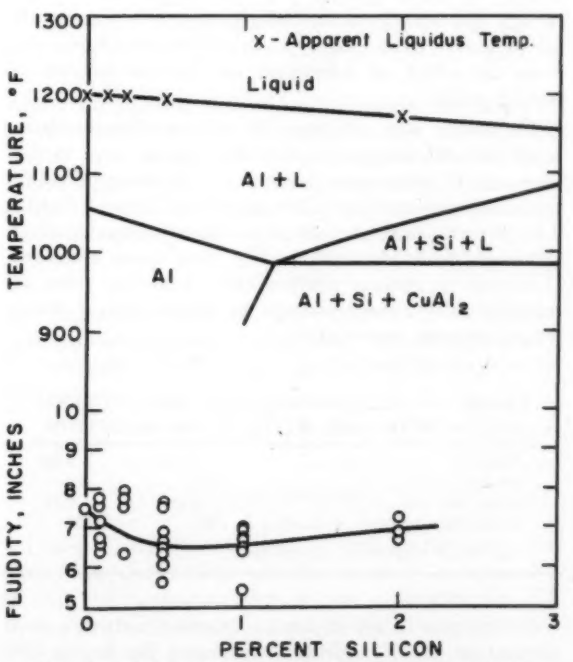



Fig. 10 — Silicon effect on liquidus temperature (top) and fluidity (bottom) of aluminum-4.5 per cent copper alloy. Fluidity tests at 1288 F, 12 in. metal head. Phase diagram from reference.[20]

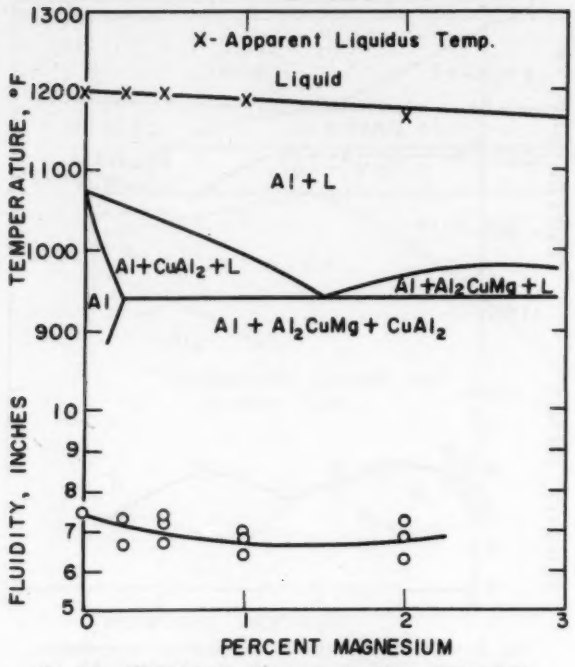



**Fig. 11 — Magnesium effect on liquidus temperature (top) and fluidity (bottom) of aluminum-4.5 per cent copper alloy. Fluidity tests at 1288 F, 12 in. metal head. Phase diagram from references.[20,22]**

earlier, was studied carefully in amounts up to 0.1 per cent, because it also was thought likely to affect the surface tension and/or surface film on flowing aluminum. Magnesium and calcium had little effect on fluidity of aluminum-4.5 per cent copper alloy (Figs. 11, 12), and beryllium had no greater effect than would be expected from phase diagram relationships; i.e., from the effect of beryllium on the mechanism of solidification.

To assure that changes in surface characteristics were not affecting fluidity in this system, motion picture records were made of the aluminum-copper alloy with and without beryllium addition, Table 1. Initial flow velocity was the same for both alloys studied, although fluid life and fluidity were quite different. If change in surface tension and/or surface films affected fluidity, some change in initial flow velocity would also be expected.[2]

**TABLE 1 — ALUMINUM-4.5 PER CENT COPPER ALLOY WITH AND WITHOUT Be* ADDITION**

| Alloy, % | 1.2 Be | 0 Be |
|---|---|---|
| Fluidity, cm | 24.5 | 19 |
| Fluid life, sec | 0.28 | 0.17 |
| Initial flow velocity, cm/sec | 160 | 160 |

*at 1288 F with 12 in. metal head

In the case of all alloying elements studied, except chromium, small additions decreased fluidity of 195 alloy. In most cases, alloys which decreased the liquidus temperature of 195 alloy most sharply also decreased fluidity most markedly (Table 2); larger additions increased fluidity. It is of interest to note that most alloys affected fluidity of 195 alloy in somewhat the same fashion as copper (Fig. 13). This is further indication that the various elements decreased fluidity by changing the mechanism of solidification, much as does increasing copper content in the binary system.[1,2]

**TABLE 2 — SMALL AMOUNTS OF ADDED ELEMENTS EFFECTS ON FLUIDITY OF ALUMINUM-4.5 PER CENT COPPER ALLOY**

| Element added | Decrease in liquidus Temp., C/per cent* | Decrease in fluidity, cm/per cent |
|---|---|---|
| Be | 13 | 12 |
| Si | 7 | 4 |
| Mg | 7 | 2.5 |
| Fe | 4 | 2 |
| Cu | 4 | 1.5 |
| Ca | 3 | 0 |
| Co | 2 | 0 |
| Mn | 2 | 0 |
| Cr | 0 | −6 |

*As measured in this work; decrease is initial slope of measured apparent liquidus temperature, Figs. 5 - 13.

## MATHEMATICAL ANALYSIS OF FLUIDITY OF ALLOYS

### *Equation and Qualitative Comparison with Experiment*

Mathematical analyses of fluidity have been made for pure metals and for alloys such as steel, and are reported in the literature.[6,15,16] However, heretofore no satisfactory analysis has been developed for mushy freezing alloys such as aluminum-4.5 per cent copper

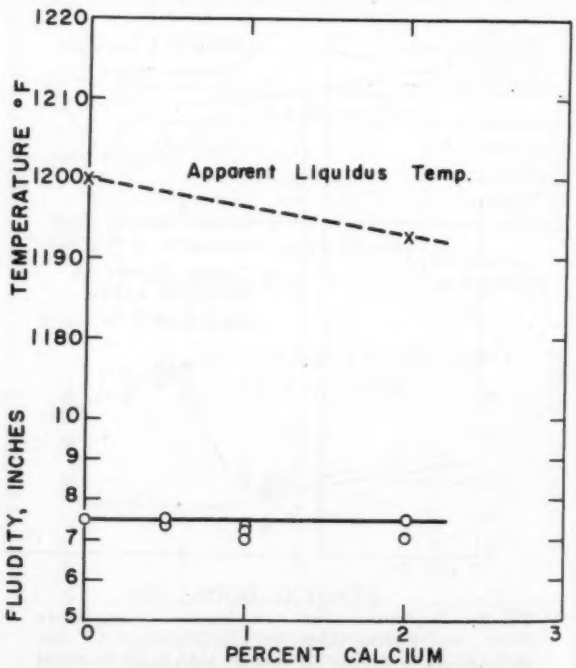



**Fig. 12 — Calcium effect on liquidus temperature (top) and fluidity (bottom) of aluminum-4.5 per cent copper alloy. Fluidity tests at 1288 F, 12 in. metal head.**

alloy. Such an analysis is presented in detail in the Appendix, and is based on a model for flow behavior of mushy alloys that assumes (1) solid particles form during flow in a fluidity channel and travel downstream with the liquid, (2) flow stops when the mean solid concentration near the flow tip reaches a certain value (critical solid concentration) and (3) flow velocity is constant until flow stops. These assumptions agree well with current understanding of the process of fluid[1,2] flow.

The basic equation derived in the Appendix expresses fluidity ($L_f$) in terms of metal and mold variables:*

$$L_f = \frac{A \rho V (k H_f + C \Delta T)}{Sh (T - Tr)} \left(1 + \frac{B}{2}\right) \qquad (1)$$

where:

$$B = \frac{h \sqrt{\Pi \alpha \Delta X}}{k' \sqrt{V}}$$

Examination of the individual terms shows that equation (1) predicts these effects in fluidity test results (in agreement with experimental results):

1) Fluidity increases linearly with the increase of superheat.
2) Fluidity is not zero at zero superheat.
3) Fluidity increases with increasing heat of fusion.
4) Fluidity increases with velocity of flow.

This (4) is generally proportional to the square root of pressure head ($V \alpha \sqrt{P}$). When $B$ is small, fluidity also is proportional to the square root of pressure head ($L_f \alpha \sqrt{P}$), and when $B$ is large fluidity is proportional to the fourth root ($L_f \alpha P^{1/4}$). In intermediate cases, the fluidity may be expressed approximately by $L_f \alpha P^n$ where $n$ is between ½ and ¼. For aluminum-tin alloys tested in the vacuum fluidity apparatus, $n$ has been determined experimentally to be approximately 0.3.[1]

5) Fluidity is proportional to the ratio of the cross-sectional area to the circumference of the channel.

This (5) was also found to be true for fluidity tests made in glass tubes of various sizes. Equation (1) indicates that the fluid life increases with an increase in superheat but decreases with an increase in pressure head. This is also in agreement with experimental results.[1,2,7]

### *Numerical Comparison with Experimental Results*

Equation (1) can be compared with experimental results in several ways. One is for the condition of fluidity at zero superheat:

$$L_{f(\Delta T=0)} = \frac{A \rho V k H_f}{Sh (T - Tr)} \left(1 + \frac{B}{2}\right) \qquad (2)$$

Another is by differentiating equation (1) with respect to superheat:

$$\frac{\delta L_f}{\delta \Delta T} = \frac{A \rho V C}{Sh (T - Tr)} \left(1 + \frac{B}{2}\right) \qquad (3)$$

and comparing the result with the slope of the curve of fluidity versus superheat.

*Definitions of symbols given in the Appendix.

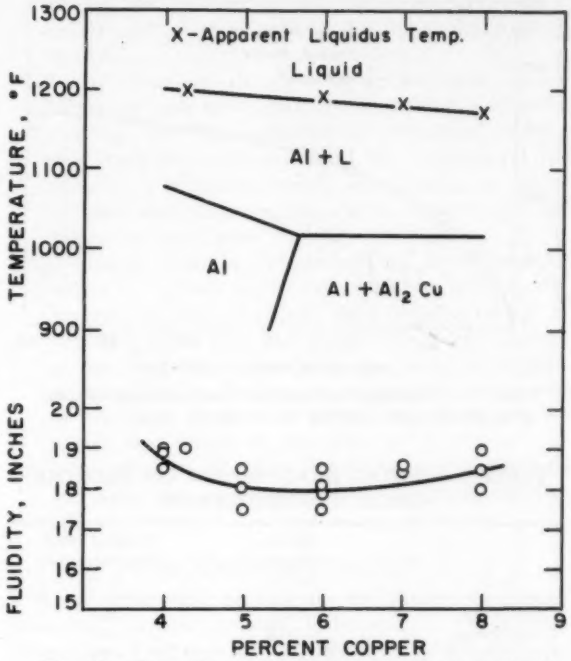



**Fig. 13 — Increasing (and decreasing) copper content effect on liquidus temperature (top) and fluidity (bottom) of aluminum-4.5 per cent copper alloy. Fluidity tests at 1288 F, 12 in. metal head.**

In solving the above equations, all terms were known or measured except k, h and ΔX. However, combining Equations (2) and (3), $k$ can be expressed in quite simple form

$$k = \frac{C}{H_f} \cdot \frac{L_{f(\Delta T=0)}}{\frac{\delta L_f}{\delta \Delta T}} \qquad (4)$$

where the specific heat $C$ and the heat of fusion $H_f$ of the metal are known, and the two other factors in the right hand side of the equation can be determined experimentally. In fact, Equation (4) is quite general. It can be shown that equation (4) can be derived from a general heat flow formula

$$q = S.f(\Theta) \qquad (5)$$

and Equations (17) and (24) can be derived without making assumptions concerning the mechanism of heat transfer. From these, the following can also be derived

$$L_f \alpha \frac{A}{S} \qquad (6)$$

$$\frac{\delta L_f}{\delta \Delta T} = \text{constant} \qquad (7)$$

Using Equation (4) and experimental data obtained in the course of the research program,[2,3] critical solid concentration $k$ was computed for 195 alloy (for glass tubes and sand molds). Data are listed in Table 3 and plotted in Fig. 14.

It is of interest that the critical solid concentration increases with increasing metal head, but that a limit appears to be reached at about 35 per cent solid.

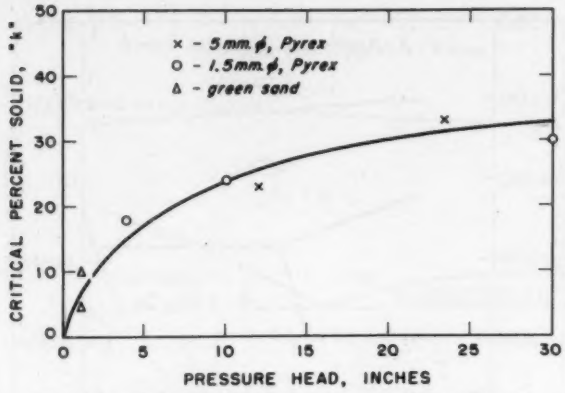



Fig. 14 — Calculated critical per cent solid at the tip of a fluidity test casting vs. pressure head.

TABLE 3 — METAL HEAD EFFECT ON CRITICAL SOLID CONCENTRATION

| Mold | Metal Head, in. Al | Critical Solid Concentration, k, % |
|---|---|---|
| Glass (5 mm) | 12 | 23 |
| Glass (5 mm) | 23.5 | 33 |
| Glass (1.5) | 4 | 18 |
| Glass (1.5) | 10 | 24 |
| Glass (1.5) | 30 | 30 |
| Sand | (effective 2 in.) | 5 (approx.) |

TABLE 4 — VALUES OF TEST VARIABLES USED IN CALCULATING h COEFFICIENT IN FLUIDITY TESTS

| Variable | Vacuum Test | Sand Mold Test | Units |
|---|---|---|---|
| A/S | 0.125 | 0.127 | cm |
| $\rho$ | 2.4 | 2.4 | gms/cm$^3$ |
| T | 700 | 700 | ° C |
| Tr | 20 | 20 | ° C |
| C | 0.26 | 0.26 | Cal/gm° C |
| $H_f$ | 94 | 94 | Cal/gm° C |
| k′ | $2.7 \times 10^{-3}$ | $1 \times 10^{-3}$ | Cal/sec cm° C |
| $\alpha$ | $6 \times 10^{-3}$ | $2.5 \times 10^{-3}$ | Cm$^2$/sec |

TABLE 5 — VALUE OF HEAT TRANSFER COEFFICIENT EFFECT*

| h, Cal/cm$^{-2}$ sec°C | $L_f$, cm | $\delta L_f / \delta \Delta T$, cm/°C |
|---|---|---|
| 0.020 | 58.1 | 0.555 |
| 0.040 | 32.2 | 0.302 |
| 0.060 | 23.2 | 0.218 |
| 0.080 | 18.8 | 0.176 |
| 0.100 | 16.1 | 0.151 |

*Effect on fluidity at 1350 F (732 C) and on slope of fluidity vs. temperature curve for aluminum-4.5% copper alloy in silica sand molds.

TABLE 6 — HEAT TRANSFER COEFFICIENT h EFFECT ON FLUIDITY

| cal/cm$^{2h}$ sec°C | $L_f$ (T = 1350 F), cm | Change,* % |
|---|---|---|
| 0.015 | 76.5 | +200 |
| 0.053 | 25.5 | Standard |
| 0.071 | 20.4 | −20 |
| 0.110 | 15.3 | −40 |

*Value for standard green sand, uncoated mold taken as standard.

Further increases in pressure do not increase the amount of solid that can be moved. This is understandable in view of data cited earlier on various types of slurries which show the viscosity of these slurries increases very rapidly at a given consistency (usually in the range of 20 to 40 per cent solid).

Another factor which may affect $k$ is solid particle size. With a constant pressure head, any change in $k$ will change $L_f$ (equation 2) but not $\delta L_f / \delta \Delta T$ (equation 3). Therefore the straight line of fluidity versus temperature will be shifted when $k$ is changed. This kind of shift was observed when titanium was added to the 195 alloy. $k$ for the alloy with titanium was found to be 19 per cent, while that for the 195 alloy was 23 per cent.

Subsequent calculations comparing theory and experiment were made using the values for $k$ predicted by Fig. 14, and assuming $\Delta X$ to be one cm. Values of other test variables are given in Table 4. One calculation was to determine the heat transfer coefficient $h$ at the mold-metal interface, for fluidity tests in glass tubes, using equations (2) and (3) independently. This was done for experiments at several different pressures, or metal heads (experimental data shown in Figs. 2 and 3). Values of $h$ obtained fell between 0.15 and 0.30 cal/cm$^2$ sec°C, which is the general range to be expected.[7]

In the case of the spiral fluidity test in green sand mold used in other parts of this research, calculations of $h$ were made from experimental observations of fluidity at 1350 F (732 C) using equation (2), and from experimental observation of the slope of the fluidity curve versus temperature using equation (3). Analysis was not made using data at zero superheat, because any temperature losses in the runner system would then have too great effect on calculated heat transfer coefficients.

In the case of spirals poured in uncoated green sand molds, the observed value for fluidity at 1350 F (732 C) is 25.5 cm, and the slope of the curve of fluidity versus temperature is 0.224 cm/°C. These observations indicate the $h$ coefficients to be approximately 0.05 cal/sec °C (Table 5).

As reported earlier, certain mold coatings such as hexachloroethane and carbon black increase fluidity in green sand molds by as much as a factor of three.[2,4] On the other hand, the use of certain core sands instead of green sand results in decreases of fluidity from 20 to 40 per cent.[3] Changes in the heat transfer coefficient $h$ that would account for such changes in fluidity were calculated, and are shown in Table 6. $k$ was assumed to be 5 per cent and $\alpha C = 2.5 \cdot 10^{-3}$ cm$^2$/sec°C.

The heat transfer coefficient need be reduced by less than a factor of 4 to result in a 200 per cent increase in fluidity (triple fluidity). Doubling $h$ results in as much as 40 per cent decrease in fluidity. It is of interest that the heat transfer coefficient need be affected only for an extremely short time to influence fluidity. For example, in a typical test in a fluidity spiral, metal velocity $V$ is 80 cm/sec, and the choking zone ($\Delta X$) is one cm. The average time the tip of the metal is in contact with any single portion of the mold is $1/80 = 0.013$ sec.

## CONCLUSIONS

1. Grain refinement by addition of 0.17 per cent titanium decreases fluidity slightly (10 to 13 per cent).
2. Elements added to aluminum-4.5 per cent copper alloy (or present as impurities) in amounts up to about 3 per cent have a relatively small effect on fluidity. Elements tested to reach this conclusion were iron, manganese, cobalt, chromium, beryllium, silicon, magnesium, calcium and copper. Maximum improvements in fluidity observed were 25 per cent at one per cent beryllium, and 28 per cent at one per cent cobalt.
3. Small amounts of solute elements added to aluminum-4.5 per cent copper alloy generally decrease fluidity slightly. Elements which cause the largest decrease in fluidity are those that decrease the liquidus temperature most markedly. Effects of small amounts of alloying elements on fluidity are quite small, and it therefore appears likely that the impurities normally present in 195 alloy (high purity or commercial purity) do not affect fluidity of the alloy significantly.
4. Increased amounts of alloying elements tend to increase fluidity. In particular, fluidity was found to increase at the composition of liquidus valleys (at or near ternary compositions where the liquidus temperature was a minimum). These increases in fluidity are attributed to (a) increased superheat resulting from the lower liquidus temperature, (b) narrower temperature range of solidification and/or (c) change in the nature of primary crystals which formed.
5. In all probability, certain of the elements added (notably magnesium, calcium, beryllium), affected surface tension and/or surface films. However, any such changes did not affect fluidity significantly.
6. Experimental results confirm the theory that solidification of mushy alloys such as aluminum-4.5 per cent copper alloy, and hence cessation of flow in fluidity tests, occurs primarily by precipitation of fine grains near the leading tip of the flowing stream.
7. On the basis of certain simplifying assumptions, including that of mushy solidification, an equation was derived to express the fluidity of an alloy in terms of metal and mold variables.
8. The features of fluidity of alloys are accounted for qualitatively by the equation:
   (a) Linearity between fluidity and superheat.
   (b) Finite fluidity at zero superheat.
   (c) Effect of heat of fusion on fluidity.
   (d) Effect of metal head on fluidity ($L_f \alpha p^n$).
   (e) Proportional dependence of fluidity on the ratio of cross-sectional area to circumference.
   (f) Decrease in fluid life with increase in pressure head.
9. The equation is in reasonable agreement quantitatively with fluidity data obtained experimentally with the vacuum fluidity tester and in the sand mold fluidity spiral, assuming heat transfer coefficients to be 0.30 cal/cm$^2$ sec° C, and 0.05 cal/cm$^2$ sec° C in the glass and sand mold, respectively, and assuming solidification occurs primarily in a narrow zone at the tip of the flowing stream.
10. Comparison of theory with experiment shows improvements in fluidity obtained by use of mold coatings can be explained by reduction of the heat transfer coefficient $h$ at the mold-metal interface. A probable reason why fluidity is lower in core sand than green sand is that $h$ may be higher in core sand molds.
11. A simple formula was derived to determine the value of critical solid concentration $k$ from experimental data (critical solid concentration is per cent solid near the tip of the fluidity spiral at the instant flow ceases). $k$ was found to depend on metal head, but was independent of the size of the test channel. The maximum value observed was about 35 per cent.
12. The effect of titanium in decreasing fluidity of 195 alloy can be explained by its grain refining effect.

## ACKNOWLEDGMENT

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## APPENDIX

## MATHEMATICAL ANALYSIS OF FLUIDITY OF ALLOYS

### *Assumptions*

1) All the solid particles formed in the fluidity channel flow downstream with liquid.
2) Flow stops when the mean solid concentration in a zone near the flow tip reaches a certain value (critical solid concentration).
3) Velocity of flow is constant until the flow stops.
4) In considering heat flow, the mold wall is regarded as flat and infinitely thick. Heat resistance at the metal-mold boundary and heat flow in the mold are considered. The correctness of these assumptions are examined below.

Consider liquid metal with some superheat entering a fluidity channel. After the metal reaches the liquidus temperature, solidification takes place along the channel. Since the alloy solidifies over a wide range of temperature, it may be assumed that all the solid is floating as particles in the liquid instead of clinging to the mold wall. These particles will flow downstream together with the liquid. At the tip of flow will be found the highest concentration of particles, since solidification rate is highest at this point because of the freshness of the mold wall.

It has been established experimentally that flow of an alloy with a wide range of solidification temperature in a fluidity channel stops when it is choked with accumulated solid particles at or near the tip.[1,2,4] This happens when the solid concentration is much less than 100 per cent. From studies on quartz and graphite particles in an organic liquid,[11] clays[17] and pulp,[18] two general conclusions can be drawn concerning the flow behavior of a mixture of liquid and solid particles:

1) There is a certain critical concentration of solid particles at which viscosity of the mixture and hence resistance to flow increases rapidly to a high value.
2) The finer the particles, the lower is the critical concentration. Probably the critical concentration also depends upon particle shape, flow velocity, channel size, etc.

In applying this concept of critical solid concentration to our fluidity problem, it is reasonable to consider the mean solid concentration in a certain finite zone near the tip, since mixing of solid and liquid takes place. For the present, this zone will be called the choking zone.

It should be possible to express the pressure drop in flow as a function of mean solid concentration at the tip and perform some integral calculations to get flow velocity as a function of time. Pressure, or head loss, due to wall friction along the channel, should be taken into consideration to account for changes of velocity with time. However, for the sake of simplicity it is assumed here that flow velocity is constant until the flow stops suddenly when the mean solid concentration in the choking zone reaches the critical value. This is, in fact, nearly what is observed experimentally.[2]

To examine the assumptions employed in heat flow calculation, the vacuum fluidity test in a glass tube is used as a model. First, the validity of the assumption of a semi-infinite mold is checked. Temperature distribution in a concave cylinder (radium $R$) heated from inside in a short time is expressed by[19]

$$\frac{T - Tr}{Tm - Tr} = \sqrt{\frac{R}{r}}\left(1 - \operatorname{erf}\frac{r - R}{2\sqrt{\alpha\Theta}}\right) \qquad (8)$$

To find the temperature rise on the outside surface of the tube immediately after completion of a test the following typical values might be substituted in equation (8)

$R = 0.25$ cm, $r = 0.35$ cm, $\alpha = 6 \times 10^{-3}$ cgs and $\Theta = 0.2$ sec.

Then

$$\frac{T - Tr}{Tm - Tr} = 0.04$$

Since the temperature rise on the outside surface is only 4 per cent, this tube can be assumed to be an infinitely thick pipe. In a sand mold, where the mold is much thicker, the validity of the assumption is apparent.

Next, the assumption of plane mold wall is checked. Disregarding the heat resistance at the boundary, heat absorption by a semi-infinite plane is[19]

$$\frac{Q}{A} = \frac{2k'(Tm - Tr)}{\sqrt{\Pi\alpha}}\sqrt{\Theta} \qquad (9)$$

while that by a semi-infinite concave cylinder is[19]

$$\frac{Q}{A} = \frac{2k'(Tm - Tr)}{\sqrt{\Pi\alpha}}\left(\sqrt{\Theta} + \frac{\sqrt{\Pi\alpha}}{4R}\Theta\right) \qquad (10)$$

The relative error due to use of equation (9) instead of the equation (10) is

$$\frac{\frac{\sqrt{\Pi\alpha}}{4R}\Theta}{\sqrt{\Theta}} = \frac{\sqrt{\Pi\alpha\Theta}}{4R} \qquad (11)$$

Substitution of numerical values for the variables gives the error to be 6 per cent, which is reasonably small. As noted in the test, the time interval that should be considered is much shorter than the overall test time taken for the above calculations ($\Theta = 0.2$ sec). As a result, errors due to both the above assumptions are even smaller than those calculated above.

*Analysis*

Considering the heat resistance at the mold-metal interface, the rate of heat transfer from metal to mold in a unit length of the channel is

$$q_1 = S\,h\,(T - Tsc) \qquad (12)$$

where the boundary temperature Tsc is unknown. On the other hand, the rate of heat flow in the mold is

$$q_2 = \frac{S\,k'\,(Tsc - Tr)}{\sqrt{\Pi\,\alpha}}\,\frac{1}{\sqrt{\Theta}} \qquad (13)$$

Since $q_1$ must be qual to $q_2$, Tsc can be eliminated by combining (12) and (13), and the heat flow rate is

$$q = \frac{S\,h\,(T - Tr)}{1 + \frac{h\sqrt{\Pi\,\alpha\,\Theta}}{k'}} \qquad (14)$$

It is seen that the type of heat flow is determined by the factor $h\sqrt{\Pi\,\alpha\,\Theta}/k'$. When $h\sqrt{\Pi\,\alpha\,\Theta}/k'$ is small (much less than one), equation (14) becomes

$$q = S\,h\,(T - Tr) \qquad (15)$$

Thus the heat flow rate is controlled by the boundary resistance and is constant all the time. On the other hand, when $h\sqrt{\Pi\,\alpha\,\Theta}/k'$ is large (greater than one), equation (14) becomes

$$q = \frac{S\,k'\,(T - Tr)}{\sqrt{\Pi\,\alpha}}\,\frac{1}{\sqrt{\Theta}} \qquad (16)$$

and the heat flow rate, controlled by the heat flow process in the mold, decreases with time. When $h\sqrt{\Pi\,\alpha\,\Theta}/k'$ has a value the order of one, then equation (14) must be used as such and the rate of heat flow is controlled both by the boundary and the mold.

Starting from the general equation (14), the heat flow rate at a point $x$ cm away from the flow tip is considered. Assuming a constant flow velocity, the time elapsed since the first metal touched the point is

$$\Theta = \frac{x}{V} \qquad (17)$$

Therefore the heat flow rate is

$$q = \frac{S\,h\,(T - Tr)}{1 + \frac{h\sqrt{\Pi\,\alpha}}{k'}\sqrt{\frac{x}{V}}} \qquad (18)$$

Notice that $q$ is a function of the position relative to the flow tip but is not a function of time. We are interested in the mean heat flow rate in the "choking zone" $\Delta X$. That is

$$qm = \frac{1}{\Delta X}\int_o^{\Delta x} q\,dx$$

$$= \frac{1}{\Delta X}\int_o^{\Delta x} \frac{S\,h\,(T - Tr)}{1 + \frac{h\sqrt{\Pi\,\alpha}}{k'}\sqrt{\frac{x}{V}}}\,dx \qquad (19)$$

and

$$qm = S\,h\,(T - Tr)\,\frac{2}{B^2}\{B - \ln(1 + B)\} \qquad (20)$$

where

$$B = \frac{h\sqrt{\Pi\,\alpha\,\Delta X}}{k'\sqrt{V}} \qquad (21)$$

Equation (20) can be approximated with errors less than 10 per cent for any value of $B$ by

$$qm = S\,h\,(T - Tr)\,\frac{1}{1 + \frac{B}{2}} \qquad (22)$$

The total heat flow in a unit length of the channel from time zero to $\Theta$ is simply:

$$Q = S\,h\,(T - Tr)\,\frac{1}{1 + \frac{B}{2}}\,\Theta \qquad (23)$$

The heat to be subtracted from the metal by the time the flow stops is a sum of the superheat plus a part of the heat of fusion

$$Q = A\,\rho\,(k\,H_f + C\,\Delta\,T) \qquad (24)$$

where k is the critical solid concentration (in per cent).

Fluid life is obtained by equating (23) and (24)

$$\Theta_f = \frac{A\,\rho\,(k\,H_f + C\,\Delta\,T)}{S\,h\,(T - Tr)}\left(1 + \frac{B}{2}\right) \qquad (25)$$

and the fluidity is

$$L_f = V\,\Theta_f$$

$$= \frac{A\,\rho\,V\,(k\,H_f + C\,\Delta\,T)}{S\,h\,(T - Tr)}\left(1 + \frac{B}{2}\right) \qquad (1)$$

This is the basic equation expressing the fluidity of an alloy in terms of metal and mold variables.

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A = Mold surface area, cm².
B = $h\sqrt{\Pi\,\alpha\,\Delta X}\,/\,k'\sqrt{V}$
$= L\sqrt{\Pi\,\Delta X}\,/\,\sqrt{C'\,\rho'\,k'\,V}$
C = Specific heat of metal, cal/gm °C.
C′ = Specific heat of mold, cal/gm °C.
$H_f$ = Heat of fusion of metal, cal/gm.
h = Heat transfer coefficient at metal-mold interface, cal/cm² sec. °C.
k = Critical solid concentration, dimensionless fraction (or per cent).
k′ = Thermal conductivity of mold, cal/sec. cm. °C.
$L_f$ = Fluidity, cm.
Q = Total heat entering mold per unit length, cal/cm.
q = Heat flow rate from metal to mold per unit length, cal/cm sec.
R = Radius of mold channel, cm.
r = Radius of concentric cylinders in mold, cm.
S = Circumference of mold channel, cm.
T = Temperature of liquid metal, C.
Tm = Melting temperature, C.
$T_{sc}$ = Mold temperature at interface, C.
Tr = Room temperature, C.
V = Velocity of metal flow, cm/sec.
$\alpha$ = Thermal diffusivity of mold = $k'/C'\rho'$, cm²/sec.
$\Delta X$ = Choking range, cm.
$\rho$ = Density of metal, gm/cm³.
$\rho'$ = Density of mold material, gm/cm³.
$\Theta$ = Time, sec.
$\Theta_f$ = Fluid life, sec.

# ENGINE CASTINGS DEVELOPMENT BY EXPERIMENTAL STRESS ANALYSIS

by L. A. Grotto


## ABSTRACT

**The development of the principle engine castings by experimental stress techniques are discussed. How data are obtained from experimental techniques, using stress coat and strain gages, is shown and correlated with material properties. The test work shows that it is important to design a casting to minimize the assembly stress, as well as the working stress. It also proved that it is not necessary to make a casting heavy to make it strong.**

**Physical testing showed that the strength of the casting could not be predicted from test bars. However, by making the foundryman a member of the team, and working closely with him, it was possible to develop the material so it would satisfy the requirements of easy machining and high strength.**

**Summarizing, the report covers the development of a casting design and its material with the cooperation of the man in the foundry.**


## INTRODUCTION

This discussion will be concerned with the development of castings that make up the crankcase assembly – namely, the crankcase, bearing caps and the cylinder liners.

The principal castings used in the manufacture of high output automotive type diesel engines must combine maximum strength and minimum weight in such a manner to provide acceptable life and reliability. This is quite an order to expect from a structural assembly as indeterminate as that making up the crankcase assembly. Therefore, the design and development of such a casting requires the use of all the experience, knowledge and specialized tools and techniques the designer can muster. Furthermore, an unusual amount of teamwork between the designer and the foundryman is necessary to obtain the required casting for the final successful product.

This assembly is a challenge to the designer. It is difficult, if not impossible to predict structural adequacy from calculations. Furthermore, the material properties are not homogeneous, and the strength depends on such things as mass and cooling rate, as well as composition and hardness. In use, this assembly is subject to high force from assembly, combustion and inertia. These forces, of course, can combine and produce stresses which will cause fatigue in the structure.

This assembly is also a challenge to the foundryman. The crankcase is a difficult casting to produce because thin walls and heavy bosses are combined in a large mass. Furthermore, it must have high strength in the main bearing area, an area of relative slow cooling, and it must be easy to machine, even in areas where the cooling rate is fast, such as the pan rail and the top deck. The bearing caps on the other hand are small, massive castings requiring consistently high strength. The cylinder liners present special problems because the material here must not only have high strength, but also good wear and skuff resistance properties.

## PREVIOUS TESTING METHOD

In the past, the design of this assembly drew heavily on the experience of the designer with prior designs operating under similar loads. Furthermore, tests under full load operation were run for thousands of hours to prove the structural integrity of the assembly. It is obvious that such a method is time consuming if the background of experience is not applicable to new design. Furthermore, the results of such tests are of the "Go or No-Go" variety, and development with this method will probably be "Add To" and "Beef It Up," resulting in unnecessary and excessive weight.

The experimental stress analysis techniques make it possible to determine if a structure is adequate, without operating the assembly for a long period of time. Quantitative values are obtained so that it is possible to know if the assembly is conservative or borderline. Areas of high stress are pinpointed as well as areas of low stress, and a concise knowledge of

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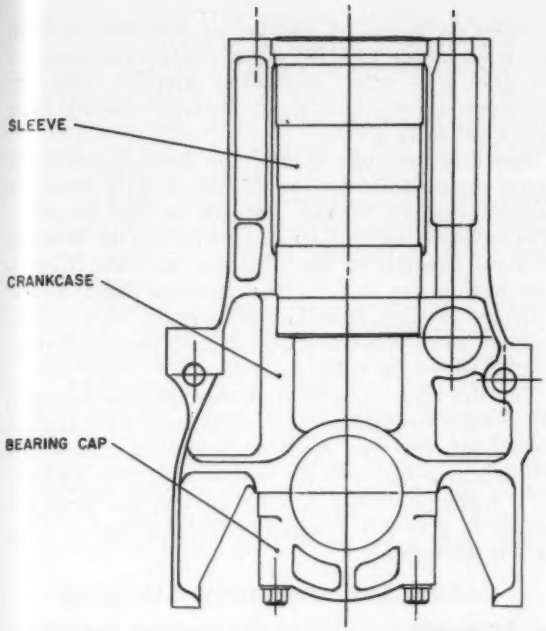



**Fig. 1 — Castings used in diesel engine crankcase assembly.**

how the loads are carried through the structures is obtained. The effect of changes can also be evaluated with concise numbers. This makes it possible to develop a satisfactory structure in less time, without extensive, specific experience.

Figure 1 is a cross-section of the assembly to be discussed, so that the relative location of the parts are apparent. This assembly makes up a large, automotive type high speed diesel engine, which has a displacement of 817 cu in., and develops 375 hp at 2100 rpm. This engine was a radical departure from engines that were previously manufactured by the author's company.

Previous engines were large, heavy, low speed engines of the precup design, which had relatively low firing pressures, while this engine is a light, high speed, high output, turbo-charged, direct injection engine with high firing pressures. This is also the first engine built in the author's company's construction equipment division in which experimental stress analysis techniques were used extensively.

## GRAY IRON STRESS ANALYSIS

Preliminary studies were carried out on gray iron which gave many interesting and surprising results. The first was that test bars and other test specimen cast with the castings had little in common with the casting, even though they were made from the same material at the same time. In general, it was found that the properties of test specimen taken from critical areas of the castings were much lower than the properties determined from a test specimen cast at the same time.

It was not uncommon to find that the test specimen had properties 50 per cent higher in tensile strength, and 100 per cent higher in bending fatigue strength than a specimen taken directly from the casting. Even the microstructures were found to be different, Fig. 2.

The effect of repeated loading on gray iron was also interesting. The first loading produced the typi-

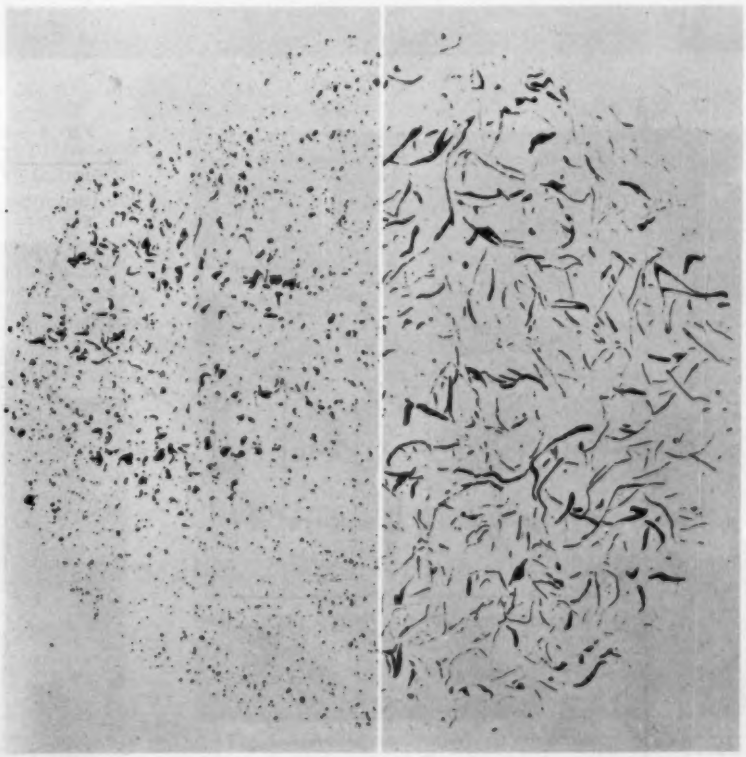


**Fig. 2 — Microstructural difference between test bar and casting. *Left* — Material cast in test bar. *Right* — material in crankcase.**

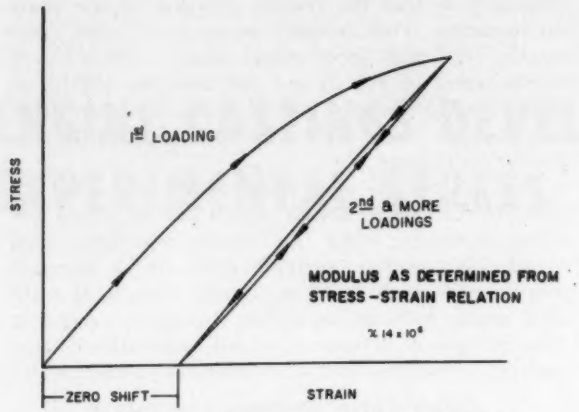



Fig. 3 — Gray iron stress-strain characteristics under repeated fluctuating load.

cal stress-strain curve for gray iron in which the modulus is not a constant. This had been observed many times during loading of a tensile specimen to the ultimate. However, when the load on the specimen was stopped short of the ultimate, released and re-applied, it was found that:

1 During the first loading, the typical gray iron stress-strain curve was observed.
2. As the load was released, the stress-strain path of unloading differed from the stress-strain path of loading, and an apparent zero shift was observed on the strain coordinate.
3. After the first loading, the material appeared to have a constant modulus of elasticity, providing the loading did not cross the zero line. This phenomena is illustrated in Fig. 3.

Based on a limited number of tests the decision was made to use 14 x $10^6$ as the modulus of elasticity for gray iron when evaluating strength, although variations in modulus were observed ranging from 11.5 x $10^6$ to 16 x $10^6$.

Based on literature studies, the decision was made to use the modified Goodman diagram as a means of determining the fatigue strength of the structure. The envelope was constructed by taking the reversed bending strength as one-third the ultimate strength and having the lines of the envelope meet at the ultimate strength. The fatigue strength with a compressive mean stress is taken to have the same range as that where the mean stress is zero.

For the types of gray iron used in this assembly, the fatigue strength at zero mean load was taken at 10,000 psi, and the envelope joined at 30,000 psi, which was considered the ultimate of the material. This is shown on Fig. 4. This analysis has been successful in gray iron castings of this type and is being used extensively.

## CRANKCASE ASSEMBLY CASTINGS

The principal casting of the crankcase assembly is the crankcase. The first crankcases were designed from experience on the large, heavy crankcases then being made. In general, the loads were carried by the walls and bulkheads. Bosses were put wherever needed to accommodate bolts, studs, oil holes, bearings, etc., and the size of these bosses were limited to the minimum needed. This design was satisfactory during the early development of the engine when operating loads were low.

However, as the development of the engine progressed higher outputs were being obtained, and higher loads were being applied to the structure until failures, such as those shown on Figs. 5, 6 and 7

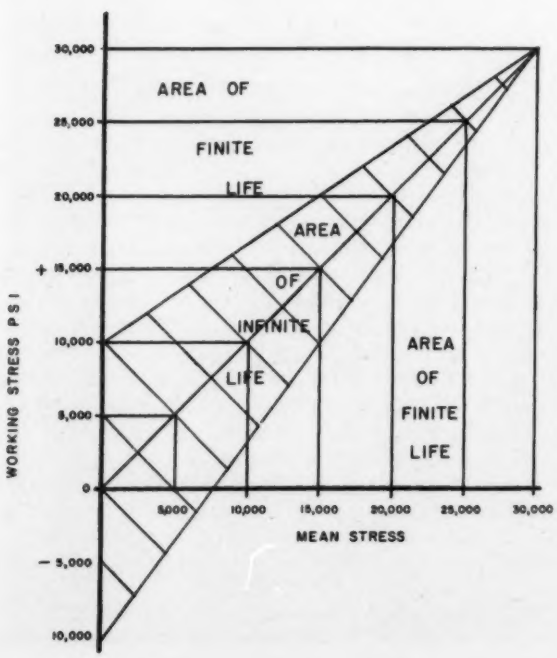



Fig. 4 — Goodman diagram for gray iron.

Fig. 5 — Crankcase failure during operation.



were experienced. The failures appeared to start at the junction of the main bearing capscrew boss and the main bearing bulkhead. Subsequent stress coat tests showed this to be an area of high stress, Fig. 8. Strain gage tests made under operating conditions showed that the working stress in this area increased with horsepower output.

At the high output, the working stress, when combined with the assembly stress was of sufficient magnitude to cause failure, when interpreted on the modified Goodman diagram, Fig. 9.

As a result of this experience, the design was modified so the loads and forces could be carried in as straight and direct a line as possible by bosses and ribs. The transitions in bosses and ribs were minimized, and the bosses around drilled passages were made as generous as possible. Figures 10 and 11 show stress coat results and Fig. 12 shows the strain gaging on the new design. Comparing Fig. 12 with Fig. 5, 6 and 7 makes the design change apparent. Static and dynamic tests showed that the redesign

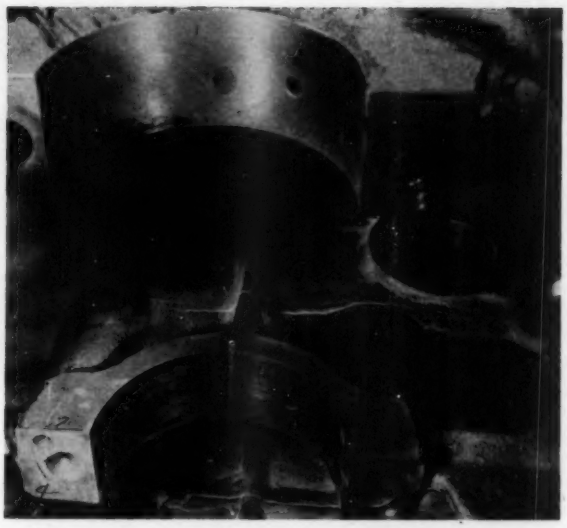


Fig. 6 — Crankcase failure during operation.

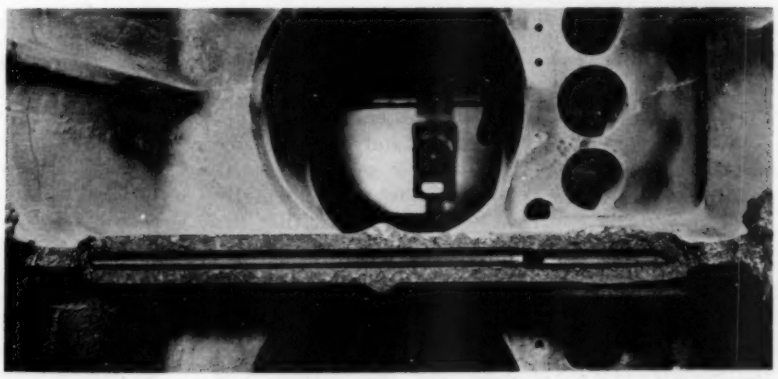


Fig. 7 — Failed section broken out of crankcase.

was satisfactory, and no further problems or failures were experienced in this area during development and final endurance testing prior to production release.

*Production Castings Failure*

However, when the engine was first built in production, failures were experienced, as shown on Figs. 13 and 14. Investigations showed the most significant change between development and production was the source of the crankcase castings. The production castings were obtained from a production foundry where large castings and small castings were made, so that a base iron was required that could be used on the extremes. Furthermore, manufacturing exerted considerable pressure on the foundry to assure castings that were easy to machine.

The drawings from engineering specified only a broad maximum and minimum hardness range, and there was nothing in the drawings to indicate that these castings required a material of a certain minimum fatigue strength in the critical areas.

The fatigue strength of the material in critical areas of these first crankcases was investigated, using

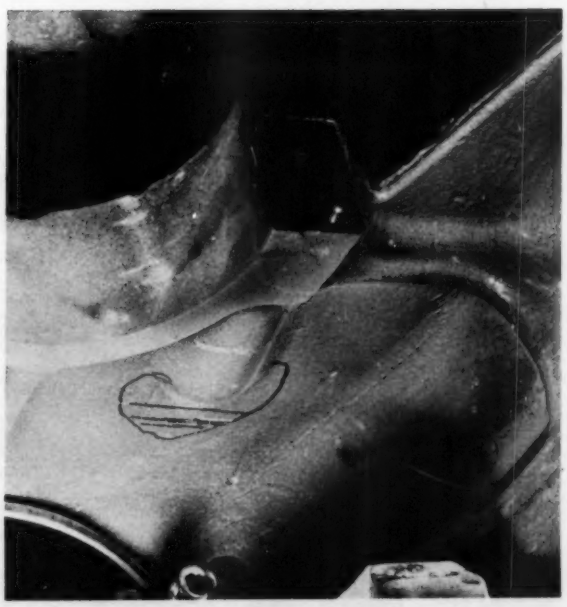


Fig. 8 — Stress coat pattern on crankcase.

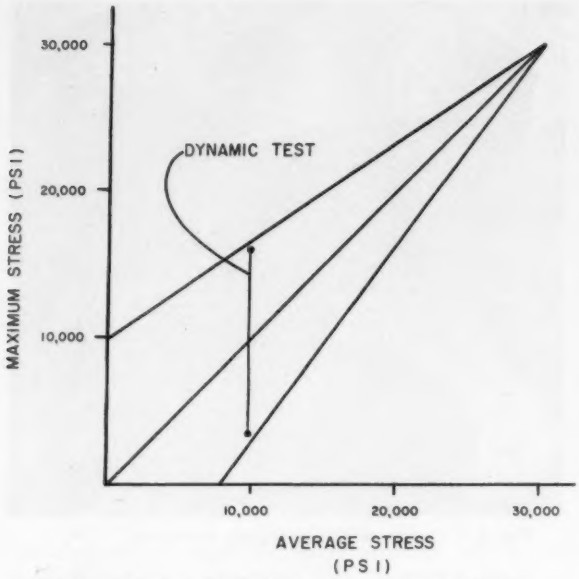



**Fig. 9 — Goodman diagram for stresses in the failure area of crankcase casting.**



**Fig. 11 — Stress coat pattern on redesigned crankcase.**

specimens of the design shown in Fig. 15. These specimen were made from the main bearing cap-screw bosses and oil gallery bosses. The size was determined by the size of the boss, and the diameter was kept as close to the diameter of the boss as possible to minimize any size effect, which incidentally was found to be appreciable. It was found that the crankcases in this lot were made of a material whose fatigue strength was considerably lower than required. Also, the hardness in the critical area was low.

Therefore, meetings were held with personnel from the foundry, manufacturing and engineering. The requirements of the casting, and the effects of not meeting the requirements, were discussed from all viewpoints. Once the requirements were clearly known, the foundry, engineering and manufacturing were able to work as a team in order to meet the necessary requirements without any detrimental effect.

Investigations showed that although scatter does exist, hardness gives the best correlation to fatigue strength. Therefore, a minimum hardness was specified in the center main bearing area after removal of the casting skin. Equipment was installed in the foundry so this could be checked on a 100 per cent basis. Checks made by manufacturing on finished machined cases showed this was an excellent method of quality control.

The foundry also investigated many modifications in materials and casting practices until all the minimum requirements were met or exceeded. Among the notable items investigated was the effect of inoculations of tin, molybdenum, varying percentages of steel scrap in the melt, cooling time in the mold,

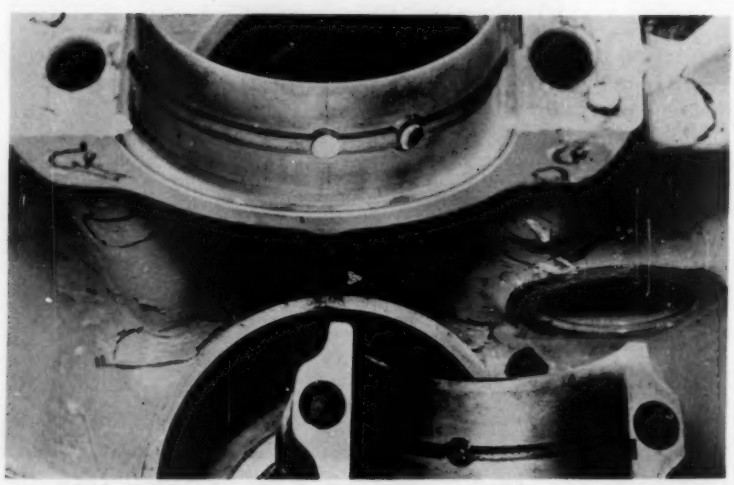


**Fig. 10 — Stress coat pattern on redesigned crankcase.**




Fig. 12 — Strain gage installation on redesigned crankcase.



Fig. 13 — Failure of redesigned crankcase.

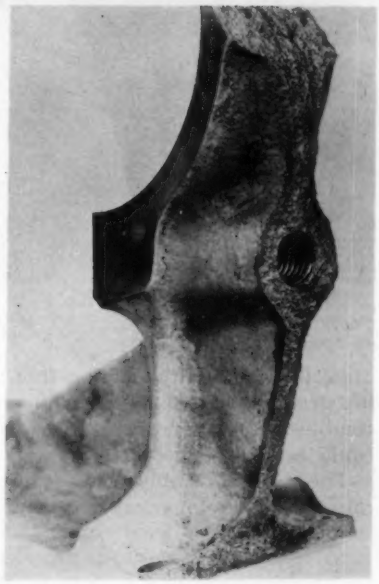


Fig. 14 — Failed section from redesigned crankcase.

and means of minimizing residual stresses in the casting.

## BEARING CAP CASTINGS

The original bearing caps were designed so that they could be cast in a cluster, and then cut into individual caps for machining, Fig. 16. Early in the development of the engine bearing cap failures were experienced, as shown in Fig. 17. Resorting to the "Add To" and "Beef It Up" manner of designing, the section modulus of the bearing cap across the area of failure was increased by more than 50 per cent. This change increased the weight of the cap by more than 20 per cent, and created a possible manufacturing problem because these bearing caps extended below the pan rail of the crankcase.

This heavier cap was found to have about the same life as the previous design, the failures being identical in appearance. Strain gages were installed in the failure area, and the stresses measured during various engine operations. It was found that the stresses increased with horsepower, and that the working stresses alone were not of sufficient magnitude to

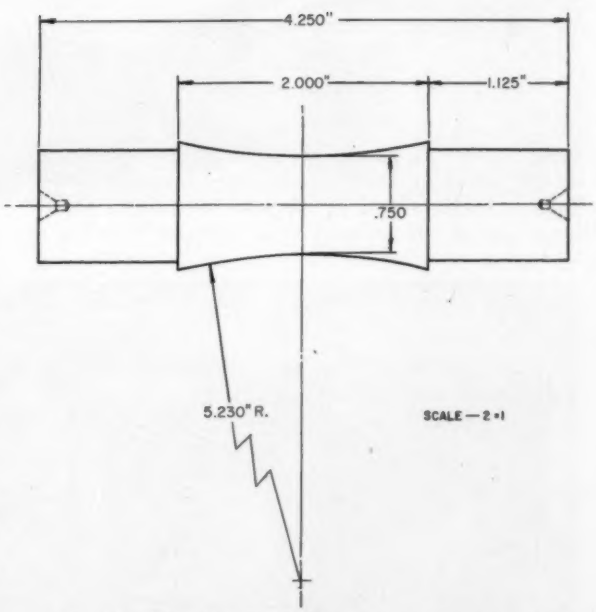



Fig. 15 — Modified SF-10-R fatigue test specimen for testing specimen taken from crankcase castings.

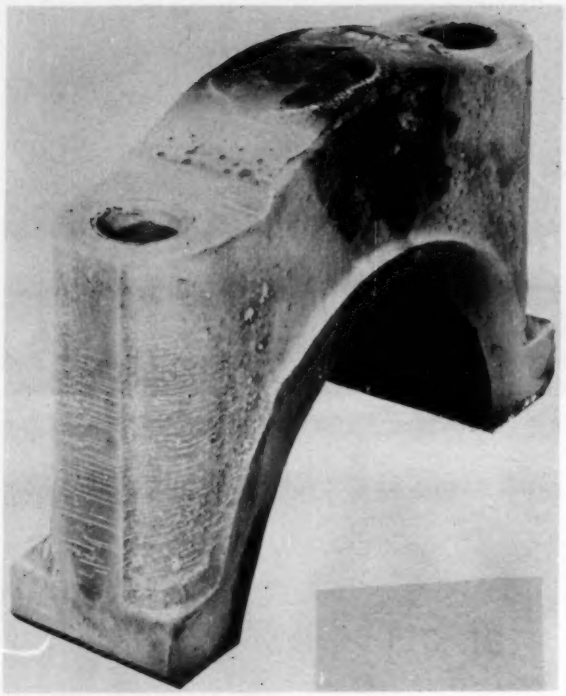


Fig. 16 — Original bearing cap.

cause failure. Additional tests showed that the assembly stress was high, and when the assembly stress was combined with the working stress, failure from fatigue could be expected.

The high assembly stress was a bending stress caused by unsymmetrical stiffness around the bearing cap capscrew bore. Increasing the section modulus, of course, did nothing to relieve this condition, and from the tests it was found that while this change did reduce the working stresses slightly the assembly stress was increased slightly, so no overall improvement was obtained.

A series of tests using stress coat was conducted, and areas of low stress and high stress were determined. These tests further verified that the high assembly stress was due to bending resulting from areas of unsymmetrical stiffness. Material was removed from the area of flow stress in such a manner as to make the stiffness around the bearing cap capscrew hole as symmetrical as possible. This reduced the weight by almost 20 per cent and increased the fatigue strength by 20 per cent. The effect of bearing cap design on bearing cap assembly stresses is graphically illustrated in Fig. 18.

The weight was further reduced by reducing the height of the bearing cap and changing the material to ductile iron. The combination design and material change increased the fatigue strength of the bearing cap to almost 40 per cent more than the original. A chart showing the relative designs with weights and safety factors is shown in Fig. 19. Figure 20 shows the bearing cap design evaluation.

## CYLINDER LINER

The cylinder liner was also the subject of considerable experimental stress analysis work. Failures were obtained, as shown on Fig. 21, with the hours to failure being relatively few hours. Stress coat tests showed that the forces from the cylinder head assembly caused patterns in the fillet below the pilot bore,



Fig. 17 — Bearing cap failure during operation.

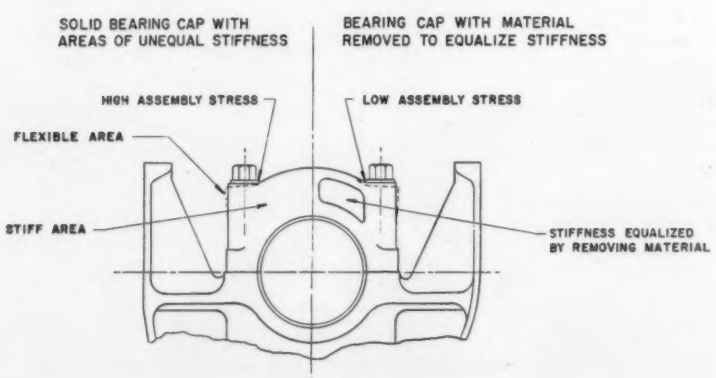



Fig. 18 — Bearing cap design effect on assembly stress.

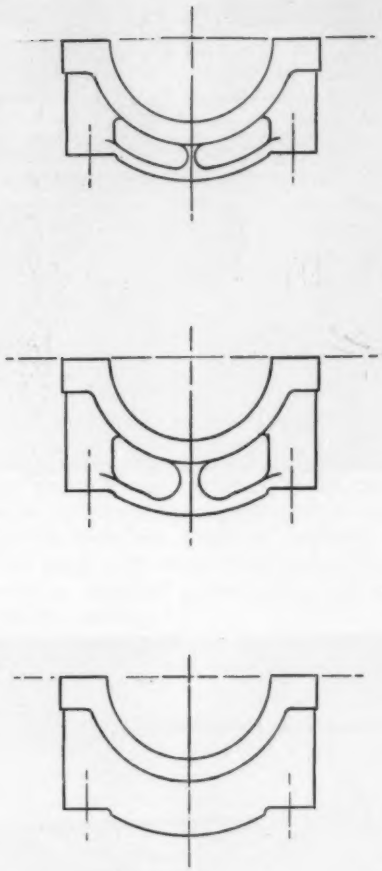


Fig. 19 — Comparable strength of main bearing caps. *Top* — old production (C.I.) — wt. — 10.37 lb. Safety factor — 1.02. *Center* — revised old production (C.I.) — wt. — 8.614 lb. Safety factor — 1.22. *Bottom* — present production (D.I.) — wt. — 7.915 lb. Safety factor — 1.40.

in the area aligned with the long studs on the camshaft side of the engine. This area was also where the predominance of operating failures occurred, indicating that the assembly stress contributed to the observed failures, Fig. 22.

Strain gages were installed on a cylinder liner which was installed in an engine, and readings taken under operation. The strain gages were installed to measure both the longitudinal strain and the hoop strain. Typical oscillograms taken from these tests showed that the assembly stress was the minimum stress, that the working stress increased with horsepower and that hoop strains were of about the same magnitude as the longitudinal strains. The strains recordings on the oscillograph closely resembled firing pressure recordings.

A static test set up in which the crankcase, sleeve, crankshaft, cylinder head, piston and connecting rod were assembled as in an operating engine. The piston was reworked to accommodate an O-ring, so that the firing pressure could be simulated by hydraulic pressure acting on the piston. It was found the results obtained on the static tests were similar to the results obtained on the dynamic test in the operating engine, and the balance of the test work was done under static conditions.

In general, the tests showed that the assembly loads and the firing pressure caused the sleeve to bend, as shown in Fig. 23. The assembly loads effected mainly the stresses in the longitudinal direction, and had little effect on the hoop strains. The firing pressure, however, affected the stresses in both the longitudinal and hoop direction causing a bi-axial stress condition. The hoop stress is generally the most critical in a cylinder under pressure. However, because of the effect of assembly loads on the longitudinal strains, the failure actually started in the longitudinal direction.

### *Proposed Solutions*

Since the mechanism of failure was known, many solutions to the problem presented themselves. Some of the proposed solutions included reducing the assembly torque, reducing the firing pressure, changing to a better material, shot peening the failure area,

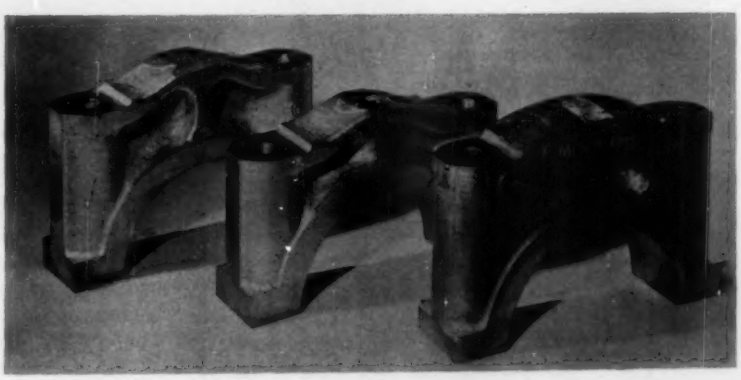


Fig. 20 — Evolution of bearing cap design.

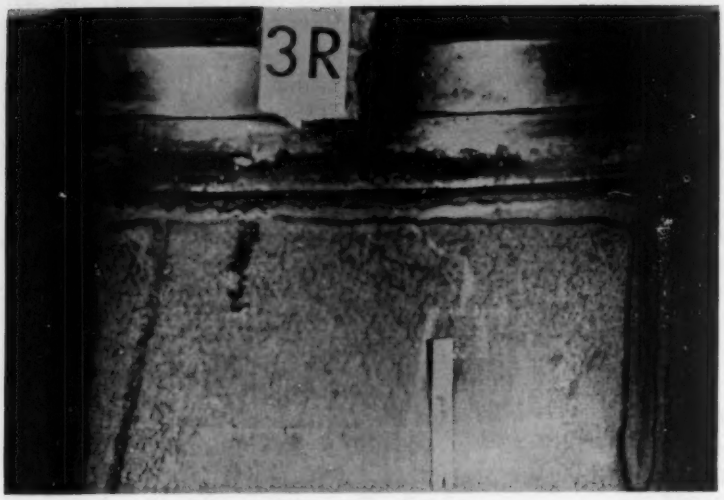



Fig. 21 — Cylinder liner failure from operation.

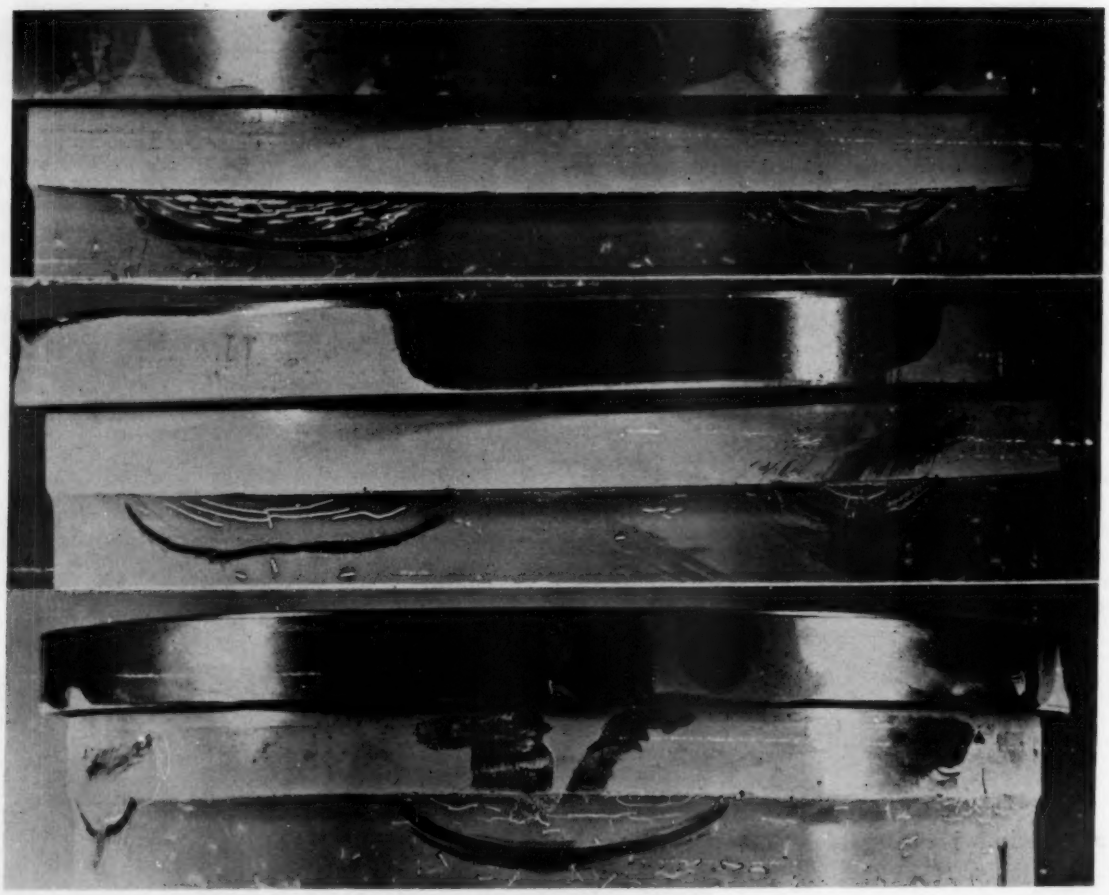
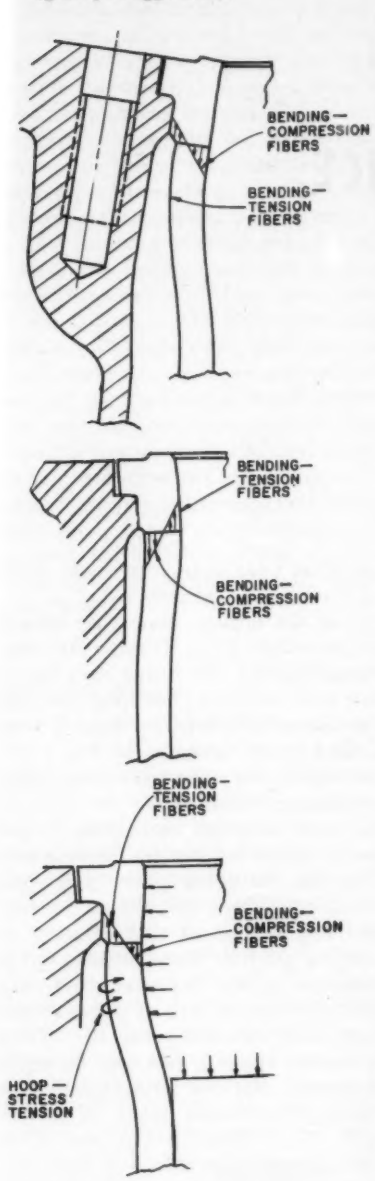


Fig. 22 — Stress coat patterns on cylinder liner. Cracks line up with large studs.

Fig. 23 — Mechanism of loading on cylinder liner (greatly exaggerated).




Sleeve deformation due to tightening torque (section at stud).
Sleeve deformation due to tightening torque (section between studs).
Sleeve deformation due to hydraulic loading (section at and between studs).

increasing the section modulus in the failure area or just increasing the support against bending. The simplest of all the proposals was to increase the support against bending, as this could be accomplished by simply reducing the clearance between the crankcase bore and the cylinder sleeve pilot.

The modified Goodman diagram in Fig. 24 shows that the safety factor increased from less than one to slightly under 1.2 by this change.

Experience to date has shown that the corrective measures made on the basis of these analyses presented has been successful. To date, with the engines that have operated in the field, some for many thousands of hours, no failures have been reported in the areas discussed. This, of course, has given a great deal of confidence in the experimental stress analysis techniques and methods, so that the thinking of management has changed.

Today, stress work is started when the first pieces are made, so that structural development parallels the performance development. In the past, stress work was conducted only after failures were experienced under operation. Furthermore, the lessons learned are being applied to new designs so that less mistakes are made on the first pieces and less corrections have to be made, since the principles of good design in gray iron castings are made readily apparent.

## SUMMARY

Some of the lessons learned from these experiences are:

1. Carry loads through members loaded principally in tension or compression, preferably in compression.
2. Avoid carrying any loads in bending.
3. Do not neglect the effect of assembly stresses.
4. Determine material properties by specimen made from the casting.
5. The modified Goodman diagram is an excellent means of analyzing a structural assembly under operating loads.
6. Work closely with the foundry so they will know where critical areas are, and what is required in the critical area.
7. Locate material controls such as hardness specifications in critical areas where the material properties are important.
8. Castings are not made stronger by making them heavier.

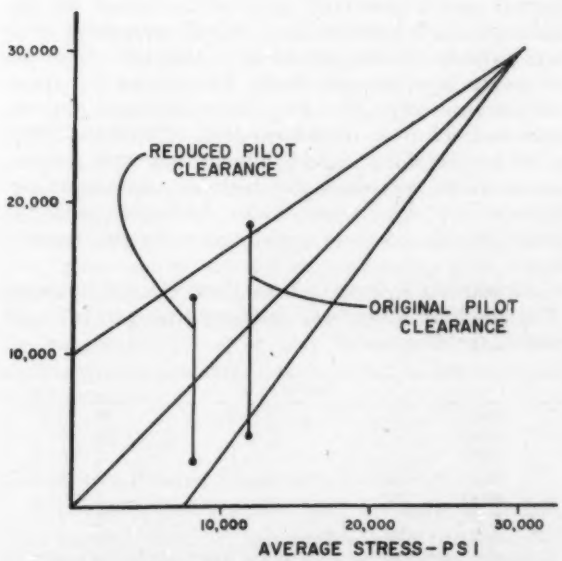



Fig. 24 — Modified Goodman diagram for maximum sleeve stresses combined by $\sigma_{MAX.} = \frac{E}{1-r^2}(\epsilon_x + \epsilon_y \mu)$ $E = 14 \times 10^6$. 2100 psi pressure.

# LADLE REFRACTORY PERFORMANCE IMPROVEMENT TEST PROGRAM

by P. J. Neff, J. J. Downs and J. T. Baker


## ABSTRACT

**Necessary increases in the temperature of metal disclosed inadequate performance of refractories commonly used in ladles. A program of testing and evaluation has revealed that such materials have little or no reserve capability to withstand increases in temperature. Magnesite furnishes this reserve for nozzles. All other refractories in the ladle are improved by the use of materials containing alumina in amounts considerably higher than are ordinarily used. The higher initial cost is justified by a reduction in cost per ton poured and by an improvement in the pouring operation.**


## INTRODUCTION

The ladle refractories and methods of application used at the authors' company for many years conformed to the standards generally accepted by the industry. Such practice had served reasonably well under the conditions prevailing at that time. Changes in production methods made an increase in metal temperatures desirable. They were increased approximately 150 F from the former level of 2800 to 2850 F (1538 to 1566 C). It is well known that such temperatures alone are sufficiently high to cause rapid deterioration of fireclay refractories. Additional problems arose because of more rapid attack by the furnace slag.

All melting is done in basic lined electric furnaces. The resulting slags have analyses that generally lie within the ranges of:

| | Analysis, % |
|---|---|
| CaO | 40 - 50 |
| $SiO_2$ | 10 - 20 |
| FeO | 2 - 10 |
| MgO | 4 - 15 |
| MnO | 5 - 10 |
| $Al_2O_3$ | 2 - 7 |
| $P_2O_5$ | 0.1 -0.5 |

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The high percentage of CaO is, of course, the principle reason for attack on clay refractories.

The ladles are all of the bottom pour type having nominal capacities of either 5 or 10 tons. As indicated by production schedules, the lining may be altered to change the ladle capacity. For example, the 5 ton ladles have at times been lined to hold 3 tons. All the ladles are brick-lined, as shown in Fig. 1. Included in this drawing are the other refractory parts of the ladle to be discussed separately.

In the main, the work that has been done to improve the performance of ladles can be divided into two categories. The first including nozzles, stoppers and sleeves directly affected the capability of pouring the heat and obviously influenced such matters as metal losses and casting quality. The performance of these parts is influenced by the fact that there may be as many as 230 openings in a heat. The second was concerned more with extending the life of the ladle lining for economic reasons, not only to equal the former performance but to substantially improve it.

## NOZZLES

Nozzles generally used in bottom pour ladles are of the fireclay type, although references can be found on zircon, magnesite and a few other variations. It seems that two virtues have been ascribed to fireclay nozzles. The first and most obvious is their low cost. The second, while more obscure, is apparently considered to be of great importance by many. This is the capability of the nozzle to soften at high temperature and thus permit deformation by the stopper. In this manner leaks arising out of poor fits or caused by damage to the stopper during pouring can usually be controlled. While the combination of the soft nozzle and the hard stopper is almost historic, it should be obvious that such deformation and the attendant erosion of the nozzle must be accompanied by transfer of the eroded material to the metal stream and eventually into the product.

Some years ago this problem of nozzle erosion, especially with steel containing 1.5 per cent manganese, was recognized, and the use of an improved nozzle was adopted. While no attempt was made at that time to analyze the material, it was known that the clay had a higher alumina content. Experience showed that its resistance to erosion was improved, although deformation of the seat was common. With increased pouring temperatures, there was an immediate and severe increase in the amount of erosion of these nozzles.

This caused a notable increase in nonmetallic particles in the castings, and also led to a vast increase in metal losses. Most of these metal loses were attributed to the failure of the nozzle resulting from excessive erosion, although there were too many instances in which there was also some sort of a failure in the stopper. All the evidence at hand supported the point of view that the composition of the nozzle was such that it simply did not have the capability to withstand the higher temperatures. Softening of the material at these high temperatures, combined with the erosive effect of the steel, and particularly the manganese in the steel, led to failure. Thus, a need was indicated for a type of material which could better withstand these conditions.

### *Magnesite Nozzles*

Of the several possibilities considered, the most obvious area for investigation was the use of magnesite. Magnesite nozzles are not new, and have been used for more or less specialized purposes for many years. A fair amount of information was available on this subject from one of the company's plants which had used magnesite nozzles for several years on certain heats. The only useful information available was that no erosion occurred during pouring, but that there was a prevalent tendency for the nozzles to leak. This was attributed to the inability of the nozzle to soften and form a perfect seat for the stopper.

Since it was most desirable to use a magnesite nozzle, consideration was given to steps which might be taken to eliminate the tendency for leakage. A number of design changes were proposed for the seat of the nozzle. For example, the conical seat, such as that used in an automobile valve to form a gas-tight seal, was considered. The common nozzle design and three other designs are shown in Fig. 2. The conventional or convex surface was rejected, since it already had a history of leakage. The conical surface was eventually rejected because it required a corresponding change in the design of the stopper.

The concave surface seemed to have the most promise. Mating of the stopper and nozzle surfaces with conventional practice is essentially on a line contact. With the inability of the nozzle to deform, there was speculation that the line bearing was not continuous and, therefore, permitted a small amount of leakage. The first design with a concave seat was such that the nozzle would conform over its entire top surface to the end of the stopper. Therefore, a large area contact would be presented. This was subsequently modified slightly, since it was foreseen that it would not be possible to consistently produce a concave surface of

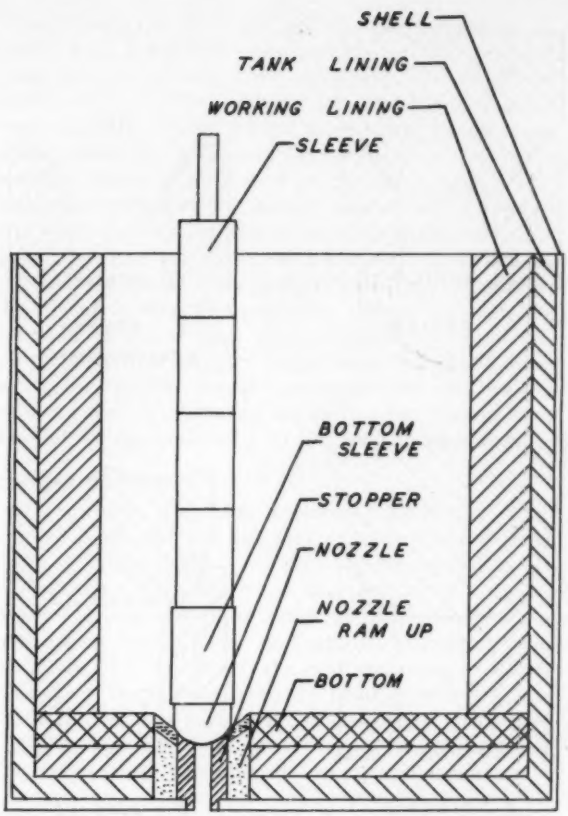



Fig. 1 — Construction of ladle.

sufficient accuracy. Therefore, the design was changed to increase the radius of the seat in the nozzle so that there would be some clearance.

A mold was built for making magnesite nozzles of this design, and samples were made and tested in production. In the initial trials, sand used during testing prior to tapping of the heat was allowed to remain in the small clearance groove around the side of the stopper. It was initially feared that the first metal into the ladle would fill this groove, and because of the high conductivity of the nozzle would freeze in place thus interfering with subsequent performance. It was later found that the use of sand in this area was unnecessary.

Nozzles with 1½-in. holes are used exclusively. However, a few one in. nozzles were tried but were abandoned, because of the tendency to freeze during pouring. This was attributed to the high conductivity of the magnesite. Thus, it may be possible that less satisfactory performance could occur at lower metal temperatures.

### *Chemically Bonded Magnesite Nozzles*

The nozzles that are now used are made of chemically bonded unburned magnesite. While this material has shown inadequate performance in other parts of the ladle, performance in nozzles has been excellent. Erosion, for practical purposes, is nonexistent, and there is no deformation of the seat during pouring

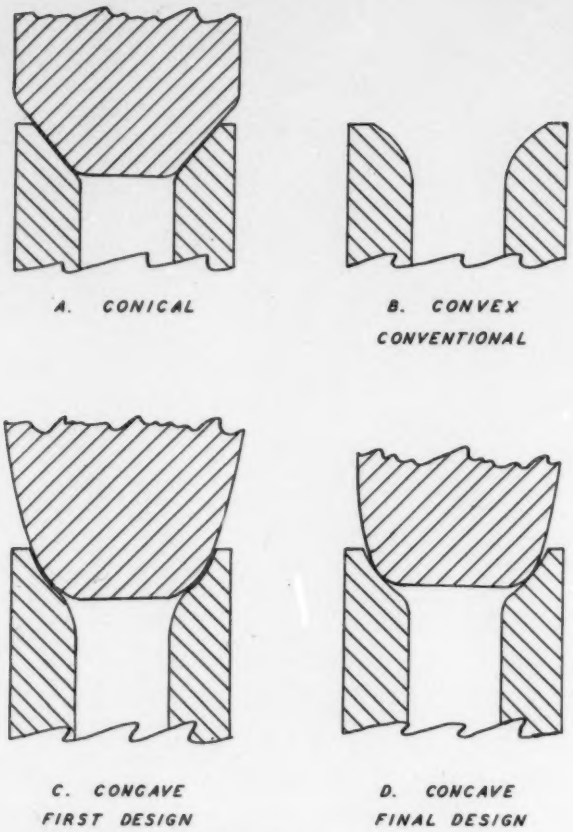



Fig. 2 — Design of nozzles.

heats under any conditions. There is no problem with cracking or spalling.

Contrary to the experience reported by others in pouring carbon and low alloy steels, there is no evidence of any buildup in the flow hole of the nozzle. Such buildup has been attributed by others to the use of aluminum in the steel, leading to a successive reduction in the diameter of the flow hole to the point where pouring of the steel is seriously impeded. There is information also presented in the literature advocating strict adherence to certain design principles in nozzles among, which is the requirement for a convex-shaped seat to produce streamline flow of metal through the nozzle. Seemingly in contradiction of these principles, the concave nozzle seat has consistently produced a smooth flowing stream.

There is little tendency for the metal stream to waver or to form an umbrella. Having arrived at the design as previously described and resorting to the use of magnesite as a material, no further changes have been undertaken in the nozzle. This design of magnesite nozzle has been used on all heats ever since it was first produced. During its use, covering a period of several years, there are no known instances of failure during pouring which could be attributed to the nozzle. Unfortunately, the successful use of magnesite nozzles did not eliminate metal losses. On the other hand, it disclosed more clearly the fact that stoppers were equally at fault and warranted intensive investigation.

## NOZZLE WELL RAM-UP MATERIAL

After the use of magnesite nozzles became standard practice, it was evident that a large amount of erosion of the nozzle well ram-up material occurred. The material normally used at that time was a clay graphite mixture. After dumping the slag from the ladle, the nozzle would protrude above the nozzle well ram-up material. This was not noticed when using the clay nozzles because they eroded at the same rate as the ram-up material. Figure 3 shows the nozzle well rammed with a clay graphite mix, and a magnesite nozzle showing this erosion. Because of the proximity of the nozzle well ram-up material to the nozzle, it is most likely that material eroded from this area would be carried into the molds.

Magnesite ramming mixes from two suppliers were tested. Both were superior in erosion resistance to the clay graphite mix. On the other hand, freshly prepared mixes dried rapidly and had to be used in

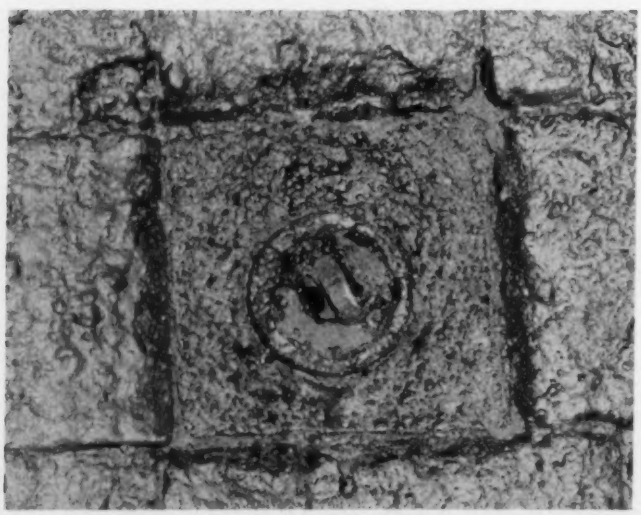


Fig. 3 — Erosion of clay graphite nozzle well ram-up from around a magnesite nozzle.

a short time. They were also more difficult to ram, and it was necessary to fire the mix immediately after ramming to prevent crumbling of the surface. For these reasons, other materials were tested.

A mullite ramming mix used in a nozzle well was found to have excellent erosion resistance. It was similar to the clay graphite mix in bench life and ease of ramming. Furthermore, it was not necessary to fire the nozzle well immediately after ramming. A sillimanite ramming mix was found to be superior to the clay graphite mix, but eroded more than the mullite ramming mix. The bench life and ramming characteristics were the same as the mullite mix. The mullite mix is used at the present time as the nozzle well ram-up material, and is applied as a 2 in. thick layer over a cheaper and less refractory base.

## STOPPERS

While magnesite nozzles were not a cause of leakers, they did accentuate a condition that probably existed all along—a condition which was disguised by clay nozzles that softened in use. The clay graphite stopper heads were cracking, spalling and eroding and evidently were completely unable to withstand the elevated pouring temperatures. Magnesite nozzles do not soften in use as did the clay nozzles previously used; therefore, when a stopper spalls or erodes and loses its contour, the nozzle will not conform to the new shape and a leak results. As spalling and erosion continue, it becomes increasingly more difficult to attain a shut-off, and eventually, in some cases, the rest of the heat has to be pigged. Not only do these stopper failures cause metal losses and a safety hazard, but they act as a source of dirt in castings. A stopper was needed which would be more resistant to thermal shock, spalling and erosion. There was also a need for a method of testing stoppers without incurring the high metal losses, which could result if an experimental stopper were to fail in a production heat.

There is a standard thermal shock test used for refractories. This test consists of heating the sample to a high temperature in a gas-fired furnace followed by air cooling under closely controlled conditions. This procedure is repeated until the refractory in question spalls or cracks. The number of cycles to failure is recorded, and this number is the test result. While this test may be satisfactory for some refractory applications, it in no way exposes a piece of refractory to the conditions approaching those to which a stopper head would be subjected in use.

Under these conditions the head is rapidly heated to temperatures much higher than those used in the typical thermal shock test. At these temperatures it is repeatedly pressed into a nozzle while hot erosive metal flows around it. What was actually needed was a service condition test, one which could simulate the conditions to which a stopper head is subjected and one that could easily be run using available facilities. The facilities available were a number of standard heat treat furnaces and three induction melting furnaces. The induction melting furnaces offered the best possibility for a service condition test. In such a furnace, ladle conditions under which the stopper head must operate can most easily be duplicated.

To conduct such a test, the sample stopper is assembled on a rod with a normal stopper rod pin and plug mix. It is then dried and preheated to about 400 F. The head is suddenly immersed into molten steel at 3000 F and held for 5 min at that temperature with power on. This subjects the head to the same thermal shock as it would receive in a ladle. The induction action of the furnace washes metal around the head, duplicating the erosive conditions of pouring. The time period for which the head is held in the metal allows it to be uniformly heated throughout. The head is then removed and allowed to cool. Inspection of the stopper starts immediately upon removal from the furnace. Visual examination if done while hot, readily reveals spalling. After the head is cool, a closer examination not only shows the spalling but allows determination of cracking or erosion.

### *Stopper Failures*

The head is then cross-sectioned transversely. This section is ground smooth and examined for various types of failures which are not necessarily visible on the surface. The effects of this test upon stopper heads are essentially the same as those observed in production heats. First, spalling is a common type of failure (Fig. 4). This is the type of flaking in which thin layers break away from the head, exposing a new surface. This new surface then is subjected to further spalling, and the head progressively disintegrates in this fashion until leaking or loss of remaining metal results. Any spalling at all will disqualify a head for a production trial.

The second type of failure is radial cracking (Fig. 5). Radial cracks usually occur at 120 degree intervals and split the head into three sections. This type of failure is considered to be more serious than spalling. Once the crack initiates it continues to widen, and leaking almost surely results. In severe cases, the heat is lost. The third type of failure commonly observed is erosion where material is washed away from the surface of the head by action of flowing metal (Fig. 6). Templates can be used to measure the degree of erosion.

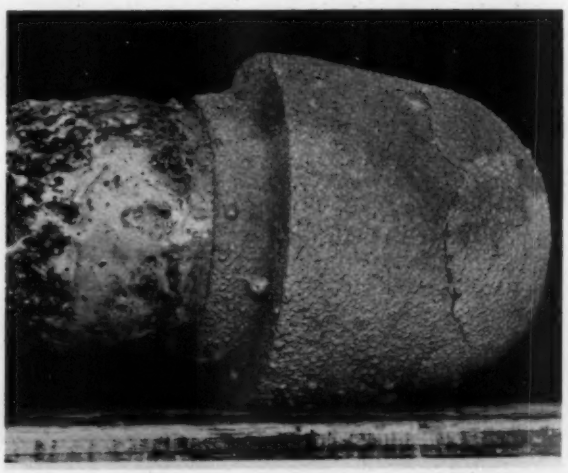


Fig. 4 — Spalling of a stopper head.



Fig. 5 — Radial cracking in a stopper head.

All sample heads are tested in duplicate and both heads must pass the test in order to warrant a production trial. The test requires that the head does not spall or crack, and that erosion is at a minimum. Experience has shown that even a slight tendency to spall or crack in the thermal shock test will result in poor performance in a production heat.

*Stopper Head Test Program.*

The failure of the clay graphite stopper heads which were being used marked the start of an intensive test program. At first many commercial heads from domestic suppliers were tested. These common heads contain about 12 per cent graphite. The ignited sample contains 60 per cent silica and approximately 35-37 per cent alumina. It was found that heads of this general composition spalled severely when shock tested, eroded rather readily and subsequently performed quite poorly under operating conditions.

When it became evident that a satisfactory head was not available, a number of special heads were produced on an experimental basis. Among the special mixes which were formulated and tested were—chemically bonded magnesite, magnesite chrome, burned chrome magnesite, high refractory clay without graphite, fused silica, periclase and mullite. All of these had a tendency to crack and break apart when tested and none was satisfactory.

A combination mullite graphite head was tested which was better than the heads formerly used, but not good enough to warrant use. A stopper head was obtained from a foreign supplier which contained 26 per cent graphite, 60 per cent alumina and 35 per cent silica, as determined on the ignited sample. Not only is the graphite much higher, but the percentages of alumina and silica are essentially the reverse of domestic stopper heads. The apparent density was 1.88. This particular head was practically inert when subjected to the thermal shock test. It did not spall or crack and erosion was nonexistent. Production trials of this type head gave excellent performance. It was subsequently adopted as part of the standard practice and became a standard of comparison for test purposes.

A large amount of test information leads to some conclusions on factors influencing the behavior of heads when subjected to a thermal shock test. The first of these factors is composition of the head. Tests to date show a correlation between spalling tendency and silica content. High silica is conducive to spalling, and it follows that the higher the silica content the more severe is the degree of spalling. As silica decreases from approximately 60 down to 40 per cent, the severity of spalling decreases proportionately un-

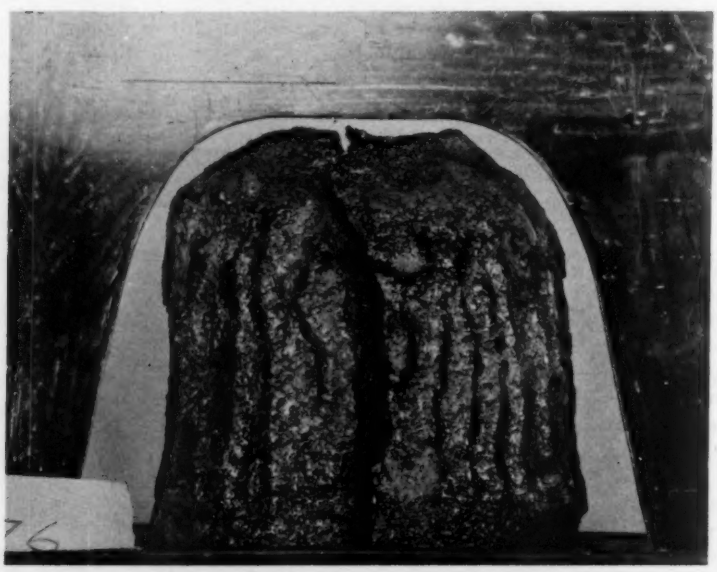


Fig. 6 — Eroded stopper head.



Fig. 7 — Badly eroded stopper rod sleeves.

til it disappears at approximately 34-40 per cent $SiO_2$. The alumina, which should make up the balance of the mix, adds to refractoriness, and seems to increase resistance to erosion.

The function of the graphite is not clear, but it apparently influences heat conductivity, thus affecting thermal expansion, a major cause of spalling and cracking. It may also influence density, the other major property affecting performance of the heads. Density may be affected not only by graphite content but also by the method of forming and firing of the heads. High density leads to radial cracking, even if alumina and silica contents are in the desired range. An experimental head was tested which had the proper amount of alumina and silica but was low in graphite and had an extremely high density of 2.18. This head showed little erosion and no spalling, but it did crack radially.

All tests to date indicate that the optimum composition for the stopper heads is one containing 25 per cent graphite and 75 per cent clay which, in turn, is composed of approximately 60 per cent alumina and 35 per cent $SiO_2$. Optimum density is around 1.88. It has also been observed that a fine grain size of the refractory particles is desirable if optimum properties are to be developed.

## STOPPER ROD SLEEVES

With increased metal temperatures the erosive effect of basic slag on common clay stopper rod sleeves created an intolerable situation. This was especially true with steels containing manganese in the range of 1.5 per cent. Figure 7 shows such a rod assembly after use. In some cases sleeves eroded early enough to expose the rod to molten metal or to molten slag, causing failure of the rod and partial loss of the heat.

In a first attempt to obtain better performance, the common clay sleeves were replaced with sleeves which the supplier indicated had a higher PCE. It was presumed that this was achieved by an increase in the alumina content. The performance of these sleeves was essentially no better, and it was concluded that any changes the manufacturer may have made were not great enough to significantly alter the properties of the parts.

To withstand the attack of the basic slag it seemed reasonable to consider the use of a basic refractory. Accordingly, a number of chemically bonded magnesite sleeves were produced. Several of these sleeves were first subjected to immersion in molten steel in an induction furnace. Several heats were also poured from a 1000-lb bottom pour ladle. In these preliminary tests it appeared that the sleeves had good erosion resistance and did not spall. On this basis, several 9-ton heats were made.

The sleeves performed satisfactorily from a mechanical standpoint but were found to evolve large volumes of gas, presumably from decomposition of the material used for bonding. As a result, many of the castings poured in the heat contained an objectionable amount of porosity. It was found that this gas forming tendency could be eliminated by preheating the sleeves to 1400 F (760 C) before use. Inasmuch as this involved an extra operation, it was not considered practical.

Chrome magnesite sleeves were found to have good resistance to erosion and spalling and were used in production for a short time. However, they developed a tendency to crack, allowing metal to penetrate to the rod. This tendency to crack seemed to become worse as the age of the sleeves increased. In general, the overall results obtained when using sleeves made of basic materials was unsatisfactory.

### *Alumina Content Variation*

Attention was then directed to the consideration of materials containing more alumina than is commonly used. Sleeves made of mullite, for instance, were found to have excellent resistance to erosion, and were practically immune to any signs of spalling or cracking. An objectionable feature of the mullite sleeves was their much higher cost. A partial cost reduction was achieved by changing the design so that the diameter was reduced from 5½-in. to 4⅝-in. The capability for making such a reduction in diameter was directly attributable to the greatly improved resistance to erosion.

The larger diameter of the common clay sleeves was partly a concession to erosion, to insure that there would still be an adequate amount of refractory remaining at the end of the heat. In spite of the reduction in cost achieved by reducing the weight of the parts, the cost of the mullite sleeves was still prohibitively high. In view of their exceptional performance, it was believed that some reduction in the alumina content could be made without seriously affecting performance.

Sleeves were prepared containing alumina in amounts of 50 and 60 per cent, as compared with the 70 per cent present in mullite. These sleeves were less subject to erosion than the common clay sleeves, but were not as good as the mullite. It was found in a general way that the extent of erosion was essentially proportional to the amount of alumina in the refractory. Furthermore, the cost of the sleeves was also directly related to the alumina content. On this basis, it was found that sleeves containing 50 per cent alumina were adequate for production purposes.

While a substantial amount of erosion could occur during pouring, the wall thickness was sufficient to protect the rod throughout the heat. An exception was made in the case of the bottom sleeve. This sleeve covers the stopper, and it is imperative that no metal enter the joint between the stopper and the sleeve. If attack in this joint is too great there is the likelihood that the stopper rod pin may be melted. Sixty per cent alumina sleeves were found to give this added protection.

## LADLE BOTTOMS

Prior to the requirement for improved ladle performance ladle bottoms were laid with brick containing 70 per cent alumina. The higher cost of this brick compared to more commonly used fireclay had been justified on the basis of a longer life. With the increased temperatures ladle bottoms deteriorated more rapidly. This was of concern not only in the consideration for the total life of the ladle, but also because of the increased likelihood that some eroded parts of the bottom could be carried into the castings.

A single attempt was made to construct a ladle bottom of magnesite brick; however, this brick cracked and spalled so severely after three heats that it had to be replaced.

Attention was, therefore, directed to the use of a monolithic lining for the bottom of the ladle. Rammed magnesite, which was considered a likely material to resist attack of the slag, gave a poor performance. Deep cracks were present after the first heat. Even though they were patched, the cracks were so enlarged after a second heat that the bottom had to be replaced. This failure was attributed partially to the alternate heating and cooling of the ladle bottom heats. Conceivably improved performance might be obtained if the ladle could be continuously maintained at a high temperature.

An attempt was also made to construct a similar type of bottom using a calcium aluminate cement. This material had given excellent results when used to line 1000 lb ladles; however, in a manner similar to the magnesite, the lining cracked severely after five heats and allowed metal to run under it. At first, cracking was attributed to damage incurred during removal of the nozzles, therefore, the test was repeated with the nozzle well surrounded with brick. This lining cracked in a manner identical to the first lining.

Although there is some consideration still being given to the use of a monolithic lining, this type of bottom has not been explored any further. On the other hand, brick with alumina content greater than 70 per cent is being actively investigated.

## LADLE LININGS

The increased metal temperatures and the greater activity of the basic slag effectively caused a marked reduction in the life of the clay ladle linings. Prior to attempts to improve the life of the linings, the sidewalls were lined with hard burned clay brick containing 36 per cent $Al_2O_3$ and 58 per cent $SiO_2$. For some time these brick had been obtained in a cupola block shape as a means of reducing the number of brick joints from what would be required when using arch brick. With the increased temperatures the life of the lining was reduced from 13 heats to more than seven heats. While several different sized ladles were used at one time or another, most work done in evaluating performance of sidewalls was done on 5 ton ladles.

The method of testing and evaluating bricks varied as experience was gained. Initially, a number of types of brick were tested by installing a complete lining. It was quickly found that a given type of brick would not necessarily perform as anticipated, and in view of the markedly higher cost of some of the bricks that were used the cost of testing an entire lining could be prohibitively high. As a result, the use of patch tests was undertaken. The literature on refractories deals to some extent with the use of patch tests and several techniques have been described.

One method suggests that such a patch should extend all the way from the top to the bottom of the ladle, so that erosion products from the upper portion will not wash down over a test section and, therefore, produce erroneous interpretations. On the other hand, in some cases the amount of brick available for such a test would not have permitted such an extensive patch. Therefore, the general technique used involved the installation of a patch approximately 15 to 20 in. square about half way up the sidewall of the ladle. In general, this could be accomplished with two courses of arch brick, each containing about five brick.

These test pieces were installed in the sidewalls of ladles made up with brick that was considered standard at the time. In some instances, at the conclusion of a test, when it became necessary to discontinue the use of a ladle because of sidewall wear the test panel would protrude from the working face of the ladle a distance of several inches. It was possible to roughly measure the amount of erosion, and thereby extrapolate to estimate what the life would have been with an entire lining of the test material. Such a technique permitted the screening of a larger number of samples without the need for lining entire ladles.

It also permitted the testing of two or three different materials in a single ladle, thereby greatly speeding up the rate at which tests could be conducted. Obviously, the more promising of these test materials could then be used for producing complete linings.

### *Basic Brick Tests*

Since attack of the basic slag on the clay brick was the principle reason for early deterioration of ladle linings, a number of basic brick shapes were tested. Forsterite (magnesium silicate) and chrome magnesite were tested in full linings. Severe spalling with both types of brick made it necessary to remove the lining from service after about 13 heats. Thus, although the ladle life was essentially doubled, the higher cost of the brick did not justify their use. Such spalling, in the main, was found to be almost a common characteristic of the basic brick that were tested. The lining, in general, performed quite satisfactorily for the first few heats, although there was a tendency for slag to adhere.

This adherence of the basic slag has been attributed to an increase in the melting point of this slag caused by a pick up of MgO from the ladle lining. If an attempt was made to remove the slag in cleaning the ladle, separation did not occur at the slag-brick interface but at some distance in the brick itself. Even though no attempt would be made to mechanically remove the slag, it would subsequently spall spontaneously and remove a substantial part of the brick at the same time. In the main, similar results were obtained on patch tests with burned magnesite, burned and unburned magnesite chrome, chrome and pure magnesite brick.

In view of the generally unacceptable results with basic materials, attention was directed to the use of materials containing higher percentages of alumina. Brick of this type is not subject to spalling or the buildup of slag that occurred with the basic materials, although their use did expose a problem in method of installation and the effect of brick joints. When it was found necessary to discontinue a test on a lining of 80 per cent alumina brick because of rapid failure at the brick joints, the method of installation was reviewed.

It was found that the practice in general use included a mortar joint generally ¼-in. thick, and in some instances as much as ½-in. thick. Rapid erosion at the corners of the brick and penetration into the mortar caused premature failure. The practice was changed to drastically reduce the amount of mortar, and to change its consistency to that of a slurry to permit laying the brick with as small a joint as possible. The effect on the performance of subsequent ladles was drastically improved.

### *70 Per Cent Alumina Brick*

The most promising results and the quickest increase in ladle life were achieved with the use of 70 per cent alumina brick using the cupola block shape. The life of such a lining was 26 heats, or an increase of more than four times that obtained with the regular clay cupola block. Although the initial cost of this lining was higher than that of the clay brick, the extended life reduced the lining cost per ton of metal to a justifiable level. As a result, the 70 per cent alumina brick was adopted as a standard to replace the hard burned clay. Such cost studies as were made were based on the direct labor for relining a ladle and the cost of materials used. A saving more difficult to evaluate was the greatly reduced rate of relining achieved by the increased life.

Mullite brick was also tested and was found to perform well; however, at the time it was not felt that the higher cost could be justified. Their excellent resistance to slag might suggest their use in the slag line area of the ladle, which is subject to the most vigorous attack.

A number of other bricks containing high alumina contents (up to 85 per cent) have been tested and have all performed well. Obviously, the increased life of the ladle must be of sufficient magnitude to justify the expense. For instance, the cost of lining a ladle with a brick containing 85 per cent alumina would require at least 45 heats in order to justify use of the brick. A recent ladle lined with this brick was used to pour 60 heats before it was retired from service. At this time additional ladles have been lined with 85 per cent alumina brick, since it is obvious that they would be more economical in the long run.

In order to keep pace with improved performance of bricks, it was necessary to make comparative studies of several different available brick mortars. In general, these were tested by using different mortars in different sections of a given ladle. The brick joints were then examined after each heat to obtain a comparative measurement of any undercutting that might occur. Of the several mortars tested, it has been found that there is little difference in their performance; however, all of the mortars that have been used in the ladles containing high alumina brick are those mortars with high alumina contents.

In general, the experience gained with the brick linings and with mortars used has paralleled experience in other parts of the ladle, i.e., that the high alumina refractories are significantly more resistant to attack by slag, and can presumably be used to extend the life of the ladle and reduce the likelihood of failure.

Experience gained in this work has shown that materials seemingly similar do not necessarily perform alike. Nor is there assurance that successive shipments of a single item will be alike. Thus, there is indication of the desirability of a test program so that individual shipments can be carefully checked to ascertain their quality level.

## ACKNOWLEDGMENT

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# MAINTENANCE MATERIALS CONTROL SYSTEMS

by K. M. Smith


## ABSTRACT

**The maintenance management viewpoint of how maintenance material control systems and material disbursing systems can best serve maintenance department needs and at the same time help a company achieve low production costs is presented. The steps leading to an adequate evaluation of need for each item is outlined, and details of crib arrangement and operation are discussed.**


## BACKGROUND KNOWLEDGE

Inventory items can be more properly controlled if these points of background knowledge have been adequately developed:

1. Materials and parts that may be required.
2. Approximate quantity of each item in use.
3. Past usage or expected future usage of each item.
4. Purchase cost of each item, and yearly cost per inventory dollar for each type of item.
5. Normal and emergency sources of each item.
6. Normal and emergency procurement times and additional emergency procurement costs.
7. Equipment downtime costs per hour, shift or day that the equipment is not available for production use.

### *Steps to Knowledge*

**1. Identify all equipment.** Each item of production equipment, as well as plant equipment, such as heating units, fans, valves, power sweepers, cranes, hoists, pumps and transformers, should be given an identifying number and name. This identifying number and name can be an embossed or stamped tag or plate, a painted legend, a decal or an adhesive label. Pipe lines should be identified by contents or service.

**2. Set up equipment record folders.** A separate equipment record folder should be setup for each equipment item or equipment category. Items such as electric drills, valves and pipeline classes are best handled in category folders, except for specialized items best kept in separate folders.

**3. Collect a full set of information.** On each equipment item there is need for full information. This information should be placed in the equipment record folder.

1. Make an index sheet for the contents of each folder to expedite the finding of the information contained in the folder, and also to serve as a reference means of replacing lost data, drawings, instruction sheets, etc.
2. List all drawings relating to a machine or equipment item and indicate the ones to be kept in the folder.
3. Make up basic and special data sheets on each equipment item.
4. Obtain a full set of operator's manuals for all cases where they exist. Study of these manuals provides a better understanding of the probable maintenance needs of a machine.
5. Obtain a full set of spare parts books or spare parts drawings for each equipment item. Do not overlook spare parts sheets for the electrical control items such as motor starters, disconnect switches, relays, limit switches, timers, etc.
6. Review spare parts books and drawings for completeness of the description of each part. Commercially available items such as bearings, seals, O-rings, couplings, roller chain, sprockets, electrical relays, brakes, should be completely identified so they can be purchased from the most economically convenient source.
7. Obtain a full set of electrical and hydraulic systems diagrams.
8. Obtain all other machine drawings that may be of value in repairing the machine.
9. Wherever practical obtain photographs or overall illustrations of the machine or equipment item.

**4. Make out a job ticket for each maintenance job performed.** When the job is completed, information on what work was done and how many of what parts were used should be added to the job ticket before the ticket is filed in the equipment record folder, unless other adequate information systems are already in use. Parts usage information will be helpful in determining minimum stock quantities. The job ticket should also show the date and the actual time required to do the job.

**5. Determine probable need for spare parts and materials.** Study the past parts requirements of each machine or of similar equipment, and develop a list of the parts and materials likely to be needed. The list may be made up by marking the spare parts lists and drawings or by making up a separate list. In some cases useful information will be available in the recommended spares list supplied by some manufacturers.

Determine the maximum number of any part that is expected to be essential for a repair job that cannot be reliably planned far enough in advance to allow for the purchase of the needed amount of materials.

Determine the approximate total installed quantity of each part.

Determine the past usage of each part or estimate the future usage of parts by comparisons with the history of known parts and machines.

**6. Establish the yearly cost for each inventory dol-**

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lar. A cost should be established for each class of material kept in inventory. This cost should be shown as per cent per year per dollar. This yearly cost should allow for interest charges, insurance, taxes, obsolescence, losses from theft and from handling of material, type and amount of material protection required and cost of storage space for each class of material.

For example, many unmachined castings, stainless steel castings and many barrelled products can be stored outside for long periods without serious damage. Other products can be kept in unheated dry storage, and machined parts are most safely kept in heated storage areas. Small parts will be kept in drawers, in bins or on shelves. Some materials can be stored as stacked pallets, and other materials must be kept in single pallets or as single layer storage in bins.

Avoid the temptation to lump together all inventory costs to obtain an average cost per dollar, because doing so will improperly burden small parts, many complicated machined parts and fragile items such as electronic tubes.

**7. Determine materials sources and procurement details.** Determine (a) the regular and emergency sources of each spare item, (b) the normal procurement time including order placing time, manufacturing time, shipping time and receiving and delivery time within the purchaser's plant, (c) emergency procurement time and (d) additional emergency procurement costs such as air express and special handling charges, and overtime charges by manufacturer.

In general the parts fabricated by the equipment manufacturer will only be available from that manufacturer. However, most manufacturers use commercially available bearings, oil seals, O-rings, bolts, belts, lubricating devices, pipe fittings, electrical components, fans, motors, pumps and hydraulic equipment.

It is advisable to determine the source or sources and full description of the commercial items, so that the replacement part to be used can be properly described in the individual spare parts book. This advance study of commercial items permits standardization of spare parts and establishes alternate sources of supply in case of emergency procurement.

**8. Determine the total cost of each item.** The list of potential spare parts for each machine should show the total cost of each item and the date for which each cost is correct. If special packaging charges, or substantial freight charges are involved but are not included in the price quoted by the vendor, the recorded cost should also show the approximate special charges.

A copy of the parts cost list should be kept in the machine record folder in each maintenance area to serve as a guide in determining the economy of repairing a part removed from service vs. the cost of buying a new part.

The inventory records for all parts which may be repairable should be marked in such a manner that a check is made with the proper maintenance supervisor before a new part is ordered.

**9. Establish production downtime costs for each piece of equipment.** A set of production downtime costs should be established for each machine or piece of equipment which could cause a loss of production in case of a failure. These costs should be shown as dollars per hour, dollars per production shift or dollars per day, depending upon the type of operation involved. If little loss occurs until the passage of a shift, a day or longer period of time, this fact should be shown. Where a machine is a part of a progressive line such that the shutdown of one machine causes losses by several machines, the dollar losses will be higher than for a separate machine installation. Thus progressive line installation should be noted on the downtime cost records.

It is suggested that the downtime costs be evaluated for conditions of low, medium and high rate of production operations. These downtime costs should be shown (a) in the equipment record folders and (b) in an index of equipment so arranged that comparative downtime cost comparisons can be conveniently made. The date applicable to each downtime cost should also be shown in the records.

## CHOOSING INVENTORY STOCK ITEMS

Two categories of items are involved in determining what should be kept in inventory. One category consists of those items not now in stock, but which are potentially desirable stock items. The second category consists of those parts currently in stock, and which should be evaluated in terms of need for greater stock quantity, smaller stock quantity or zero stock. The evaluation methods are similar for both categories up to the point that surplus stock of an item is evident. Disposal of surplus inventory will be discussed separately.

For infrequently used parts the cumulative cost of holding a part in inventory for multiple years should be known. The table shows the actual charge accumulated against each dollar of initial inventory cost at various annual charge rates compounded semi-annually for holding periods of up to ten years. When comparing downtime costs with inventory costs, the cost of a rarely used part held in inventory is the accumulated cost, not the original cost. The table makes no allowance for dollar inflation or for the possible price increase on parts for obsolete equipment.

**CUMULATIVE FINAL COST PER DOLLAR OF ORIGINAL COST FOR ITEMS KEPT IN INVENTORY FOR MULTIPLE YEARS**

| Kept in Inventory Until End of Year Shown | Annual Charge Rate — Compounded Semi-Annually, $ | | | | | |
|---|---|---|---|---|---|---|
| | 8% | 10% | 15% | 20% | 25% | 30% |
| 1 | 1.08 | 1.10 | 1.16 | 1.21 | 1.27 | 1.32 |
| 2 | 1.17 | 1.22 | 1.35 | 1.47 | 1.60 | 1.75 |
| 3 | 1.27 | 1.35 | 1.54 | 1.77 | 2.03 | 2.31 |
| 4 | 1.36 | 1.48 | 1.78 | 2.14 | 2.56 | 3.06 |
| 5 | 1.48 | 1.63 | 2.06 | 2.59 | 3.25 | 4.05 |
| 6 | 1.60 | 1.80 | 2.38 | 3.14 | 4.11 | 5.35 |
| 7 | 1.73 | 1.98 | 2.75 | 3.80 | 5.20 | 7.08 |
| 8 | 1.87 | 2.17 | 3.18 | 4.60 | 6.59 | 9.35 |
| 9 | 2.05 | 2.41 | 3.68 | 5.56 | 8.34 | 12.40 |
| 10 | 2.19 | 2.66 | 4.25 | 6.73 | 10.55 | 16.40 |

**10. Determine the practicality of and the cost of immediate repair of each part.** If a part is expected to fail by breakage, can the part be welded quickly

and successfully? If the part is expected to be replaced because of wear, can the part be repaired quickly? In many cases worn parts can be detected soon enough to plan for suitable repair during a nonproduction shift.

Once the possibility of repair, the downtime required and the cost of repair are known, the total of the repair costs plus the downtime cost can be compared with the inventory cost of a new part. If repairing the part provides the lowest cost, no inventory should be kept. If a part needs to be kept in inventory, it can be a repaired part if the repair costs are less than the cost of a new part.

Warning – A crib full of repaired parts not kept in dollar inventory may increase the total storage costs normally charged entirely to the new parts kept in dollar inventory. Perhaps required parts should be kept on or at the machine using the parts, although this decreases the reliability of having adequate spares.

**11. Determine the possibility of substituting another part on a permanent or temporary basis.** Determining in advance that a single part will serve satisfactorily in more than one installation permits a smaller number of dollars in inventory to give greater protection against downtime losses. An adequate description of the substitute part to be used, and any unusual conversion procedures to be followed, should be clearly indicated in the parts list for the machine.

If a part is to be used on a temporary basis and then be returned to inventory, it will be necessary to flag the inventory record to show where the part is being used in order to avoid the purchase of another new part for the inventory. It is suggested that the records for the machine, which normally requires the part in temporary use elsewhere, should be marked to indicate that such usage may take place and where it may take place.

**12. Compare downtime cost of immediate local purchase vs. inventory costs if part is kept in stock.** Many items such as bearings, belts, couplings, sheaves and electric controls can be obtained from local sources within a few hours time, if proper advance arrangements have been made for quick procurement of emergency items. An alternate plan is to keep an emergency minimum stock which would be replenished by quick procurement methods. Quick procurement is desirable where the minimum stock protects a number of machines.

Since the actual or estimated parts usage should have been developed earlier, it can be readily determined if it is economically practical to keep a part in stock. The table will be helpful in determining the cost of keeping parts which are used infrequently.

**13. Compare downtime cost plus extra costs of emergency purchasing vs. inventory costs if part is kept in stock.** Before this method can be used, it must be determined that the part manufacturer keeps the specific part in stock at all times. If the part is continuously available, and a reliable procurement time has been established, the practicability of keeping zero stock can be determined. The table will be helpful in determining the cost of keeping parts which are used infrequently.

Here again it may prove to be more desirable to keep an emergency minimum stock, which is replaced by the quick recording of items used on many machines.

**14. Establish a minimum stock system for all parts which are to be kept in inventory.** The quantity of a part to be kept in stock will be governed by (a) the original decisions that established the need for the part, (b) whether expedited or routine procurement is required, (c) the usage history of the part, (d) the delivery time for the part, (e) the most economical lot size in keeping with procurement cost per individual order and (f) discounts that may apply to quantity orders even though delivery may be scheduled for portions of the total order.

The following minimum stock designation system is suggested as a means of telling the Material Control and Purchasing Departments what procurement conditions apply to each part:

1E. One part is to be purchased by emergency procurement procedures when the one part in stock is withdrawn from stock.

2E. Enough parts to bring stock up to 2 parts are to be purchased by emergency procurements when either one or two parts are withdrawn from stock.

0E. No stock kept on hand but emergency procurement procedures are to be used when an order is placed.

2R. Enough parts are to be purchased by routine procurement methods to bring the stock up to the total amount indicated as the minimum. This minimum is unlikely to apply to items involved in quantity purchase.

0R. The zero minimum is a means of mechanically cross indexing all parts contained in the machines and equipment of a plant. Under this system all machine parts will have an assigned item number which may be coded if desired. This assigned item number is unlikely to be the same as the machine manufacturer's part number. The suffix *E* or *R* indicates the procurement method to be followed if a part is ordered.

4C. A sufficient quantity should be purchased to insure a continuous minimum stock of the indicated quantity. Allowances will need to be made for the requirement procurement time and for usage during the procurement period. Quantity discounts and minimum economical order quantity considerations will also affect order size.

Significant changes in material delivery times will affect the minimum stock required. For example, wartime shortage conditions or a change in status from "manufacturer in stock" to "manufacture upon receipt of order" will require a view of the stock minimum.

When abnormal amounts of material will be required for new equipment installations, the Inventory Control Department should be notified of the impending need as soon as practical so order quantities can be properly scheduled. The Inventory Control records should flag the abnormal quantities required for new construction work so that the correct allowances will be made for normal needs.

*The "Insurance" Minimum*

In many cases the evaluation of need for a specific part under the previously outlined steps will result in the decision that a certain minimum stock is essential to insure the maximum continuity of production operations. The minimum stock decided upon in these cases can appropriately be called "insurance" minimums. Subsequent usage of some of these parts may raise the minimum stock level for the item, but the master inventory control records should be marked with this insurance minimum to insure that the minimum stock level is not reduced below this insurance level.

**15. Periodically review the cost and usage of spare parts for out-of-date models of equipment and machines.** Parts for out-of-date models of machines or small equipment items, such as portable power tools and hoists, are usually higher priced than comparable parts for current production models. Delivery time on such parts is longer, and the total inventory investment and yearly usage costs will be higher than for current model parts. In many cases the current equipment model will also clearly provide improved service life between repairs. Thus, a periodic review of total repair and inventory costs for out-of-date equipment and machines will reveal the point where it is economically practical to replace out-of-date equipment.

## DISPOSAL OF SURPLUS INVENTORY

All items kept in inventory should be periodically evaluated for the continuing need for the item. It is suggested that the usage history of each inventory item be examined at six months intervals to detect (a) those items which have not been issued for the past year, (b) those items which have not been issued for two or more years and (c) those items whose recent usage has declined to the point where the stock level is too large for the current rate of usage.

On items not issued for two or more years, the existance of the machine containing the item and the continuing need for the item should be verified. On items not issued for one or more years, the current stock level should be compared with the insurance minimum for the item. Stock in excess of the insurance minimum should be evaluated for disposal.

On those items where usage has declined sharply, an explanation for the decline should be determined. If usage is unlikely to increase for a long time, the surplus stock should be evaluated for disposal.

*Evaluation for Disposal*

Immediate disposal of surplus inventory is not automatically a good business practice. A series of evaluation steps should be performed to determine if reasonable dollar savings can be expected to result from the disposal of the surplus inventory.

**A — Determine if the original vendor or original manufacturer will accept the return of the material at a reasonable return charge.** Many commercial items such as pipe, pipe fittings, valves and bearings may be returnable to the original vendor who will allow a credit of original cost less a restocking charge. In some cases the original manufacturer will accept the return of items that he is currently manufacturing. Specially manufactured items and parts for noncurrent machine models are rarely returnable for credit.

Occasionally nonreturnable items can be sold commercially by consignment to resale specialists. Salvage and sale of components is sometimes a possibility.

**B — Re-evaluate the desirability of surplus disposal when neither a significant per cent of original cost nor a reasonable number of dollars is recoverable.** When a surplus inventory item can only be sold for a small part of its original cost or for scrap value, the item should be evaluated as to whether it would be an economic purchase if the price was lowered to the recoverable value plus 52 per cent of the difference between the recoverable value and the original cost. Fifty-two per cent of the difference enters the value calculation because the full inventory loss upon disposal of an item is an operating cost which reduces income before taxes. Thus, 52 per cent of the inventory loss is recovered as an income tax reduction. Surplus items subject to significant deterioration during prolonged storage should not be retained in stock.

If a surplus inventory item can remain in a drawer or on a shelf together with the normal amount of inventory of the same item without requiring additional handling costs, and if the space vacated by disposing of the surplus will not be converted to the profitable storage of other materials, then the actual yearly inventory charge to use in evaluating the desirability of retaining the surplus would be not more than 8 per cent per year per dollar of original cost. This 8 per cent charge would cover taxes, insurance, interest, obsolesence, clerical and supervision charges.

In other words a partially filled storage bin or shelf will have the same space charge as a full bin or shelf. If a fixed crib area is used for storing spare parts, the statistical yearly charge per dollar of inventory will rise during periods of low inventory and will fall with an increasing inventory value. Consideration should be given to the effect of a major dollar volume of surplus inventory being disposed of within any specific period because of the effect upon the profit picture of the company for that period.

Anyone having knowledge that an inventory item is to become obsolete should promptly report the information so that disposal arrangements can be completed at the most advantageous time. The best way to dispose of surplus inventory is to not buy too much in the first place. In other words keep inventory acquisition practices under close control.

## MAINTENANCE MATERIALS CRIB OPERATION

An important part of prompt and effective maintenance is adequate availability of the appropriate spare parts and maintenance materials at locations convenient to the Maintenance Departments.

The preceeding discussion was concerned with how to evaluate the need for parts which should be kept in inventory, and the following discussion of maintenance crib operation concerns those items expected to be kept on hand within the manufacturing plant. The small plant will usually operate from a single maintenance materials disbursing crib, but as the plant size increases, the number of cribs will increase to the point that multiple cribs will be supplied from a master crib.

### *Locate Cribs to Minimize Travel Time*

All maintenance material cribs should be conveniently located within the plant area served by the respective maintenance groups in order to minimize the walking or travel time by maintenance men. This time saving also results in lower machine downtime losses. The internal arrangement of the sub-crib should be planned with the thought of reducing travel by the crib attendant. Heavy usage items should be kept near the crib window. Items likely to be used together should be stored close together. Electrical items should be stored in a group separate from pipe items and separate from mechanical items.

### *Know How Much Is Available and Where It Is*

The maintenance material control system should keep its records in such a manner that the exact quantity on hand of each item should be known until the item is issued to the maintenance man. In the case of the master crib-sub-crib system, it will be necessary to keep track of the material in stock in the master crib and in each sub-crib. This multi-crib record keeping increases the clerical operations, but effective maintenance operations can then be supported by a smaller inventory through the redistribution and sharing of available inventory. This also permits a more economical combined volume purchasing of the item for distribution to the various sub-cribs.

Each maintenance sub-crib should contain all spare parts and maintenance materials used in only that plant division serviced by the corresponding maintenance group. Such items should not be stocked in the master crib in order to avoid unnecessary parts delivery delays. Each maintenance sub-crib should also contain in appropriate amounts those spare parts and maintenance materials that have plantwide use, and which may also be stocked in a master crib.

### *Provide Good Indexes of Available Items*

Each maintenance materials crib should be provided with a complete set of indices of all items available from any maintenance materials crib. By thoroughly describing each item at least once in the index listings, cross reference listings under other code words applicable to the item and cross reference listings under the usage classification for the item will make the entire inventory more readily available for multi-application usage.

For example, it is suggested that all items associated with gear reducers should be indexed under "Gear Reducer." If the complete gear reducer is kept in stock, a typical listing could be "348573 Gear Reducer Complete – Model 10RH Ratio 20-1 Right Hand Output Shaft." The separate parts could be typically listed as "348574 Gear Reducer Part – 1151 Output Shaft for Model 10RH Ratio 21-1." All parts kept for the repair of condensate traps could be listed as "656921 Trap – Condensate Return – Part 157 Float for size ¾-in. FT15." This part would also be cross indexed under the key word "Float."

All parts kept for the repair of a specific machine model should be indexed under a common heading. For example, "515303 Lathe Part – 1285 Spindle Gear for Model 12T Serials 125, 185 (X Co. M9100, M9780)." This item would be cross indexed under "Gear." A commercially obtainable bearing for the above lathe would be completely described in the bearings index, and the listing in the lathe parts index would be "152300 Lathe Part – 1287 Spindle Bearing (ND-7200) for Model 12T Serials 125, 157."

Many items are commonly known under several names. Gasket materials may be sheet packing, cork, rubber or asbestos; hence, cross references will be needed under the key words gasket, packing, cork, rubber and asbestos.

### *Keep a Master Index*

A master index should be kept which includes a complete description of the item and a listing of all known machine installations using the item. The machine owners identification number and machine location should also be given. If another stock item can be temporarily used as a substitute, the master index should so indicate. Note that in the typical item descriptions given above, each description begins with an item number such that the item number alone can serve to identify, locate, requisition, purchase, receive and store the item.

### *Crib Indices*

The indices at each crib should be marked with the exact storage location for all items stored in that crib. It is suggested that the master crib index should be marked with the storage locations for all cribs that stock each item. Marking the crib storage locations in the indices permits the maintenance man to write the storage location on the requisition he gives to the crib serviceman, which speeds up crib service and holds down the number of crib servicemen needed.

Marking all crib locations on the master crib index expedites the re-allocation of available stock of all items.

The physical form of the indices can vary to suit the individual company, but the system used should meet the test of (1) being readily updated and (2) being conveniently usable by the personnel who need the indices. It is suggested that the indices be mounted on rotary or swinging carriers along a wall of the crib, in such a manner that the maintenance men can use the indices from one side and the crib servicemen can use them from the other side in order to determine the storage location for incoming items.

It is suggested that the individual item data should be individually removable from the index assembly. Either the individual item strips or the entire index page should be protected by a plastic cover to minimize soiling of the index information. Where automatic machine accounting systems are used for inventory control, the accounting machines can be used to prepare any desired number of duplicate item identification strips for new items or for description revisions to existing items.

For greater accuracy in charging repair parts to the proper machine and to the proper department, it is suggested that area layout maps showing the machine number, machine name and machine model should

be posted near the item indices. These maps often will help the maintenance man find the appropriate group of parts in the indices.

### *A Typical Storage Location System*

It is suggested that the crib storage shelves and bins be marked with a location designation system so arranged that the system will locate an item in a drawer, bin or shelf. This can be done by assigning a section number to each major section of shelves, bins or drawers and to locked cabinets. This major section can be of any length desired. The second step is to assign letters in vertical sequence to each section starting from the floor and proceeding upward with the vertical space assigned to each letter being equal to the minimum height drawer or shelf used in the major section.

If some portion of a major section is equipped with widely spaced shelves only those letters corresponding to the shelf heights will be used. The third location identification step is to number the storage spaces from left to right in each major section. Usually the horizontal storage spaces will be taken as being 4 or 6 in. wide, but other widths can be used as desired.

A similar floor area marking arrangement for locating pallet storage positions could consist of the section number followed by letters to designate the sequential positions starting from the floor for multi-deck pallet or tote box storage, and followed in turn by the numerical sequence of the pallet starting from one end of the row.

### *Mark Identity Number on Most Stock Items and on Racks*

To the major extent practical, each item in stock should be individually and clearly marked with its identification number. This identification marking on each item reduces errors in handling items within the crib, and it helps to accomplish accurate return-to-stock operations when unused items are returned to the crib.

As a parallel to the marking of the individual item in stock, the bin, shelf or drawer locations should also be marked with labels showing the item number, storage location and at least a partial description. These labels help to keep stock items in their proper location, and they help during re-stocking work.

To assist in the proper identification of newly purchased items, the shipment items should be identified by the use of the vendor's shipping list and by comparison with a copy of the purchase order. Vendors should be encouraged to show the purchase item numbers on the shipping invoice, but this information should be checked against the purchase order copy to avoid clerical errors in descriptions.

### *Proper Storage Is Essential*

All items kept in stock should be adequately protected against moisture damage, corrosion damage, mechanical damage, high temperature damage and low temperature damage. Certain items require dry heated indoor storage, others require only unheated indoor storage, or exterior storage.

Fragile items, such as electronic tubes, require careful handling at all times. Many items as received from the vendor will require additional protection and individual boxing to adequately protect the item against damage during storage.

### *Provide Prompt and Dependable Crib Service*

At the crib service window it is suggested that all maintenance and crib personnel be instructed in the use of a preferential service arrangement whereby any maintenance man needing a part for the emergency repair of critical equipment can ask for and receive immediate service.

An adequate number of crib service personnel should be available in order to minimize the in-line waiting for crib service on the part of maintenance men. If one or more maintenance men are consistently waiting in line to present a material requisition to a crib serviceman, additional crib service personnel are needed.

It is suggested that consideration be given to the use of a self service system of withdrawals from stock and return to stock of maintenance stores. For this system to work reliably, charge-out cards containing the material description should be kept beside or near the individual item. In this way the proper inventory record can be charged or credited as items are removed from or returned to the maintenance stores. This self service system is currently being successfully used to control 35,000 different items in the case of one company which employs 260 maintenance men.

For all items stocked in several cribs, the stock in one crib should not be reduced to zero without redistribution of the remaining stock, unless such redistribution leaves an ineffective quantity in each crib. In this case the material control supervisor should evaluate the situation for the most appropriate replacement method.

In all cases prompt and dependable delivery service should be provided between the master cribs and the sub-cribs and between the various sub-cribs. The lack of prompt service at a crib window or the lack of dependable delivery of items from another crib are both conducive to the hoarding of spare parts by maintenance men.

### *Salvage and Return Useable Items to Cribs*

In maintenance operations it is practical to salvage many items as spare minutes work. The material inventory system should permit the convenient return of these salvaged items to the appropriate storage point. The maintenance management must exercise good judgment in selecting the items on which salvage work is profitable. Salvage work should include the removal and replacement of damaged components in stock items such as electrical controls, valves, etc. For example it may be practical to individually purchase the components required to repair a single stock item.

## CONCLUSION

Effective management of maintenance material control and disbursing systems can help a company achieve both lower production costs and improved maintenance operations.

# HIGH STRENGTH ALUMINUM CASTING ALLOY M517-T61

## *A preliminary evaluation*

by W. A. Bailey, E. N. Bossing and F. H. Roebuck


### ABSTRACT

**A preliminary evaluation of M-517-T61 as a high strength aluminum alloy casting material is given. The authors' company is currently procuring structural missile castings of a 356 type aluminum alloy varied by increased magnesium and an addition of beryllium. One of these castings poured in M-517 alloy showed more uniform and higher mechanical properties than those poured in the 356 variant alloy. These results are given with a general discussion of the potential of this experimental alloy.**


### INTRODUCTION

Several structural missile components are presently cast in a 356 variant aluminum alloy to the chemical and mechanical property requirements of the authors' company Specification D.M.S. 1721. This specification, in addition to describing an alloy composition similar to that of high purity A-356 but with increased magnesium and a beryllium addition, requires ultimate strengths of 50 ksi, yield strengths of 40 ksi and 5 per cent elongation within a one in. gage length in test coupons excised directly from critical areas of the casting.

These mechanical properties are optimistic, and border on the upper limit of the present state of light metal foundry art.

The required strength level of the castings presently purchased to Specification D.M.S. 1721 is verified by the tensile testing of several integrally cast coupon blanks fixed to each individual part of each configuration. The technique was used because of the assumption that the area of the casting directly adjacent to the integral bars is of essentially the same strength level as are the tested bars. This assumption was based upon previous experience with smaller less complicated configurations in which mechanical properties varied but little from area to area.

In larger, more complex castings with considerable variation in section size and in degree of chilling, the assumption is clearly invalid. In unpublished work by W. A. Bailey, marked variations have been observed within extremely short distances, especially in ductility, where 15 per cent elongation directly adjacent to a chill has been seen to progressively fall to 3 per cent ¾-in. from that chill.

#### *Interrelated Metallurgical Problems*

Thus, at present, three interrelated metallurgical problems exist for castings produced under the material specification:

1. The problem of consistent foundry conformance with the required mechanical properties.
2. The problem of limited sources for procurement.
3. The problem of lack of correlation between the tensile strengths of the integrally cast test bars and the mechanical properties of coupons excised from the casting matrix.

The simplest theoretical solution to all of these simultaneous problems would be an alloy change, which would permit the pouring of castings to D.M.S. 1721 mechanical properties by foundries of perhaps less technical competence than is now required to properly handle the 356 variant aluminum alloy. In short, ideally, what is needed is a forgiving and more castable aluminum alloy which could meet or better

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the mechanical properties now obtained from coupons excised from the matrix of the high strength light metal alloy castings.

## EXPERIMENTAL ALLOY

Previous unpublished work by W. A. Bailey and E. N. Bossing had indicated that the new experimental M-517 copper-silicon-magnesium-aluminum alloy might have some potential in this direction. The limited tensile strength data obtained from castings were promising. Hypothetically, the nominal 9 per cent silicon would give some increase in fluidity, which would in turn permit a lowering of pouring temperature with a resultant decrease in grain size. This lower pouring temperature should, in theory, also reduce the potential level of included hydrogen and perhaps even slightly lessen the hazard of shrinkage and dross. The copper addition might give added strength at both ambient and elevated temperatures. With all these potential advantages a further investigation of M-517 aluminum alloy was definitely indicated.

Toward this direction, a configuration was selected which had been previously produced in 356 variant aluminum alloy, and for which considerable mechanical property data existed which could be used to compare the relative strengths of the two materials. The selected configuration, a 36 in. diameter ring, is diagramed in the figure. Historically, the part had been poured in limited production to D.M.S. 1721 by Foundry B, and is at present cast to the same specification in almost identical shape under a different part number by Foundry A.

The initial pattern equipment was utilized by Foundry B in pouring several pieces of the ring configuration in M-517 aluminum alloy. Pouring, chilling and gating techniques for these castings were said by Foundry B to be identical to their production of the same part in 356 variant aluminum alloy.

The experimental castings were heat treated according to schedule — solution heat treat 12 hr at 980 F, 180 F water quench, artificial age 6 hr at 340 F. One of these castings was submitted to the company's Materials Research and Process Engineering for evaluation. The present paper describes the results of this evaluation.

## PROCEDURE AND RESULTS

The casting submitted was clean and acceptable on fluorescent penetrant inspection. Radiographs of the casting showed a quality comparable to that obtained from the production castings of this part poured in 356 variant aluminum alloy by both Foundry B and Foundry A. The overall defect level lay somewhere between Grade B and C of MIL-C-6021D.

A spectrographic chemical analysis of the submitted casting is reported in Table 1. Also appearing in Table 1 is the nominal composition of M-517 aluminum alloy as reported by Foundry B, as well as, the composition limits imposed by Material Specification 1721 for 356 variant aluminum alloy. All production castings of the subject configuration shipped to the company by both Foundry B and Foundry A, of which mechanical property data are herein reported, were within these chemical limits.

TABLE 1 — CHEMICAL COMPOSITION, PER CENT

| | Nominal M-517 Aluminum Alloy As Reported | M-517 Aluminum Alloy As Determined | 356 Variant Aluminum Alloy Per D.M.S. 1721 |
|---|---|---|---|
| Magnesium | 0.5 | 0.61 | 0.45-0.75 |
| Silicon | 9.0 | 8.9 | 6.5 -7.5 |
| Iron | — | <0.10 | 0.20 max. |
| Manganese | — | <0.07 | 0.20 max. |
| Zinc | — | — | 0.20 max. |
| Copper | 1.8 | 1.48 | 0.20 max. |
| Titanium | — | 0.105 | 0.20 max. |
| Beryllium | — | <0.001 | 0.20 max. |
| Other Elements, each | — | — | 0.05 |
| Other Elements, total | — | — | 0.15 |
| Aluminum | Remainder | Remainder | Remainder |

The five integrally cast test coupons as well as a number of excised coupon blanks were machined from the submitted M-517-T61 aluminum alloy casting in the locations shown in the figure. These were machined into standard A.S.T.M. Type 2 tensile specimens, and were subsequently tested in tension. A summary of the results of these tensile tests appears in Table 2 for the integral test bars and Table 3 for the excised coupons.

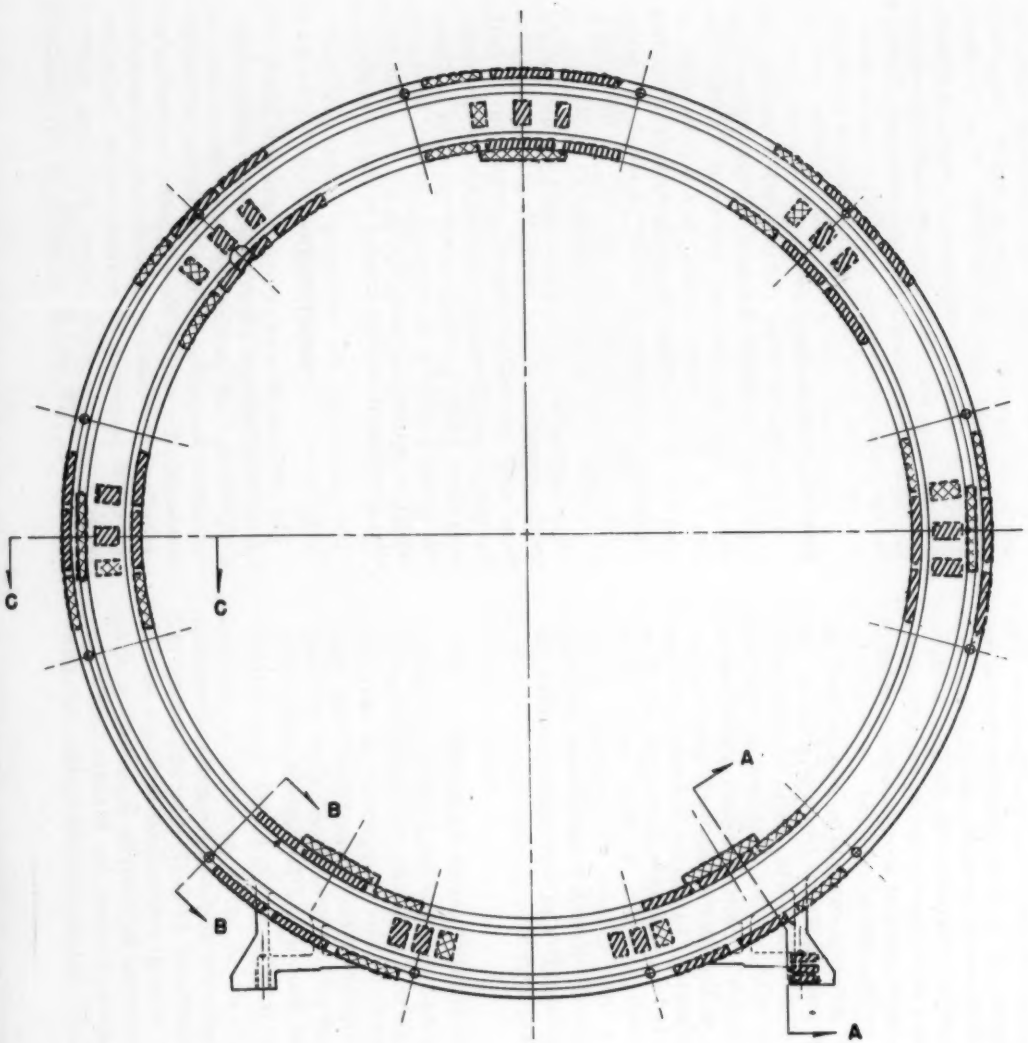
### *Bending Modulus Specimens*

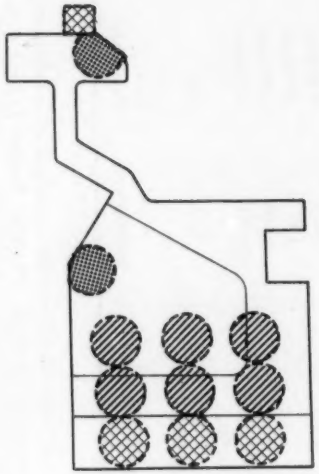
Also machined from the M-517-T61 casting, excised from locations indicated in the figure, were several bending modulus specimens. These were tested in three point loading over a 2 in. span, as described in the paper by J. C. Graddy, "Bending Technique for Evaluation of Cast Materials and Structures." AFS TRANSACTIONS, *vol. 67*, p. 166 (1959). A summary of these results appears in Table 3.

For comparison, Table 2 also lists a summary of the tensile properties of the integrally cast test coupons from parts poured for limited production by both Foundry B and Foundry A in 356-T6 variant aluminum alloy. Table 3 carries a summary of the mechanical property results of coupons excised from the matrix of several of these 356-T6 variant aluminum alloy production castings poured by each foundry.

## DISCUSSION

The observed increase in tensile strength and bending modulus, brought about by a simple alloy change is significant, but perhaps of equal importance to the aircraft industry is the fact that in the M-517-T61 aluminum alloy casting, the mechanical properties of the integral test bars are approximately equal to the mechanical properties of coupons excised from the casting proper. In short, within individual M-517-T61 castings, the tensile strengths of the integral

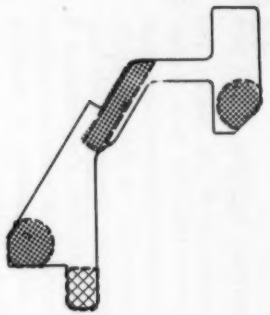


SECTION A-A



SECTION B-B



SECTION C-C

BEND SPECIMENS

TENSILE SPECIMENS

BOTH BEND & TENSILE SPECIMENS

bars may accurately reflect the actual strength of the part which they represent. This, has not been historically true with 356-T6 variant aluminum alloy.

Assuming that the gating, risering and chilling of the M-517-T61 aluminum alloy casting was identical to that used in pouring the 356-T6 variant aluminum alloy production castings and realizing further that the integral test bars on all the castings were chilled, it appears that M-517 aluminum alloy might conceivably be less chill sensitive than is 356 variant. If this reasoning is valid and the limited data reliable, M-517 could potentially have two additional advantages over 356 variant beyond the probable increase in strength level and the possible increase in validity of the integral test bar results.

### *Metallurgical Requirements*

First, the foundry metallurgical requirements might be lessened as the necessity of extreme attention to progressive solidification is reduced. Second, the foundry might more easily meet casting dimensions and tolerances because the placement of a complex chill system sharply increases production difficulties. This should be as attractive to the founder as to the customers because chill blows are always a calculated risk. Also a slightly lower solution heat treat temperature might conceivably reduce the amount of warpage which occurs during water quenching. If the advantages exist, even to a degree, they would be favorably reflected by lower costs in both time and money.

While the M-517 aluminum casting alloy appears to be a most promising material, it must be remembered that any conclusions reached or assumptions made in this report are only tentative and are based on limited data. Much more work must be performed before these observations can be statistically confirmed or rejected. Further, at this moment there is little direct information concerning the foundry behavior of this alloy, its corrosion and stress corrosion characteristics, its heat treat response, fatigue properties and many other factors.

Despite the dearth of such information, M-517 aluminum alloy should be thoroughly investigated. Inherent in the alloy could be, if not a panacea, at least a partial solution to a number of design and procurement problems. It is in this direction that the authors' company is proceeding.

**TABLE 2 — MECHANICAL PROPERTIES OF INTEGRALLY CAST TEST BARS FOR M-517-T61 AND 356-T6 VARIANT IN SIMILAR CAST CONFIGURATIONS**

| | M-517-T61 (Foundry B) | 356-T6 Variant (Foundry B) | 356-T6 Variant (Foundry A) |
|---|---|---|---|
| Number of Tests | 5 | 60 | 35 |
| Ultimate Tensile Strength, ksi | | | |
| Mode | — | 50.5 | 49.4 |
| Median | 55.6 | 51.0 | 49.4 |
| Mean | 54.7 | 50.0 | 49.0 |
| Maximum | 58.9 | 55.5 | 55.7 |
| Minimum | 49.4* | 44.6 | 45.3 |
| Yield Strength, ksi | | | |
| Mode | — | 40.2 | 40.1 |
| Median | 43.7 | 41.0 | 41.1 |
| Mean | 44.1 | 41.2 | 41.4 |
| Maximum | 45.1 | 45.7 | 46.0 |
| Minimum | 43.2* | 37.5 | 37.2 |
| Elongation, % in one in. | | | |
| Mode | — | 5.0 | 3.0 |
| Median | 4.5 | 4.0 | 5.0 |
| Mean | 4.3 | 4.3 | 5.2 |
| Maximum | 7.5 | 10.0 | 10.0 |
| Minimum | 1.5* | 0.5* | 1.0 |

*Inclusion

**TABLE 3 — MECHANICAL PROPERTIES OF COUPONS EXCISED FROM THE MATRIX OF M-517-T61 AND 356-T6 VARIANT ALUMINUM ALLOYS IN SIMILAR CAST CONFIGURATIONS**

| | M-517-T61 (Foundry B) | 356-T6 Variant (Foundry B) | 356-T6 Variant (Foundry A) |
|---|---|---|---|
| Number of Tests | 22 | 46 | 52 |
| Ultimate Tensile Strength, ksi | | | |
| Mode | 56.3 | 47.6 | 48.8 |
| Median | 54.6 | 47.2 | 49.0 |
| Mean | 55.0 | 46.8 | 50.1 |
| Maximum | 58.9 | 53.3 | 54.0 |
| Minimum | 48.1 | 36.4 | 44.6 |
| Yield Strength, ksi | | | |
| Mode | 44.0 | 43.2 | 43.5 |
| Median | 44.0 | 42.4 | 42.4 |
| Mean | 43.9 | 42.1 | 42.2 |
| Maximum | 46.9 | 44.2 | 46.2 |
| Minimum | 41.7 | 37.1 | 38.6 |
| Elongation, % in one in. | | | |
| Mode | 4.5 | 2.5 | 2.0 |
| Median | 4.5 | 2.5 | 2.0 |
| Mean | 4.5 | 2.8 | 2.6 |
| Maximum | 9.5 | 6.5 | 6.0 |
| Minimum | 1.0* | 1.5 | 1.0 |
| Bending Modulus, | | | |
| Number of Tests | 44 | 52 | 23 |
| Mode | 103.0 | 89.1 | — |
| Median | 102.7 | 84.7 | 90.0 |
| Mean | 101.8 | 84.9 | 90.9 |
| Maximum | 114.3 | 100.8 | 108.6 |
| Minimum | 82.6 | 66.2 | 78.5 |

*Inclusion

## CONCLUSIONS

As can be seen by comparing the summary of results in Tables 2 and 3, two observations are evident —(1) M-517-T61 aluminum alloy appears to have a marked mechanical property advantage over 356-T6 variant aluminum alloy cast rings and (2) the integral test bars fixed to the M-517-T61 aluminum alloy casting appear to more accurately reflect the strength of the matrix of the casting than do the integral bars of the 356-T6 variant aluminum alloy castings.

## ACKNOWLEDGMENT

The authors wish to thank the Aluminum Company of America Research Laboratories for their work in the development of this alloy, and for supplying the test castings used in this research.

# ESTABLISHING WORK STANDARDS BY SAMPLING

by E. C. Keachie


## ABSTRACT

**The author offers an introduction to work sampling giving a comparison of this method with that of various other methods of measuring work. These methods are —stopwatch timestudy, standard data and predetermined work times. The ideas and practices of work sampling are simple and adaptable for obtaining data on the shop floor. It is statistically sound for getting representative samples and for using these samples as a basis for cost control and work improvement.**

**This introduction to work sampling was presented with an explanatory film as a contribution of the AFS Industrial Engineering and Cost Committee.**


## WHAT AND WHY

Work sampling is a simple but effective method of determining the time of production operations as a basis for cost reduction, control and general improvement. It is comparatively new, and is increasing in popularity because of its advantages over other work measurement methods. This is especially true where management has relied too much on history and estimate.

In brief, work sampling consists of making random observations of what is done, sometimes with refinements as to the exact nature of the work and performance rating. Observations of several jobs can be made by one observer; the sequence and timing are planned to ensure that the samples taken are truly representative. The total calendar time involved for a single study usually varies from a few days to two weeks.

Results are often startling in revealing the true situation. For example, the total of nonproductive time may be as much as 30 per cent. Corrective action is usually apparent, and the nature of the technique is such as to facilitate making needed changes through employee cooperation.

## COMPARISON OF METHODS

A quick look at other means of measuring work will help place each in perspective. To begin with, all management has some idea of what constitutes a fair day's work, if only from habit or vague experience. Next, there are historical records on which both management and labor may rest content, if not efficiently. Often these data are the basis for bids. The data may be highly refined, but may or may not be based on effective methods and standards work.

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Accepted industrial engineering practice includes methods analysis plus stopwatch timestudy, standard data and predetermined work times. These can give precision results, and each had advantages in particular applications.

Stopwatch timestudy is widely used in work measurement. Begun by F. W. Taylor in the late 19th century, it is now a precision tool. Given a thorough methods description, timing by elements and careful setting of allowances, it yields a standard time for operations based on actual work done. Work sampling is also based on actual observation of work done, extended to cover all events during the typical day.

Standard data consists of elemental times for new items that are based on the bank of data already established by stopwatch for similar items. Thus, an operation on a new size of a previously made piece can be timed by reference to standard times on other sizes by interpolation.

Predetermined human work times are operation standard times made up of the sum of ideal component times for each movement involved. Several systems have been developed and are commercially promoted. They have the virtue of requiring a good methods description and close attention as a basis for detailed planning and estimating.

In addition to the direct methods of securing standard times, one should be alert for other indications that may affect the total picture. For example, the manufacturing progress function or so-called learning curve may contribute to improving efficiency over periods of months or longer, with a consequent reduction in manhours per unit produced. This is especially true where designs are complex and subject to fairly rapid change. Numerical or graphical indications of such reduction might indicate the need for new work sampling.

## SUMMARY

The ideas and practices of work sampling show simplicity and adaptability of the technique on the shop floor. The process is a simple but sound statistical basis for getting representative samples and using them as a basis for cost control and general improvement.

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61-126

# STEEL CASTING REPAIR BY FLUX CORED $CO_2$ WELDING PROCESS

by M. B. Prisuta


## ABSTRACT

**This report covers a welding process that makes possible significant savings for the steel foundryman in casting repair. The paper discusses (1) The principle of the $CO_2$ flux cored welding process and the equipment used and (2) The use of the process for repair welding of mild and alloy steel castings. A discussion on no. 2 gives appropriate data of welding speeds, deposition rates, labor and time savings and physical properties obtained. The range and type of steel castings that can be welded with this process are covered.**


## INTRODUCTION

The author's company began development work in the 1940's on a flux cored $CO_2$ welding process. The process, following its commercial introduction, has been used successfully throughout industry to weld all kinds of steels, and the weld metal obtained has satisfactory mechanical and chemical properties. The process has been used for casting repair work in a few steel foundries. The results of this work indicate that this process can provide some real savings for foundries.

The flux cored wire used with the process, becomes a filler metal for casting voids and cavities caused by sand erosion and other casting problems. Generally speaking, costs and welding time are cut almost in half, as compared to stick electrode methods formerly used. Savings from 40 to 65 per cent are realized. The weld deposition rates are significantly higher than those obtained with stick electrode.

The purpose of this report is to tell the foundry industry about this process, and how it has helped the few foundries in which it has been used.

## BASIC ELEMENTS OF PROCESS

The flux cored $CO_2$ welding process uses the basic ingredients of (1) a continuously fed, flux cored electrode, (2) carbon dioxide gas and slag as shields and (3) suitable welding gun, control unit, and welding machine. All of these will be explained in detail later.

The flux cored electrode gets its name from the fact that the flux compounds are added to welding wire, which has been formed into a tubular shape. The wire is closed completely after all of the compounds have been added. These compounds include ionizers, which stabilize the welding arc, deoxidizers, which prevent gas porosity and slag formers, which perform a cleaning action and provide a slag shield. Figure 1 shows the compounds on the inside of the electrode.

The carbon dioxide shields the arc. None of the excessive spatter associated with $CO_2$ bare wire welding, occurs with this process. The carbon dioxide gas also costs much less than other gases which might be used for other types of repair welding. This cost factor will be brought out in detail later. The power source is a constant voltage-type direct current welding machine. Reverse polarity is used with this process.

## ELECTRODE TYPES AVAILABLE

There are four types of electrodes available for use with the process:

1) The no. 110 electrode, used for welding plain carbon steels having a maximum carbon content of 0.35 per cent. This electrode can also be used to weld steel plate covered with mill scale and rusty plate. It is considered a single-pass electrode, even though it can be used quite successfully up to a maximum of 3 passes.
2) The no. 111A electrode is also used to weld plain carbon steels, but it is a multiple-pass electrode. It has been certified by the American Bureau of Shipping to meet A.S.M.E. Boiler Code Specification A-233.
3) The no. 150 electrode is used to weld low alloy, high tensile steels like T-1 and A.I.S.I. 4130. It is a multiple-pass electrode.
4) The no. 151 electrode is also used to weld low

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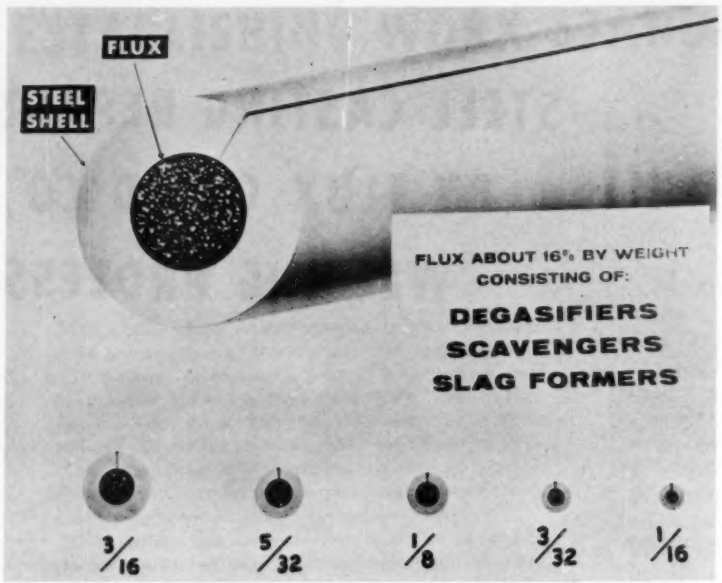



Fig. 1 — Flux compounds on inside of electrode.

| A.I.S.I. classification | Use electrode |
|---|---|
| Carbon steels 1005-1029, 1030-1040 | 110 and/or 111A |
| Ni-Cr steels 3115-3130, 3135-3150<br>Mo steels 4017-4028, 4032-4068<br>Cr-Mo steels 4130-4132, 4135-4150<br>Chromium steels 5120-5130, 5132-5152<br>Cr-V steels 6120<br>Ni-Cr-Mo steels 8615-8627, 8630-8660-8720-8735-8750-9747-9763<br>Mn-Ni-Cr-Mo steels 9437-9445 | 150 and/or 151 |

Fig. 2 — Steels weldable by the flux cored $CO_2$ welding process.

alloy, high tensile steels, particularly those used in code work. It also is a multiple pass electrode.

## STEELS WHICH CAN BE REPAIR WELDED

Figure 2 shows a list of steels which can be repair welded by the flux cored $CO_2$ welding process, as well as the electrodes which are used.

## EQUIPMENT USED

As mentioned previously, the equipment used with the process includes a flux cored electrode, carbon dioxide gas, welding gun, control unit and welding machine. The electrode is fed through a semi-automatic gun. The gun is water cooled to withstand the high heats developed during welding. The electrode feed rate is controlled by a speed regulating governor on the control unit.

The feed rate of the electrode is not dependent upon arc voltage or arc current, but is set independently by the operator for the correct operational speed required. The trigger on the gun is a latching type, and controls the starting and stopping of the welding operation. Various size gas cups are available for use with the gun, and are specially designed to insure that a proper gas shield is provided.

The circuity of the manual gun control is arranged to provide touch starts. It is only necessary to have the wire projecting about 3/8-in. from the cup of the welding gun and touch it to the work. These actions then take place:

1) The shielding gas starts to flow (the water is always on).
2) The previously chosen welding current burns off the electrode.
3) The wire feed motor starts and feeds the electrode to the arc at the preset speed.

The constant potential direct current supply is best suited to this operation. Once the proper settings of the equipment have been made, a reasonably wide range of welding voltages may exist across the arc without having them seriously vary the current supplied to the welding arc. The welding machines used with this equipment are of the constant voltage type.

### *Carbon Dioxide*

The carbon dioxide gas is supplied from cylinders or bulk storage containers. The quantity of $CO_2$ used is only the amount required to maintain a shield around the welding zone. The average flow-rate is from 25 to 45 cu ft/hr. The $CO_2$ should be welding grade, having a dewpoint of −40 F or lower.

Figure 3 shows the gun, control equipment, welding machine and $CO_2$ cylinders used with the flux cored $CO_2$ welding process.

### *Welding Technique*

The repair welding technique used with this process is relatively simple. The operator need only move the welding gun at the correct speed along the void to be filled. When the weld is completed, he simply releases the trigger on the gun to stop the weld. Only flat, horizontal and semi-vertical welds can be made with the present equipment available.

Table 1 shows the amperage and voltage figures for the different sizes of electrodes used. The amperage

and voltage used also depends upon the thickness of the casting on which the repair welding is done; hence, the range of figures given. The wire feed, travel speed and $CO_2$ flowrates are also given in this table.

TABLE 1 — OPERATING DATA FOR REPAIR WELDING

| Electrode Size in In. | Amps | Volts | Wire Feed in In./Min | Travel Speed in In./Min | $CO_2$ Flow-rate in Cu Ft/Hr |
|---|---|---|---|---|---|
| 3/32 | 350 | 26 | 105 | 40 | 45 |
| | to | to | to | to | |
| | 620 | 29 | 264 | 44 | |
| 1/8 | 640 | 29 | 240 | 36 | 45 |
| | to | | to | to | |
| | 680 | | 269 | 45 | |
| 5/32 | 650 | 30 | 95 | 39 | 45 |
| | to | | to | to | |
| | 750 | | 122 | 22 | |
| 3/16 | 800 | 27 | 81 | 27 | 45 |
| | to | to | to | to | |
| | 950 | 29 | 102 | 54 | |

### *Stress Relieving*

In the case of the low carbon, unalloyed steel castings repair welded with this process, stress relieving is not essential. The high carbon and alloy steel castings normally require stress relieving after welding. The stress relieving treatment usually involves heating the casting slowly to a temperature of 1100 to 1200 F (593 to 648 C). This temperature range should be held for a period equivalent to one hr/in. of heaviest section. Air cooling or water quenching should follow the stress relieving.

As indicated in the tables showing mechanical data on the weld metal, mechanical properties obtained by using this process are satisfactory with or without stress relieving for the steels tested.

## ECONOMICS OF THE PROCESS

The author's company's engineers conducted some tests at a steel foundry, Foundry *A*, in order to obtain quantitative information on the economics of the flux cored $CO_2$ welding process. This foundry was using an inert gas shielded process to repair weld their castings. The castings welded at Foundry *A* had carbon contents of 0.20 to 0.30 per cent and thicknesses ranging from 1/4 to 6 in. A step-by-step analysis of the process was made during the test period. Three factors were considered—the gas, weld quality and cost analysis.

### *The Gas*

The use of inexpensive carbon dioxide as the shielding gas was of interest to Foundry *A*, but they were skeptical. They had tried $CO_2$ before with a bare wire welding process and had experienced excessive spatter. However, the flux cored process does not use a bare wire, but a flux cored wire, which eliminates the spatter. The tests conducted indicated that good gas shielding without spatter could be obtained using the carbon dioxide gas.

### *Weld Quality*

The next factor that had to be studied carefully was that of weld quality. Several large carbon steel castings were welded at Foundry *A*, with these results:

1. Penetration was good.
2. Porosity was virtually eliminated.
3. No spatter on welds.
4. Fast deposition rates.
5. Entirely satisfactory welds.

Tests were then run using the inert gas welding process, in order to compare the quality of welds made using the two processes. Virtually the same results were obtained, except that the deposition rates for the flux cored process were greater than those obtained with the inert gas process. This meant that the flux cored process now had two economic advantages, by virtue of the inexpensive gas and higher deposition rates.

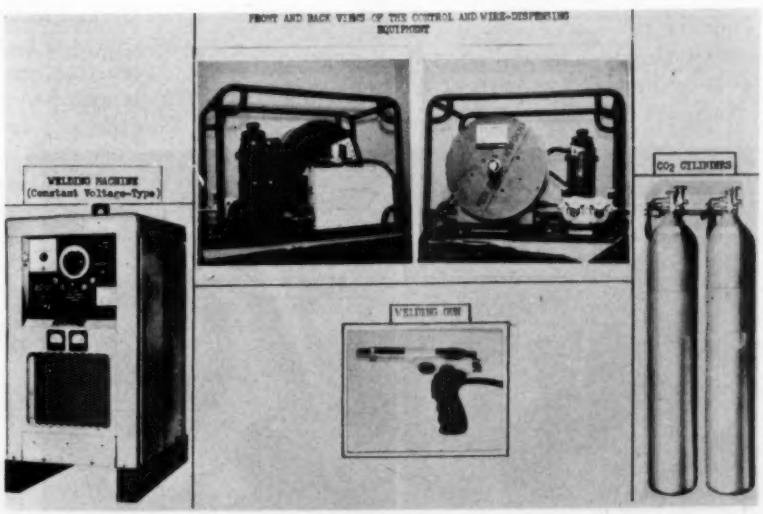



Fig. 3 — Equipment used with the flux cored welding process.

TABLE 2 — DATA OBTAINED AT FOUNDRY A — FLUX CORED $CO_2$ vs. INERT GAS PROCESSES

| | Test One | | Test Two | |
|---|---|---|---|---|
| | Flux Cored | Inert Gas | Flux Cored | Inert Gas |
| Arc time/hr, in min | 29 | 40 | 31 | 36 |
| Weight of Metal Deposited, lb | 6.28 | 5.92 | 6.64 | 5.36 |
| Gas Flowrate, cfh | 35-40 | 40 | 35-40 | 40 |
| Inert Gas Cost at 12c/cu ft, $ | — | 3.20 | — | 2.98 |
| $Co_2$ Cost at 1½c/cu ft, c | 0.27 | — | 0.29 | — |
| Cost of wire used with inert gas process at 35.5c/lb, $ | — | 2.09 | — | 1.90 |
| Cost of flux cored wire at 28c/lb, $ | 1.76 | — | 1.86 | — |
| Labor Costs, $ | 2.32 | 3.20 | 2.48 | 2.88 |
| Total Costs, $ | 4.35 | 8.49 | 4.63 | 7.66 |
| Approximate cost/lb of metal deposited, $ | 0.69 | 1.43 | 0.70 | 1.42 |
| Difference in cost/lb of metal deposited, c | 0.74 | | 0.72 | |

A minimum savings of about $0.73/lb of metal deposited could be realized by using the flux cored $CO_2$ welding process. This would amount to a savings of about $1150/month or about $13,800/year, based on 20 working days per month and average monthly wire consumption for the operation of two units on two shifts per day. The flux cored electrode size is 3/32-in.

### *Cost Analysis*

In order to get a completely accurate economic picture for the flux cored $CO_2$ welding process, it was decided to run two more tests comparing it with inert gas processes. The object of these particular tests was to obtain and list all the pertinent operating data which would permit a valid evaluation. Welding was done with a 3/32-in. electrode on the same type of casting in each test. Table 2 gives all of these data. As a result of these tests, Foundry *A* decided to use the flux cored $CO_2$ welding process to repair their castings on a production basis.

## FOUNDRIES USING THE PROCESS

In order to obtain information on actual production use of the process in steel foundries, three other foundries (using the process on a regular basis) were visited, foundries B, C and D.

### *Foundry B*

This foundry uses the no. 110 electrode to repair weld castings used as parts of machinery frames and air cylinder trunnions. These castings vary in thickness from 1 to 2½-in. A very clean flat weld is obtained, requiring little grinding. Savings of 50 to 60 per cent have been realized over the stick electrode method formerly used. These savings are due to the much higher deposition rates obtained with the flux cored electrode, as well as the low cost of the $CO_2$.

Figure 4 shows two representative castings made by Foundry *B* and repaired with the flux cored $CO_2$ welding process.

### *Foundry C*

This foundry makes castings for the rubber, tool and gear industries. The castings repaired are C-1026, C-1045 and 4037 steels and vary in size from ¼ to 10 in. in thickness, and from one to 5 ft. in diameter. Formerly, these castings were repaired using stick electrode. Now the no. 111A flux cored electrode is used, with good results. The mechanical properties of the finished welds are approximately:

| Mechanical Properties of Weld Metal (as-welded) | |
|---|---|
| Tensile Strength, psi | 60 to 70,000 |
| Yield Strength, psi | 35 to 40,000 |
| Elongation in 2 in., % | 30 |
| Reduction of Area, % | 45 |
| Brinell hardness, 3000-kg load | 160 |

Figures 5, 6, 7 and 8 show photographs of some of the castings repair welded with the flux cored $CO_2$ welding process at Foundry *C*.

### *Foundry D*

This foundry uses the flux cored $CO_2$ welding process to repair castings used for diesel locomotives, machine tools, rolling mills and railroad frogs. The 111A electrode (3/32-in. size) is used to repair weld these castings, which are in the S.A.E. 1020, 1030 and 1040 range, and are all sizes from one in. up. At the present time, Foundry *D* is also conducting tests using the 151 electrode to repair weld high tensile nickel castings.

This foundry formerly used stick electrode welding to repair their castings. One of their most important reasons for changing to the flux cored $CO_2$ process can best be illustrated by an incident, which occurred after they had been using the process. A casting repair job that was being done using the flux cored $CO_2$ process had to be stopped because the foundry had

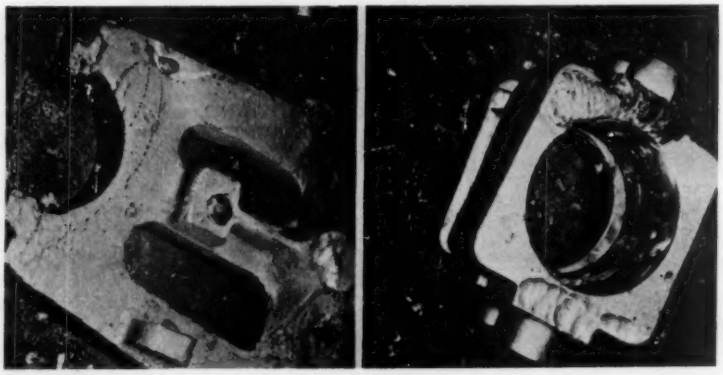
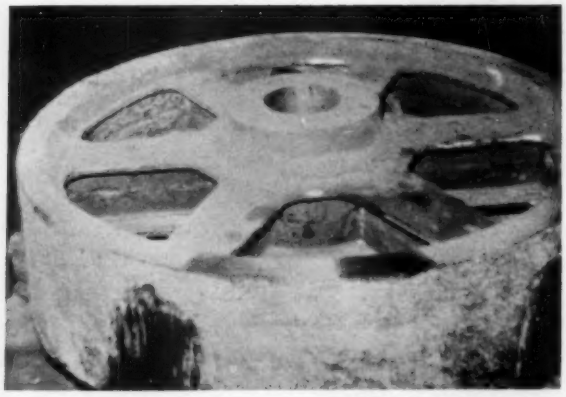
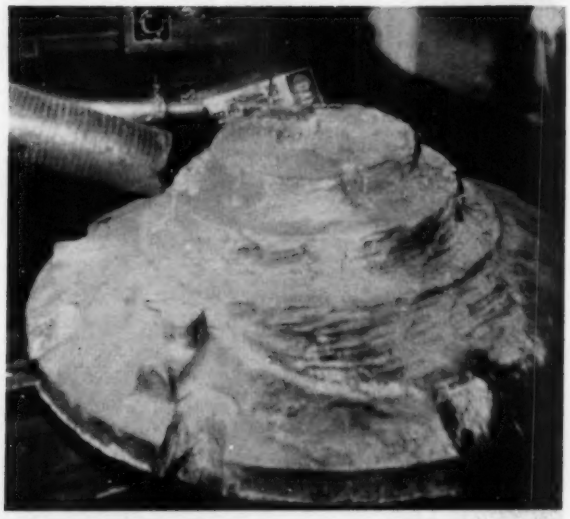


Fig. 4 — Two representative castings repaired with flux cored welding process (from Foundry *B*).









**Figs. 5, 6, 7 and 8 — Castings repair welded with flux cored $CO_2$ welding process at Foundry *C*.**

temporarily run out of the flux cored welding wire. In order to get the job done, stick electrode welding had to be used. Three welders had to do this work in order to keep up with the welding rate produced by just one welder using the flux cored process.

The welds obtained meet the physical specifications for steels in the S.A.E. 1020, 1030 and 1040 range. These specifications are about the same as indicated for the 111A electrode given previously.

Costs in this repair welding operation have virtually been cut in half as a result of using the flux cored $CO_2$ welding process.

Figure 9 shows a typical Foundry *D* casting on which flux cored repair welding is used. Figure 10 shows this casting being welded, and Fig. 11 shows the finished weld on this casting.

Examples of the depth of voids on castings repaired with the flux cored process are shown in Fig. 12. This picture of a casting made at Foundry *D* was taken just prior to repair welding with the flux cored process. The resulting welds adequately filled the voids.

**TABLE 3 — 110 ELECTRODE SINGLE-PASS PLAIN CARBON STEEL**

| A285, ½-in. Plate, As-Welded | | |
|---|---|---|
| Physical Properties | 1 Pass | 3 Passes |
| Tensile Strength, psi | 86,000 | 90,000 |
| Yield Strength, psi | 80,000 | 81,500 |
| Elongation in 2 in., % | 30 | 26 |
| Reduction of Area, % | 52 | 51 |
| Brinell Hardness, 3000 kg Load | 185 | 230 |

**TABLE 4 — 111A ELECTRODE, MULTIPLE-PASS PLAIN CARBON STEEL**

| A285, ¾-in. Plate, As-Welded | |
|---|---|
| Physical Properties | Values |
| Tensile Strength, psi | 72,000 |
| Yield Strength, psi | 60,000 |
| Elongation in 2 in., % | 28 |
| Reduction of Area, % | 50 |
| Brinell Hardness, 3000 kg Load | 162 |

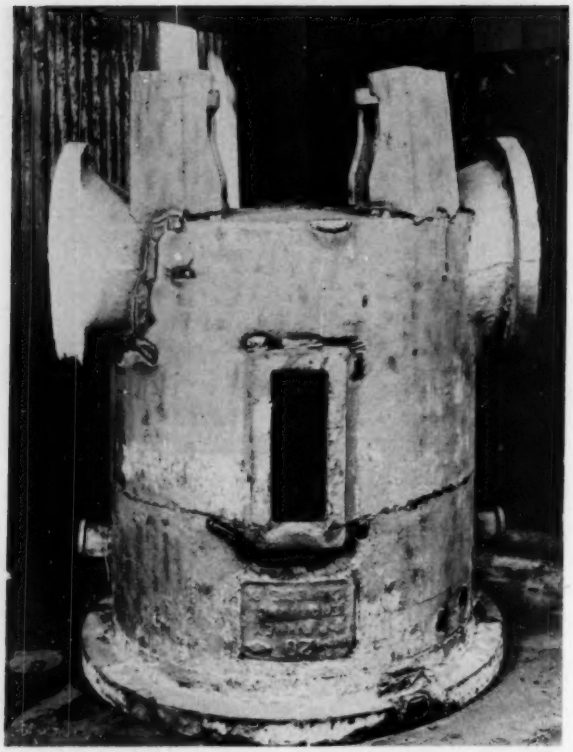


Fig. 9 — Casting flux cored repair welded at Foundry *D*.

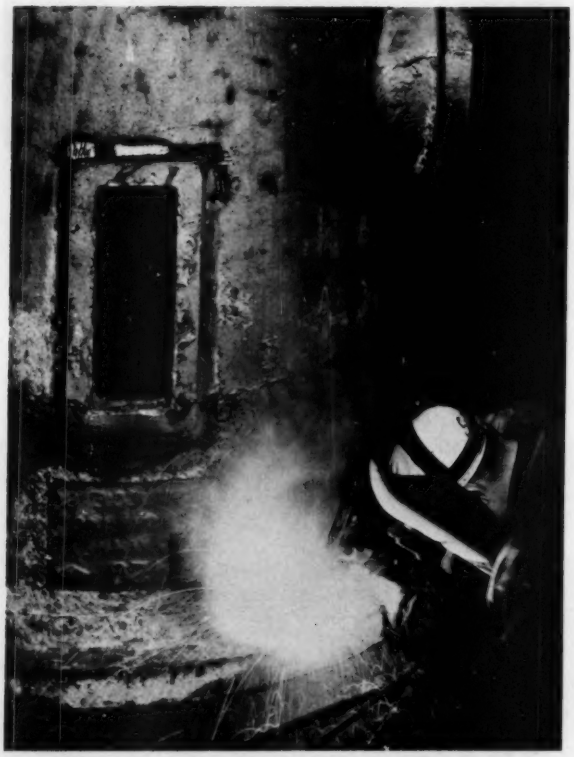


Fig. 10 — Same casting as Fig. 9 in process of being welded.



Fig. 11 — Finished weld on casting shown in Fig. 9.



Fig. 12 — Depth of voids repaired with flux cored process.

The railroad frogs previously mentioned are shown in Fig. 13, and some finished welds on a frog are shown in Fig. 14.

## MECHANICAL DATA ON WELD METAL

Tables 3 through 6 show the mechanical properties obtainable with each of the four electrodes available. Table 7 shows the chemical compositions of the weld deposits obtained, and Fig. 15 charts the impact values of the weld deposits.

## SUMMARY

The flux cored $CO_2$ welding process offers these advantages to the steel foundryman:

1) Higher metal deposition rates.

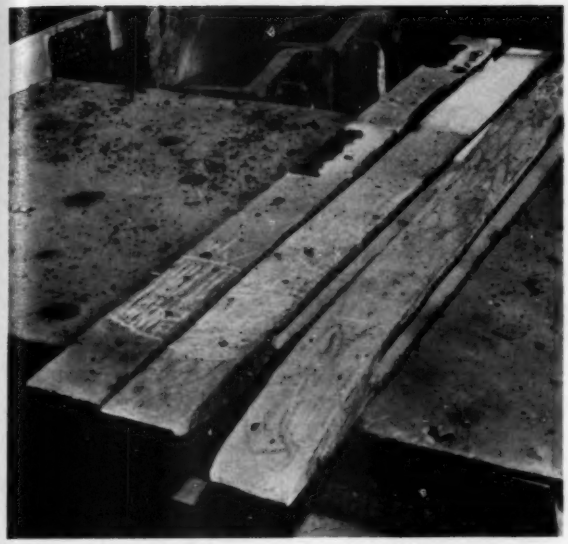


Fig. 13 — Railroad frogs to be repair welded.

| | | ELECTRODE | | | |
|---|---|---|---|---|---|
| | | 110 | 111A | 150 | 151 |
| AT | 70F | 50 | 85 | 65 | 19 |
| AT | 0F | 20 | 51 | 45 | 13 |
| AT | -25F | 14 | 45 | 38 | 8 |
| AT | -50F | 11 | 10 | 20 | 8 |

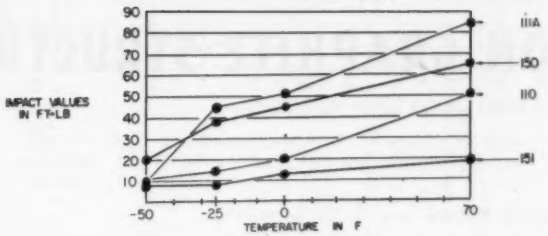



Fig. 15 — Impact values in ft-lb, Charpy V-notch, of weld deposits.

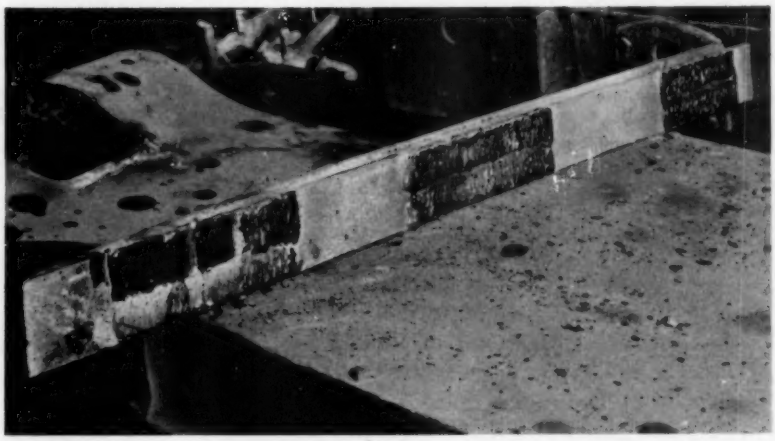


Fig. 14 — Railroad frogs showing finished welds.

TABLE 5 — 150 ELECTRODE, MULTIPLE-PASS LOW ALLOY HIGH TENSILE STEEL

| Physical Properties | A285 ¾-in. Plate Stress Relieved | T-1 ¾-in. Plate As-Welded | T-1 ¾-in. Plate Stress Relieved |
|---|---|---|---|
| Tensile Strength, psi | 88,000 | 108,500 | 107,750 |
| Yield Strength, psi | 76,000 | 97,000 | 96,500 |
| Elongation in 2 in., % | 22.5 | 22 | 24 |
| Reduction of Area, % | 55 | 45 | 51 |
| Brinell Hardness, 3000 kg Load | 228 | 228 | 228 |

2) Low-cost shielding gas (carbon dioxide).
3) Faster welding rates.
4) Savings in time and labor.
5) Ease of using the process.
6) Weld deposits with satisfactory mechanical properties.

TABLE 6 — 151 ELECTRODE, MULTIPLE-PASS LOW ALLOY (1¼ $C_R$-½ $M_{OLY}$) HIGH TENSILE STEEL

| Physical Properties | A285 ¾-in. Plate Stress Relieved |
|---|---|
| Tensile Strength, psi | 130,000 |
| Yield Strength, psi | 100,000 |
| Elongation in 2 in., % | 16 |
| Reduction of Area, % | 40 |
| Brinell Hardness, 3000 kg Load | 283 |

TABLE 7 — CHEMICAL COMPOSITION OF WELD DEPOSITS ON A285 ½-IN. PLATE

| Element | 1 Pass 110, % | 3 Passes 110, % | 111A, % | 150, % | 151, % |
|---|---|---|---|---|---|
| Carbon | 0.10 | 0.08 | 0.08 | 0.05 | 0.14 |
| Manganese | 1.29 | 1.73 | 0.85 | 1.50 | 0.70 |
| Phosphorus | 0.02 | 0.02 | 0.02 | 0.02 | 0.02 |
| Sulfur | 0.03 | 0.03 | 0.02 | 0.02 | 0.02 |
| Silicon | 0.82 | 1.06 | 1.40 | 0.27 | 0.60 |
| Chromium | — | — | — | — | 1.25 |
| Molybdenum | — | — | — | 0.50 | 0.50 |

# HEAVY DUCTILE IRON CASTINGS COMPOSITION EFFECT ON GRAPHITE STRUCTURE

by I. Karsay and R. D. Schelleng


## ABSTRACT

**Large hypoeutectic ductile iron castings, requiring several hours to freeze, contain coarse and irregularly shaped graphite particles. However, true spheroids of graphite form from hypereutectic ductile iron melts, even if the solidification extends over a period of many hours. Ferrosilicon inoculation is essential to the production of spheroidal graphite structures in heavy castings. The improved microstructure achieved in inoculated, hypereutectic, slowly cooled ductile iron castings is reflected in better mechanical properties.**


## INTRODUCTION

It is generally known, that the slow solidification of heavy castings results in coarser microstructures and poorer properties than are encountered in more rapidly cooled samples of the same composition. This is particularly true in gray irons. Coarse graphite flakes occur in large gray iron castings, and the ultimate strength of such irons might be as low as 15,000 psi or less because of this coarseness. The strength of ductile iron also is often considerably lower in heavy castings than in light castings, not only because of the coarsening of the microstructure but also because the graphite particles tend to lose their spheroidal shape.

Figure 1 shows the graphite structure found in a ductile iron ingot mold. The ingot produced in this mold measured 18 x 18 x 40 in. The mold wall was 6 in. thick. Although the graphite particles have a radial structure, they are large and irregularly shaped.

Although the coarseness of the microstructure of ductile iron in heavy castings was expected, no explanation could be found for the deterioration of the shape of the graphite particles. Therefore, a study was made to determine what factors control the graphite shape in slowly solidified ductile irons, and to develop methods of improving the structure and properties of such irons.

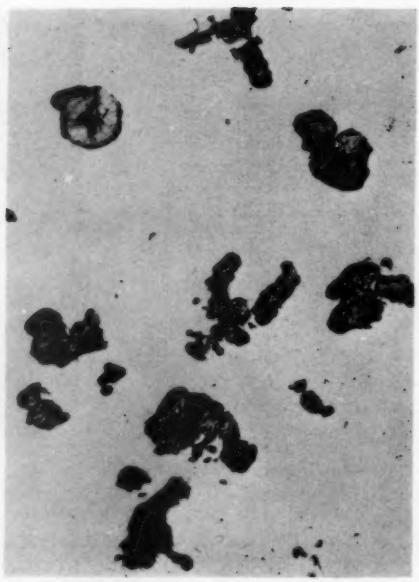


**Fig. 1 — Microstructure of ductile iron ingot mold for an 18 x 18 x 40 in. ingot. Mold wall — 6 in. thick. Composition, per cent — TC — 3.46; Si — 1.40; Mn — 0.28; P — 0.048; Mg — 0.13. Unetched. 50 ×.**

## PROCEDURE

Several test castings varying in size from 1 x 1 x 1 to 10 x 10 x 10 in. were made to determine the effect of solidification rate on the microstructure and mechanical properties of hypoeutectic ductile irons of the composition normally employed in heavy castings. The nominal composition of these castings was:

| TC, % | Si, % | Mn, % | Mg, % |
|---|---|---|---|
| 3.3 | 2.0 | 0.4 | 0.07 |

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The carbon equivalent of this composition is significantly lower than the value of 4.3 usually used in light castings. Platinum-platinum-10 per cent rhodium thermocouples in quartz protection tubes were inserted in the castings, and the time intervals between pouring and the end of solidification were determined.

Large irregular graphite particles similar to those observed in large commercial castings were found in the 10-in. casting. A time interval of about 2¼-hr was required for the completion of solidification in this casting. A 7 in. cube casting poured in a preheated, thermally insulated mold was found to solidify in about the same length of time and was used for most of the other experiments.

It was thought that the primary dendrites in a hypoeutectic iron, between which the graphite is nucleated, might restrict the normal spherical growth of the graphite particles. Therefore, many melts of hypoeutectic, eutectic and hypereutectic compositions were studied. The carbon equivalents were varied through changes in both carbon and silicon contents. Several melts with varying magnesium and manganese contents also were examined.

All the laboratory melts were made in an indirect arc furnace. The molten alloy was heated to 2900 F (1593 C), then tapped and cooled to 2700 F (1482 C).

At this temperature the iron was poured onto the magnesium alloy and immediately reladled again for ferrosilicon inoculation. The magnesium treatment was performed using nickel-magnesium alloy, with the exception of a few heats indicated in Table 2 for which a ferrosilicon-base alloy was employed. Generally 0.25 per cent ferro$_{85}$ silicon (calcium bearing) was added as inoculant except if otherwise indicated.

Chilled samples for carbon analysis were taken in order to determine the carbon equivalent of the melts at the time of pouring. Samples for chemical analyses were also taken from the slowly cooled cubes at several points along a vertical line aiming for an average but not using the exterior one in. area at the top and bottom. Examinations of the microstructures of the castings were carried out on samples removed from the most slowly cooled portions of the test blocks.

## RESULTS AND DISCUSSION

### *Properties of Hypoeutectic Irons*

The time intervals required for solidification of ductile iron castings varying in size from a one in. cube to a 10 in. cube are shown in Fig. 2. The solidification time of one in. cube is only a few minutes, while a 10 in. cube requires about 2½-hr to solidify completely. These data are believed to be applicable for both gray and ductile irons, with slight deviations depending upon pouring temperature.

The size of the graphite particles observed in the test castings increases with longer solidification times, as shown in Fig. 3. The graphite size increases rapidly to 0.0068 in. in the 6 in. cube. In the larger cubes the rate of increase of the size of the graphite particles is not as rapid, resulting in a particle size of 0.0087 in. in the 10 in. cube.

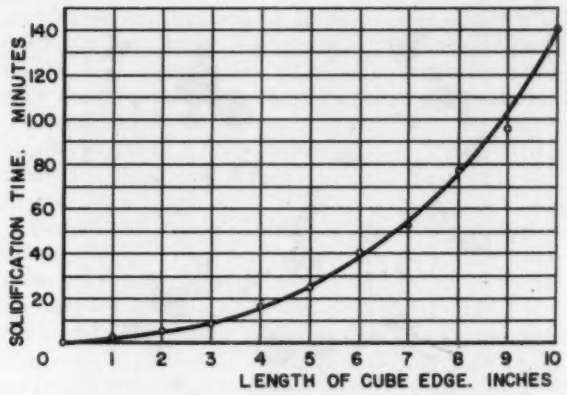



**Fig. 2 — Size of cube castings effect on length of time required for solidification. Pouring temperature — 2600 F.**

The results of tensile tests on these irons are presented in Table 1. In all conditions of heat treatment, lower strength and ductility were found in the heavier castings. For example, in the partially annealed samples with a relatively constant hardness of about 174 Bhn, the elongation decreased from 12 per cent in the 5 in. cube to 6 per cent in the 10 in. cube. The concurrent change in tensile strength was from 73,400 to 67,500 psi. The yield strength was unaffected. Similar behavior was observed in the as-cast and annealed specimens.

The decrease in both strength and elongation in

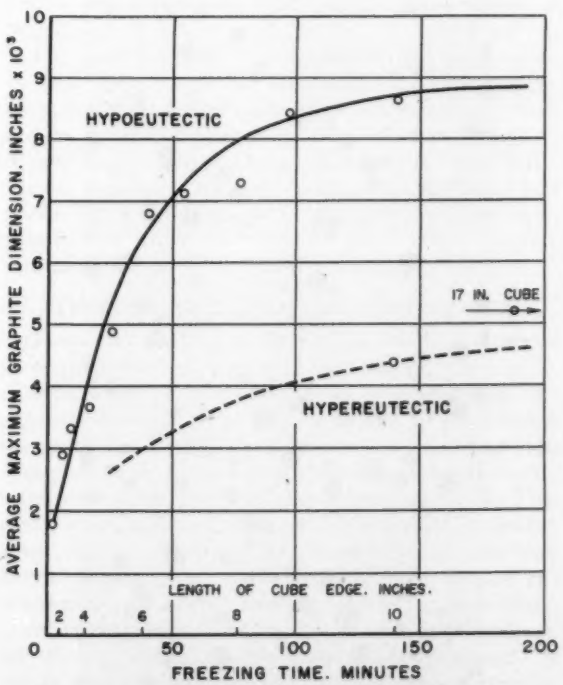



**Fig. 3 — Solidification time effect on size of graphite particles in hypoeutectic and hypereutectic ductile irons. The largest particle in a 0.024 in. field, selected at random, was measured. Each value is the average of approximately 20 such measurements.**

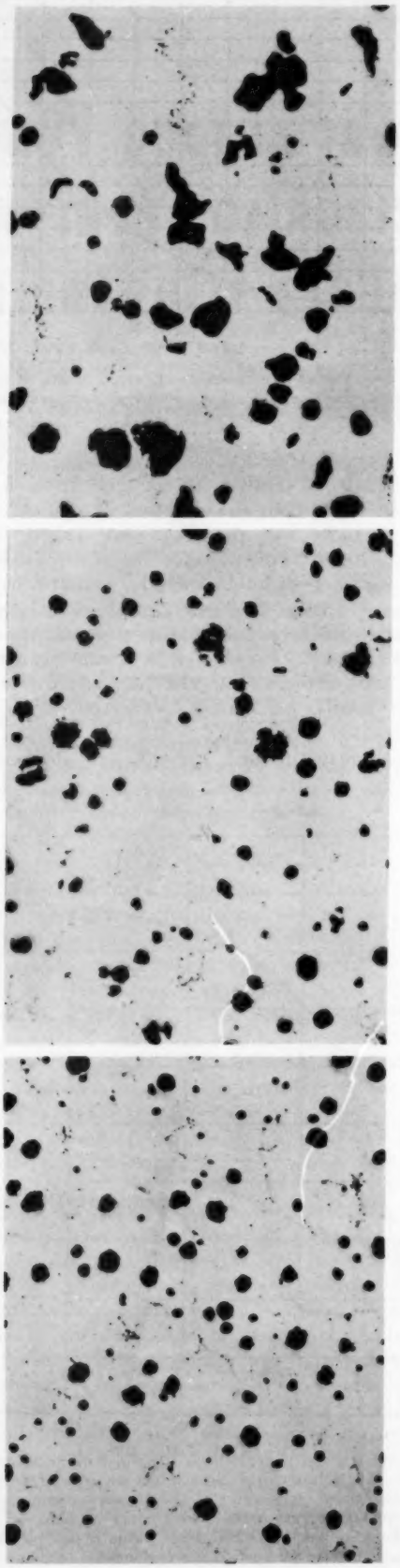


Fig. 4 — Freezing rate effect on graphite structure in hypoeutectic ductile iron. *Top to bottom* — cube size 7 in., freezing time — 53 min; cube size — 5 in., freezing time — 25 min; cube size — 3 in., freezing time — 8 min. Unetched. 40 ×.

the larger castings is attributed, at least in part, to a deterioration of the shape of the graphite particles which accompanies the increased size. Figure 4 illustrates this change. Although the shape of these particles is irregular, the inner crystalline structure is still that of a spheroid, since a radial structure can be seen under polarized light.

### *Composition Effect*

**Carbon Equivalent.** The deterioration of the spheroidal shape of graphite in hypoeutectic irons due to slow cooling has been described. It has been found however, that as soon as the carbon equivalent of the liquid alloy after magnesium treatment and inoculation exceeds the value of 4.3, the spheroidal shape of the graphite is preserved even if the casting cools slowly. For example, melt 6 in Table 2 with a carbon equivalent of 4.23, contained large irregular spheroids while melt 10 with a carbon equivalent of 4.34 contained small well formed spheroids. Figure 5 shows the graphite structure of such a hypereutectic iron.

Not only are the graphite particles in hypereutectic ductile iron more spheroidal, they also are smaller. The slowest solidification rate in this investigation produced graphite particles about 0.0042 in. in diameter in hypereutectic irons, while hypoeutectic irons contained graphite particles twice this size. These data are shown in Fig. 3. The size of the spheroids

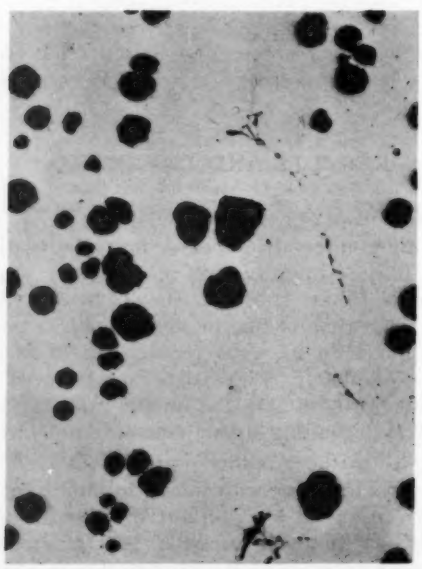


Fig. 5 — Graphite structure in a 7 in. hypereutectic ductile iron cube (melt 15). Solidification time 150 min. Unetched. 40 ×.

in a commercially made 17 in. cube of hypereutectic ductile iron[1] was also determined and found to be 0.0052 in.

In the laboratory melts a small decrease in carbon content occurred at the time of the magnesium treatment, as shown in melts 20, 21 and 22 in Table 2. The decrease varied from 0.04 to 0.12 per cent. The carbon contents prior to and after the magnesium treatment were determined using chilled pin samples. The cause of the carbon loss is not known. The magnitude of the change may be dependent upon the conditions under which the iron is treated. This effect should be taken into account in practice.

The final carbon content in the body of a heavy hypereutectic iron casting will be approximately eutectic. The hypereutectic carbon is rejected from the liquid iron during freezing and floats to the cope surface of the casting.[2] The thickness of the segregated carbon layer will depend upon the carbon equivalent of the iron and the height and cooling rate of the casting.

**Silicon and Manganese.** In order to separate the effects of carbon and silicon, both hypoeutectic and hypereutectic melts were prepared with various silicon contents. The silicon range for the hypoeutectic alloys was 0.6-1.58 per cent and for the hypereutectic alloys, 0.77-2.77 per cent. No significant effect of the silicon content on the graphite structure has been found except through its influence on the carbon equivalent, as shown by melts 13, 14, 15 and 21 in Table 2. Although the magnesium contents of these melts are quite low, no flake graphite was observed except in melt 21. An evaluation of the size and shape of the spheroids was possible in melt 21, even though some flake graphite was present.

Two melts were made with low manganese contents to determine the influence of a change in this direction (low manganese ingot molds had exhibited coarse irregular graphite structures). A decrease in manganese from 0.59 to 0.07 per cent did not produce coarse graphite in hypereutectic melts (melts 18 and 19 in Table 2).

**Magnesium and Cerium.** In the past it has been common practice to employ high magnesium contents in irons intended for heavy castings. The satisfactory structures achieved in melts 15, 16, 18 and 27 with low magnesium contents indicate that magnesium contents higher than normally employed in light ductile iron castings are not necessary. In addition, melts 7, 8, 9 and 10 indicate that high magnesium contents in the particular base iron employed are detrimental resulting in mixtures of flake and spheroidal graphite, as shown in Fig. 6. Another base iron yielded a satisfactory structure with 0.11 per cent magnesium in melt 24. However, spheroidal graphite structures were always obtained at 0.04-0.05 per cent magnesium, indicating that this level is optimum irrespective of casting size and base iron.

The ability of cerium to counteract the flake forming effect of titanium and other detrimental elements is well established.[3] An addition of cerium to melt 26, however, gave an unexpected result. This iron in the absence of cerium contained spheroidal graphite and fine, intercellular flake graphite. The addition of 0.012 per cent cerium was not beneficial. The amount of fine flake graphite was decreased, but at the same time a large amount of chunk graphite was produced. The microstructure of this iron is shown in Fig. 7. Smaller additions may be satisfactory, but it seems clear that large amounts of cerium in heavy castings should be avoided.

**Ferrosilicon Inoculation.** Each of the experimental alloys was inoculated with ferrosilicon after the magnesium treatment except melts 20, 21 and 22, in which the magnesium treatment was performed with a ferrosilicon-base calcium bearing alloy which presumably contributed some inoculating effect. In melt 23 no inoculant was added, and an unusual importance of the ferrosilicon inoculation was established. Referring to Table 2 it is evident by comparison with melts 15, 16 and 18 that melt 23 was hypereutectic and would have contained spheroidal graphite if post inoculated, but in the absence of inoculation large irregular graphite particles formed.

An attempt was made in melt 6 to obtain well formed spheroids in a hypoeutectic iron by means of greater inoculation. Twice the amount of ferrosilicon normally added was employed, but the resultant structure showed no improvement.

Since the carbon equivalent of the melt is critical only at the moment of pouring, the ferrosilicon inoculation can easily be used to control the final carbon equivalent. Thus, in melt 10 a hypoeutectic base iron was converted to a hypereutectic composition by means of the final ferrosilicon addition, and the resultant graphite structure was satisfactory. This procedure probably would lend itself well to commercial practice.

The control of the carbon content of a hypereutectic base iron might be difficult because of kishing. However, the carbon content of a hypoeutectic melt should be stable. The conversion of such iron to a hypereutectic composition by means of the final ferrosilicon inoculant after the magnesium addition will result in a good graphite structure, as previously shown, and might be a satisfactory method of composition control.

## HYPEREUTECTIC IRON PROPERTIES

Comparison of the properties of the hypereutectic irons, presented in Table 3, with the data on hypoeutectic irons in Table 1, shows that although the graphite structure is greatly improved in hypereutectic melts, the strength and elongation in the section sizes poured in this investigation may not be markedly improved, particularly in the as-cast condition. This study was concerned mainly with graphite structure, since it was felt that the first step in achieving better properties in heavy castings must be the improvement of the graphite structure. Apparently, other factors, outside the scope of this investigation, are also important. Intercellular inclusions, such as can be seen in Fig. 5, because of their semi-continuous distribution, probably exert a large influence upon the properties of the material.

In heavier castings, such as the hypereutectic 17 in. cube, the improved graphite structure obtained in

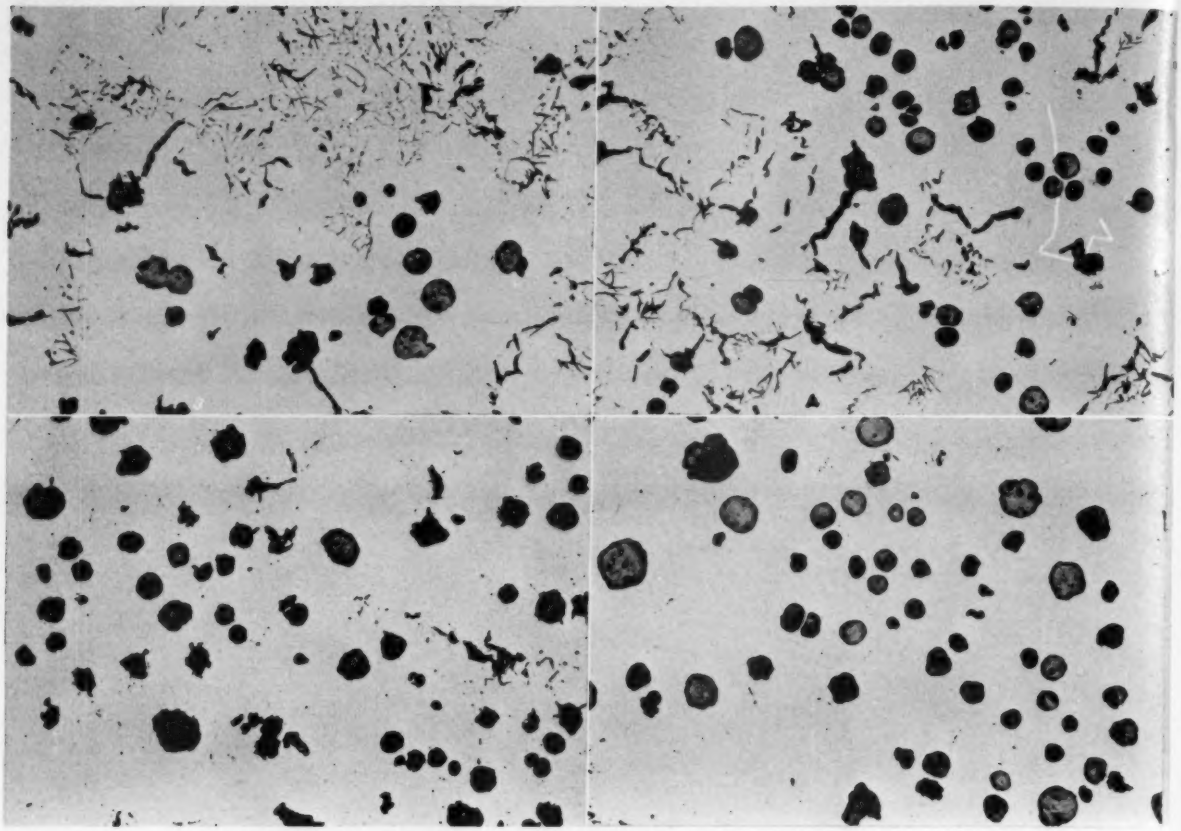


Fig. 6 — Magnesium content effect on graphite structure in slowly solidified ductile iron. *Top row — left to right* — melt 7, 0.14 per cent Mg; melt 8, 0.08 per cent Mg. *Bottom row — left to right* — melt 9, 0.059 per cent Mg; melt 10, 0.044 per cent Mg. Unetched. 40 ×.

hypereutectic irons gives improved properties. These data are also shown in Table 3.

A peculiarity of the microstructure of hypereutectic ductile iron leads to somewhat different properties in the vertical and horizontal directions. Figure 8 shows a macrograph of one of these irons. During solidification large dendrites form, and on the underside of these dendrites graphite spheroids collect, apparently by flotation. This microsegregation of the graphite is different from the segregation of hypereutectic graphite to the cope surface. Only graphite precipitating during eutectic solidification is involved. This type of solidification produces a layered structure. Figure 5 shows a layer of spheroids at a high magnification. The area around the spheroids becomes ferritic as customary in ductile irons. In Fig. 8 the dendritic areas are ferrite, which appears dark because oblique illumination was used. Since flotation is involved the layers tend to occur in horizontal planes, resulting in slightly lower ultimate strengths and elongation in the vertical direction. The data on melt 15 in Table 3 illustrate this behavior.

## CONCLUSIONS

Hypereutectic ductile irons, i.e., irons with carbon equivalents equal to 4.3 or more, contain well formed graphite spheroids in large castings which freeze slowly. Hypoeutectic irons, however, contain large irregularly shaped graphite particles in large castings.

Inoculation with calcium bearing ferrosilicon is essential to the formation of spheroidal graphite structures in heavy castings. Uninoculated irons, even though hypereutectic, contain large irregular graphite particles.

In large castings, hypereutectic ductile iron provides greater strength and ductility than does hypoeutectic iron. Stratification of graphite spheroids in hypereutectic irons causes lower ductility in a vertical direction than in bars oriented horizontally.

Magnesium contents in excess of 0.05 per cent are not necessary, and sometimes are detrimental in heavy castings.

## ACKNOWLEDGMENT

The authors are grateful to The International Nickel Co., Inc., for permission to publish the results of this investigation. Ductile iron is a patented material produced under license from The International Nickel Co., Inc. The material presented in this paper is the subject of a patent application.

## REFERENCES

1. Unpublished Research, W. R. McCrackin Cooper-Bessemer Corp.

TABLE 1 — CASTING SIZE EFFECT ON TENSILE PROPERTIES OF HYPOEUTECTIC DUCTILE IRON

| Melt No. | Casting Size, in. | Mechanical Properties: As-Cast Tensile Str., psi | As-Cast 0.2% Yield, psi | As-Cast Elong., % | As-Cast Bhn | Partially Ferritic Tensile Str., psi | Partially Ferritic 0.2% Yield, psi | Partially Ferritic Elong., % | Partially Ferritic Bhn | Annealed — 100% Ferritic Tensile Str., psi | Annealed — 100% Ferritic 0.2% Yield, psi | Annealed — 100% Ferritic Elong., % | Composition, % TC | Si | Mn | Ni | P | Mg |
|---|---|---|---|---|---|---|---|---|---|---|---|---|---|---|---|---|---|---|
| 1 | 3x3x3 | 116,900 | 73,200 | 4 | 243 | | | | | | | | 3.06 | 1.97 | 0.42 | 1.26 | 0.042 | 0.075 |
| | | 104,200 | 69,500 | 3 | 258 | | | | | | | | | | | | | |
| | 4x4x4 | 97,000 | 66,100 | 1½ | 244 | | | | | | | | | | | | | |
| | | 90,400 | 66,100 | 2 | 245 | | | | | | | | | | | | | |
| 2 | 5x5x5 | 95,300 | 59,000 | 4½ | 200 | 73,400 | 51,300 | 12 | 170 | | | | 3.34 | 2.14 | 0.40 | 0.83 | 0.040 | 0.050 |
| | | 94,800 | 59,300 | 4½ | 207 | | | | | | | | | | | | | |
| | 6x6x6 | 90,300 | 55,500 | 5 | 206 | 72,200 | 50,900 | 9½ | 167 | | | | | | | | | |
| | | 82,800 | 54,300 | 3 | 207 | | | | | | | | | | | | | |
| | 7x7x7 | 80,100 | 52,600 | 4 | 196 | 73,800 | 51,300 | 8½ | 171 | 65,300 | 46,100 | 11 | | | | | | |
| | | | | | | 72,300 | 51,200 | 7 | | | | | | | | | | |
| | | | | | | 71,900 | 50,300 | 8½ | | | | | | | | | | |
| 3 | 8x8x8 | 76,700 | 50,200 | 3½ | 195 | 69,600 | 50,000 | 7 | 169 | 64,700 | 45,900 | 8½ | 3.32 | 2.12 | 0.39 | 0.84 | 0.037 | 0.036 |
| | | 73,900 | 49,500 | 3½ | 192 | 71,900 | 50,800 | 7 | 174 | | | | | | | | | |
| | | | | | | 70,800 | 50,400 | 7 | | | | | | | | | | |
| | | | | | | 72,500 | 50,400 | 8 | 175 | | | | | | | | | |
| 4 | 9x9x9 | 68,600 | 53,000 | 2 | 195 | 68,900 | 50,800 | 6 | 166 | 63,800 | 46,400 | 6½ | 3.29 | 2.22 | 0.41 | 0.84 | 0.036 | 0.067 |
| | | 71,400 | 52,000 | 2½ | 190 | 69,100 | 50,500 | 6 | 178 | | | | | | | | | |
| | | | | | | 71,100 | 51,900 | 7 | 174 | | | | | | | | | |
| 5 | 10x10x10 | 72,400 | 51,500 | 3 | 187 | 66,000 | 48,000 | 6½ | 170 | 61,700 | 45,700 | 6½ | 3.33 | 2.18 | 0.40 | 0.89 | 0.036 | 0.072 |
| | | 68,400 | 49,400 | 2½ | 188 | 69,900 | 51,300 | 5½ | 179 | | | | | | | | | |
| | | | | | | 66,800 | 50,100 | 5 | 170 | | | | | | | | | |

TABLE 2 — CHEMICAL COMPOSITIONS, CASTING SIZES AND GRAPHITE STRUCTURE OF VARIOUS SLOWLY COOLED DUCTILE IRON CASTINGS

| Melt No. | Casting | Base TC | Carbon Equivalent of Melt at Time of Pouring $CE = TC + \frac{Si}{3.2}$ | TC Chill | TC Casting | Si | Mn | Mg | Ce | Graphite | Remarks |
|---|---|---|---|---|---|---|---|---|---|---|---|
| 6 | 7 in. cube in insulated and preheated mold | | 4.23 | 3.75 | 3.77 | 1.53 | 0.34 | 0.084 | | D | 0.5% FeSi Inoculation |
| 7 | 7 in. cube in insulated and preheated mold | | >4.3 | | 3.65 | 1.70 | 0.57 | 0.14 | | B | |
| 8 | 7 in. cube in insulated and preheated mold | | 4.33 | 3.65 | | 2.16 | 0.59 | 0.08 | | B | 0.5% FeSi Inoculation |
| 9 | 7 in. cube in insulated and preheated mold | | >4.3 | | 3.50 | 1.59 | 0.27 | 0.059 | | A | |
| 10 | 7 in. cube in insulated and preheated mold | | 4.34 | 3.96 | 3.73 | 1.23 | 0.30 | 0.044 | | A | |
| 11 | 7 in. cube in insulated and preheated mold | | 4.62 | 4.22 | 3.87 | 1.28 | 0.24 | <0.01 | | C | |
| 12 | 7 in. cube in insulated and preheated mold | | 4.65 | 4.25 | 3.77 | 1.28 | 0.30 | 0.055 | | A | |

*(continued on next page)*

*(Table 2 — continued)*

**TABLE 2 — CHEMICAL COMPOSITIONS, CASTING SIZES AND GRAPHITE STRUCTURE OF VARIOUS SLOWLY COOLED DUCTILE IRON CASTINGS**

| Melt. No. | Casting | Base TC | Carbon Equivalent of Melt at Time of Pouring $CE = TC + \frac{Si}{3.2}$ | TC Chill | TC Casting | Si | Mn | Mg | Ce | Graphite | Remarks |
|---|---|---|---|---|---|---|---|---|---|---|---|
| 13 | 7 in. cube in insulated and preheated mold | | 4.08 | 3.59 | 3.53 | 1.58 | 0.39 | 0.025 | | D | |
| 14 | 7 in. cube in insulated and preheated mold | | 4.20 | 4.01 | 3.89 | 0.60 | 0.37 | 0.016 | | D | |
| 15 | 7 in. cube in insulated and preheated mold | | 4.57 | 4.33 | 3.88 | 0.77 | 0.40 | 0.027 | | A | |
| 16 | 7 in. cube in insulated and preheated mold | | 4.92 | 4.12 | 3.47 | 2.57 | 0.35 | 0.035 | | A | |
| 17 | 7 in. cube in insulated and preheated mold | | 4.80 | 4.0 | 3.49 | 2.55 | 0.39 | 0.015 | | B | |
| 18 | 7 in. cube in insulated and preheated mold | | >4.3 | | 3.58 | 2.06 | 0.07 | 0.033 | | A | |
| 19 | 7 in. cube in insulated and preheated mold | | 4.52 | 3.90 | 3.60 | 1.99 | 0.55 | | | A | |
| 20 | 7 in. cube in insulated and preheated mold | 4.03 | 4.64 | 3.91 | 3.56 | 2.32 | 0.10 | 0.014 | | B | No post inoculation, treated with 1.2% FeSi base Mg alloy[1] |
| 21 | 7 in. cube in insulated and preheated mold | 3.93 | 4.72 | 3.89 | 3.40 | 2.77 | | 0.010 | | B | No post inoculation, treated with 1.7% FeSi base Mg alloy[2] |
| 22 | 7 in. cube in insulated and preheated mold | 3.83 | 4.19 | 3.75 | 3.60 | 1.42 | | 0.035 | | D | No post inoculation, treated with 2.0% FeSi base Mg alloy[1] |
| 23 | 7 in. cube in insulated and preheated mold | | 4.51 | 3.71 | 3.58 | 2.55 | (0.4) | 0.027 | | D | No post inoculation |
| 24 | 8 in. cube in green sand | | >4.3 | | | (2.0) | (0.5) | 0.11 | | A | |
| 25 | 8 in. cube in green sand | | >4.3 | | | (2.0) | (0.5) | 0.066 | | B | |
| 26 | 8 in. cube in green sand | | >4.3 | | | (2.0) | (0.5) | 0.10 | 0.012 | E | |
| 27 | 10 in. cube in green sand | | >4.3 | | 3.44 | 2.32 | 0.42 | 0.019 | | A | |
| 28 | 17 in. cube | | 4.5 | 3.63 | 3.26 | 2.61 | 0.3 | 0.057 | | A | |

LEGEND: Values in parenthesis are calculated.

A — Spheroidal graphite (Fig. 5)
B — Spheroidal and flake graphite (Fig. 6)
C — Flake graphite
D — Irregular graphite (Fig. 1)
E — Chunk graphite (Fig. 8)

| | Mg, % | Si, % | Ca, % | Rare Earths, % |
|---|---|---|---|---|
| 1. | 9 | 45 | 2 | — |
| 2. | 9 | 45 | 2 | 1.1 |

TABLE 3 — MECHANICAL PROPERTIES OF HYPEREUTECTIC DUCTILE IRON IN HEAVY SECTIONS

| Melt No. | Composition, % TC | Si | Mn | Mg | Melt C.E. | Mechanical Properties: As-Cast Tensile Str., psi | 0.2% Yield, psi | Elong., % | Bhn | Annealed Tensile Str., psi | 0.2% Yield, psi | Elong., % | Bhn | Normalized Tensile Str., psi | 0.2% Yield, psi | Elong., % | Casting |
|---|---|---|---|---|---|---|---|---|---|---|---|---|---|---|---|---|---|
| 10 | 3.73 | 1.23 | 0.30 | 0.044 | 4.34 | 64,300 | 49,700 | 1 | | | | | | | | | 7 in. cube in preheated insulated mold |
| | | | | | | 66,100 | | 2 | | | | | | | | | |
| 15 | 3.88 | 0.77 | 0.40 | 0.027 | 4.57 | 71,000 | 39,000 | 3 | | | | | | 81,500 | 40,900 | 5½ | 7 in. cube in preheated insulated mold |
| | | | | | | 64,400 | 39,000 | 3 | | | | | | | | | |
| | | | | | | 68,300 | 39,300 | 2 | | | | | | | | | |
| | | | | | | 62,700 | 39,400 | 2(V) | | | | | | | | | |
| | | | | | | 63,000 | 39,500 | 2(V) | | | | | | | | | |
| 16 | 3.47 | 2.57 | 0.35 | 0.035 | 4.92 | 63,700 | 51,100 | 2(V) | 175 | 64,400 | 51,500 | 4(V) | 174 | | | | 7 in. cube in preheated insulated mold |
| | | | | | | 62,500 | 51,800 | 2(V) | 174 | 66,700 | 51,400 | 4½(V) | 179 | | | | |
| | | | | | | 64,800 | 51,400 | 3(V) | | 68,100 | 51,800 | 5(V) | | | | | |
| | | | | | | | | | | 67,900 | 51,400 | 5(V) | | | | | |
| 28 | 3.44 | 2.32 | 0.42 | 0.019 | >4.3 | 60,900 | 44,600 | 4½ | | 59,100 | 40,900 | 11 | | | | | 10 in. cube in green sand |
| | | | | | | 64,900 | 44,800 | 6½ | | 58,900 | 40,900 | 10 | | | | | |
| | | | | | | 64,700 | 45,000 | 7 | | 59,400 | 40,400 | 11½ | | | | | |
| 29* | 3.26 | 2.61 | 0.30 | 0.057 | 4.5 | | | | | 66,000 | 51,900 | 14½(C) | 165 | | | | 17 in. cube in dry sand |
| | | | | | | | | | | 67,100 | 62,000 | 5½(C) | | | | | |
| | | | | | | | | | | 65,700 | 60,600 | 5(C) | | | | | |
| | | | | | | | | | | 66,000 | 58,700 | 3(S) | | | | | |
| | | | | | | | | | | 64,800 | 60,400 | 4(S) | | | | | |
| | | | | | | | | | | 65,800 | 51,600 | 13(S) | | | | | |
| | | | | | | | | | | 65,000 | 51,000 | 11(S) | | | | | |

*Ref(1).

LEGEND: (V) — Vertical bar.
(C) — From center.
(S) — 4 in. from surface.

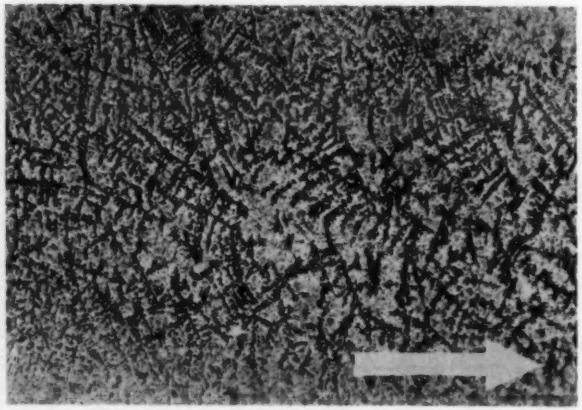


Fig. 8 — Dendrite like pattern of the ferrite (appearing black) in a 7 in. hypereutectic ductile iron cube, melt 16. Solidification time, 150 min. The arrow points toward the top of the casting. 4 per cent nital etch. Natural size. Oblique illumination.


2. A. H. Rauch, J. B. Peck and G. F. Thomas, "Carbon Flotation in Ductile Iron," AFS TRANSACTIONS, *vol. 67*, p. 263 (1959).

3. H. Morrogh, "Influence of Some Residual Elements and Their Neutralization in Magnesium-Treated Nodular Cast Iron," AFS TRANSACTIONS, *vol. 60*, p. 439 (1952).

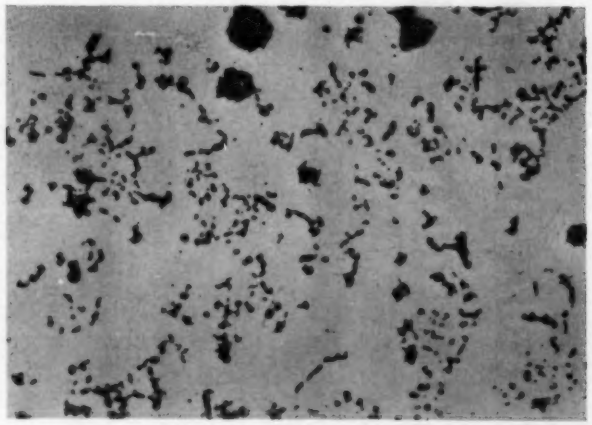



Fig. 7 — Graphite structure in an 8 in. hypereutectic ductile iron cube containing cerium, melt 26. Light etch with 2 per cent nital. 100 ×.

# VACUUM INDUCTION MELTING ALLOY QUALITY REQUIREMENTS

by T. F. Kaveny


## ABSTRACT

In spite of the large quantity and often the good quality of the information published about vacuum melting there has been no public consideration given to raw materials specifications. Generally the vacuum melter has simply taken the stand of purchasing only those materials low in their content of gases and other residual elements. This survey was undertaken to analyze the requirements of alloys to be used in vacuum induction melting.

The problem has been approached by the means which presented themselves. The literature on vacuum melting was surveyed and discussions were held with many active in this field. A thermodynamic approach was taken to the problems of gases, but this was tempered by using as a basis of such work some practical operating information. Finally some work carried out at the laboratories of the author's company was used to illustrate some points regarding deoxidation of the bath.


## INTRODUCTION

The quality of raw materials used in vacuum induction melting can be judged in large measure by their gas contents, i.e., hydrogen, oxygen and nitrogen. These gases can determine the ultimate service ability of a finished product, and their evolution on melting can affect the length of the melting process. Inasmuch as such high physical properties are expected of vacuum melted alloys, the presence of tramp elements such as sulfur and phosphorus is even more harmful in these than in conventionally melted alloys. Although undesirable elements such as lead and bismuth can be removed from a melt by volatilization, the process is not generally relied on for such purification. The presence of all these impurities in raw materials is of concern to the vacuum melter.

Vacuum melting has developed in some measure an alloying system of its own, and elements such as silicon and manganese which are beneficial in air melting may be totally undesirable in vacuum melted alloys and hence place additional restrictions on raw materials.

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## OXYGEN

In essence the element controlling deoxidation is that which at a given temperature is in equilibrium with less oxygen than any other element in the alloy system. Deoxidation with elements resulting in solid or liquid reaction products are insensitive to pressure, and behave in vacuum exactly as they do in air. Carbon, however, which on reaction with oxygen results in a gaseous product, becomes a more effective deoxidizer under vacuum. The principle of vacuum melting therefore is to control the carbon content and the pressure such that carbon lowers the oxygen content below that level at which it would be in equilibrium with the other elements in the system.

There is no evidence to lead us to believe that metal oxides will, of themselves, reduce, although some oxygen may be carried from the system by volatilization of certain oxides. Excluding, therefore, the use of hydrogen or elimination by flotation of oxides, the removal of oxygen in commercial vacuum systems is accomplished largely by the carbon-oxygen reaction

$$\underline{C} + \underline{O} = \overline{CO}$$

is characterized at 2912 F by the equilibrium

$$(\%\,\underline{C})\,(\%\,\underline{O}) = P_{(CO)}/433$$

where $P_{(CO)}$ is the partial pressure of CO above the metal.

Under commercial vacuum conditions, carbon is much less a deoxidizer than would be predicted by use in the above equation of the gage pressure recorded in the tank. A large part of this difference is, of course, accounted for by the change in activity of carbon and oxygen as alloy is added to the melt. In addition, ferrostatic pressure of metal, and the difficulty of overcoming surface tension before a CO



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bubble can form and the reaction proceed, both place serious restrictions on the deoxidizing potential of carbon.[1] Both these factors are viewed here as increasing the effective tank pressure. The whole matter of vacuum decarburization is further complicated by reoxidation of the bath by breakdown of the furnace lining.

### *Carbon Deoxidation*

Prediction of the extent to which carbon can deoxidize a bath relies therefore on the availability of practical operating data to couple with thermodynamic principle. Aksoy[2] has reported some actual carbon-oxygen products measured at various working pressures. These are shown in Table 1.

In addition to the measured carbon-oxygen products for iron baths which have been given, Aksoy has also determined similar information for this reaction in other materials. These are presented in Table 2.

TABLE 1

| Gage Pressure, atm | Products in Liquid Iron, %C, %O | | |
|---|---|---|---|
| | Theoretical Product | Measured Product[1] | Measured Product[2] |
| $10^{-7}$ | $2.3 \times 10^{-10}$ | $1.0 \times 10^{-6}$ | $1.0 \times 10^{-5}$ |
| $10^{-4}$ | $2.3 \times 10^{-7}$ | $2.5 \times 10^{-5}$ | $6.0 \times 10^{-5}$ |

1. Small furnaces — up to 50 lb capacity.
2. Large furnaces — up to 2000 lb capacity.

TABLE 2

| Metal | Products Measured Averages, %C, %O | Crucible |
|---|---|---|
| Nickel | $2.5 \times 10^{-6}$ | Magnesia |
| Cobalt | $5.0 \times 10^{-6}$ | Magnesia |

A comparison of Tables 1 and 2 shows that cobalt and nickel base melts should be more readily deoxidized with carbon than are iron melts. Thus, whereas iron base alloys are considered here, the results shown will be on the conservative side for cobalt and nickel base alloys.

Hilty, Rassbach and Crafts[3] have shown the relationships existing between chromium and carbon in an oxygen saturated bath at approximately one atmosphere pressure. Their curve at 2912 F (1600 C) is presented in Fig. 1 (this and the family of curves drawn at different temperatures comprise one of the basic relations in stainless melting as normally practiced today). Figure 1 shows the minimum level at which carbon will control deoxidation where excess oxygen is present in the bath.

The Hilty relationship is redrawn in Fig. 2 for a low pressure. In this case, operating information is chosen from Aksoy's data to reflect a system operating at a pressure of the order of $5\mu$. The effective pressure as discussed earlier will actually be higher.* The low pressure line is about parallel to the original, but is located at a much lower carbon level. This curve would indicate, for example, that a chromium of 20 per cent carbon would control deoxidation so long as it is at a level of 0.011 or higher.

### *Final Oxygen Content*

Figure 2 would indicate that thermodynamically the oxygen content of a vacuum melted alloy is largely dependent on its final chemistry and not on the oxygen content of the raw materials. This has been substantiated by experimental work performed at the

*For purpose of these relations a C • O product of $3 \times 10^{-5}$ is used to represent a system in a large furnace under a tank pressure of $5\mu$. Under these conditions, the "effective pressure" is calculated to $10{,}000\mu$.

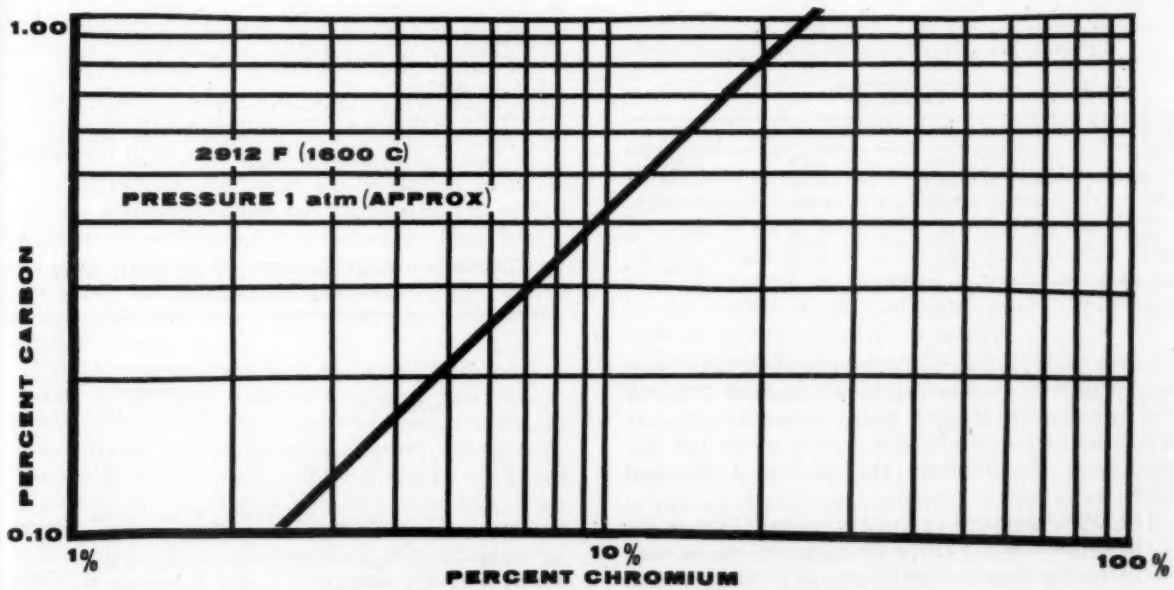



Fig. 1 — Equilibrium between chromium and carbon in an oxygen saturated bath at 2900 F and one atm pressure.

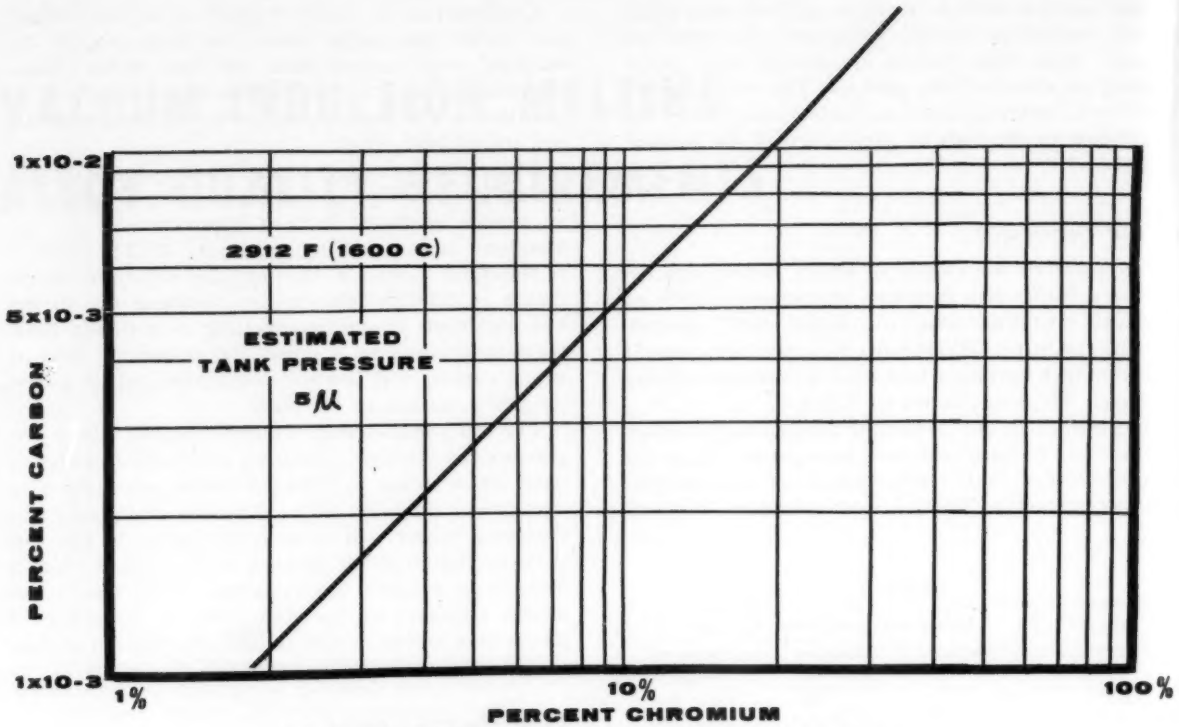



Fig. 2 — Equilibrium between chromium and carbon in an oxygen saturated bath at 2900 F and reduced pressure (approximately 5 $\mu$).

laboratories of the author's company. In this work, six 30 lb experimental heats of a 17 per cent chromium alloy were produced under vacuum conditions. Two heats were produced from each of three different sources of chromium metal, 0.55 per cent O electrolytic chromium, 0.08 per cent O lump, chrome metal and 0.05 per cent O vacuum processed chrome pellets. The results of these tests are shown in Table 3.

TABLE 3

| Heat No. | Chrome Source | Final C, % | Final O, % |
|---|---|---|---|
| 165 | 0.05% O pellets | 0.003 | 0.024 |
| 166 | 0.05% O pellets | 0.002 | 0.025 |
| 167 | 0.55% electrolytic | 0.006 | 0.020 |
| 168 | 0.08% O lump | 0.005 | 0.015 |
| 169 | 0.55% O electrolytic | 0.006 | 0.017 |
| 170 | 0.08% O lump | 0.008 | 0.012 |

These tests show fairly conclusively that final oxygen is mostly a function of final carbon. Thus, so long as sufficient carbon is added to remove whatever oxygen is charged, the oxygen content of the raw materials will not influence the quality of the final product.

The principles used in the mathematical treatment of chromium versus carbon deoxidation can be used in comparing deoxidation by any two elements. Unfortunately data are not always available to permit such a treatment. In particular such data are often lacking for the strong deoxidizers such as aluminum and titanium, which are so important in many vacuum melted alloys.

Given sufficiently low pressure and sufficiently high carbon, carbon deoxidation will predominate over deoxidation by solid deoxidizers. In the case of chromium such predominance is large and when treating the stronger deoxidizers such as aluminum the influence of carbon is not nearly so great. Visharev[4] has developed figures comparing the reduction of chromium oxides and aluminates by carbon. These results are shown in Tables 4 and 5. Although similar results for titanium are not shown, it can be assumed that these would be intermediate between chromium and aluminum, probably much closer to the latter.

TABLE 4 — DECOMPOSITION OF $Cr_2O_3$ AND $Cr_2O_3$ FeO INCLUSIONS

| Heat | In Billet, % C | Mn | Si | Cr | C in v.t. Metal, % | Inclusions, % Before v.t. | After v.t. | Means Inclusions Decomposing, % |
|---|---|---|---|---|---|---|---|---|
| 212 | 0.02 | 0.12 | tr | 6.81 | 0.007 | 0.394 | 0.331 | 19.4 |
| 213 | 0.02 | 0.12 | tr | 6.81 | 0.004 | 0.394 | 0.304 | |
| 221 | 0.09 | 0.16 | tr | 5.33 | 0.06 | 0.088 | 0.019 | 76.2 |
| 222 | 0.09 | 0.16 | tr | 5.33 | 0.08 | 0.088 | 0.023 | |
| 237 | 0.42 | 0.07 | 0.02 | 6.95 | 0.39 | 0.0103 | 0.0024 | 78.4 |
| 238 | 0.42 | 0.07 | 0.02 | 6.95 | 0.42 | 0.0103 | 0.0031 | |
| 247 | 1.08 | 0.13 | 0.01 | 5.06 | 1.01 | 0.0074 | 0.0012 | 81.6 |
| 249 | 1.08 | 0.13 | 0.01 | 5.06 | 1.05 | 0.0074 | 0.0015 | |

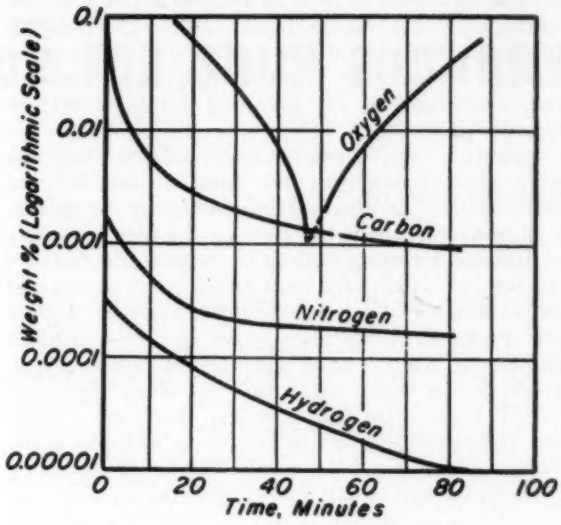



Fig. 3 — Degassing during vacuum induction melting of iron in a magnesia crucible.

The results show strongly that oxides of aluminum are relatively stable, and that the presence in the system of this element, or other strong deoxidizers, will tend to limit the effectiveness of carbon deoxidation. This is particularly true for baths of low carbon content. They also show conclusively that the oxygen content of late additions of such elements should be as low as possible. This same conclusion can be reached for almost any element added late to a low carbon bath.

TABLE 5 — DECOMPOSITION OF ALUMINATE INCLUSIONS

| Heat | In Billet, % C | Mn | Si | Al | C in v.t. Metal, % | Inclusions, % Before v.t. | After v.t. | Means Inclusions Decomposing, % |
|---|---|---|---|---|---|---|---|---|
| 411 | 0.03 | 0.09 | 0.04 | 0.24 | 0.012 | 0.310 | 0.282 | 13.6 |
| 412 | 0.03 | 0.09 | 0.04 | 0.24 | 0.009 | 0.310 | 0.257 | |
| 421 | 0.10 | 0.18 | 0.01 | 0.17 | 0.08 | 0.082 | 0.064 | 26.8 |
| 422 | 0.10 | 0.18 | 0.01 | 0.17 | 0.09 | 0.082 | 0.056 | |
| 432 | 0.44 | 0.12 | 0.07 | 0.22 | 0.42 | 0.0157 | 0.0091 | 35.0 |
| 433 | 0.44 | 0.12 | 0.07 | 0.22 | 0.43 | 0.0157 | 0.0103 | |
| 448 | 0.98 | 0.07 | 0.03 | 0.14 | 0.95 | 0.0098 | 0.0062 | 34.6 |
| 449 | 0.98 | 0.07 | 0.03 | 0.14 | 0.97 | 0.0098 | 0.0066 | |

Visharev's results were developed by remelting steel of known oxide content in a vacuum at 20 mm (20,000 $\mu$) pressure. After holding this small melt for 5 to 15 min the metal was permitted to freeze and the number of inclusions recounted. The results obviously were developed at pressures much higher than Americans would normally operate, but the melt was small and would be expected to give results closer to equilibrium than would our commercial vacuum melting units.

The chemistry of the bath will control the oxygen content of the bath (there is a limit to this control, a limit imposed by the transfer of oxygen from the furnace lining). A number of investigators have proved that the longer a heat is left in a molten condition in a vacuum, the higher will be its oxygen. This oxygen creep is attributed to the breakdown of the basic lining. This phenomenon, the influence of which is shown in Fig. 3,[5] suggests the importance of tapping the heat as soon as possible after the carbon boil has substantially subsided.

### *Oxygen Conclusions*

The evidence indicates that the final oxygen content of a vacuum melted heat is largely a function of final chemistry (primarily carbon), the vacuum and pick-up from the lining. The oxygen content of nonoxidizing or weakly deoxidizing raw materials exerts no discernable influence on the oxygen analysis of the final product – providing sufficient excess carbon is charged to remove the oxygen in the charge. Of course, where it is necessary to make additions of these elements late in the heat when the carbon content of the bath is low and no more can be added conveniently, their oxygen content must be a matter of concern to the quality-minded melter.

Strong deoxidizers such as aluminum, and to a smaller extent titanium, will suppress deoxidation by carbon. They should, therefore, be added late in the heat (after deoxidation by carbon is completed), and they should be as free from oxygen as is possible.

## NITROGEN

The removal of nitrogen from a vacuum induction furnace is more difficult than is the removal of oxygen. Unless the nitrogen content of a bath is extremely high compared to its equilibrium concentration, the tendency to form bubbles of nitrogren in the bath or on the refractory wall of the furnace is low. The escape of nitrogen is, therefore, a matter of diffusion to the surface of the bath, and since the nitrogen atom is large its diffusivity is small.

The solubility of nitrogen obeys Sievert's Law for diatomic gases. This is expressed:

$$\underline{N} = K\sqrt{P_{n_2}}$$

Langenberg and Day[6] have developed a relationship to predict the nitrogen content of alloy steel at one atmosphere pressure of nitrogren. This may be worked out for any iron base alloy the result substituted for $K$ in the above equation. For instance the predicted solubility of nitrogen at 2900 F in a 20 per cent chromium alloy at one atmosphere pressure of nitro-

gen is 0.27 per cent. Substituting this value in the Sievert equation the predicted solubility of nitrogen in this alloy at 7 micron pressure ($10^{-5}$ atm) is 0.0008 per cent. This type of relationship will indicate the ultimate that can be expected of nitrogen removal, but unfortunately the predicted result is seldom reached in practice.

Induction stirring and extended holding times will assist nitrogen removal, but these are not 100 per cent effective. Another aid in eliminating nitrogen is a carbon-oxygen boil. By creating bubbles or centers of diffusion for nitrogen deep in the bath, the distance through which nitrogen must diffuse is minimized and its removal enhanced. Where other analytical factors are the same, it has been noted that low nitrogen contents go hand in hand with the low carbon-oxygen products – or with alloys which have been vigorously boiled. Where otherwise practical, therefore, a vigorous carbon-oxygen boil is helpful to nitrogen elimination.

In nickel and cobalt base alloys, the solubility of nitrogen is less than it is in iron – consequently, the elimination of nitrogen from such materials is less a problem than in iron base materials. Since nitrogen removal from vacuum furnaces is so inefficient, all raw materials should be scrutinized carefully for their nitrogen content. This is particularly true of nitride formers such as titanium. Any nitrogen present in the bath when the titanium addition is made, or any present in the addition agent itself will be tied up as TiN or TiCN, and will probably result in excessive inclusions in the final product.

## HYDROGEN

The mechanism by which hydrogen is removed from solution is thought to be largely akin to that by which nitrogen is eliminated, i.e., diffusion. Actually there may be some elements of a hydrogen-oxygen reaction involved. The extent or conditions under which this latter reaction occurs is unknown, but it is probably of academic interest only. The hydrogen atom is small and diffuses readily, so much so that hydrogen is seldom a problem in vacuum melting. To some extent hydrogen can be tied up by such elements as titanium, so care should be taken in choosing such alloys for hydrogen content – otherwise hydrogen is not critical in alloy selection.

## MISCELLANEOUS ELEMENTS

Among the group of harmful miscellaneous elements are sulfur, phosphorus, lead, tin, tin, etc. Such elements are no more harmful to vacuum melted materials than to similar air melted alloys, but the high physical properties and consistency of performance expected of the former makes these tramps particularly obnoxious.

**Phosphorus.** There is no known way to eliminate phosphorus from a vacuum melt, and its introduction in any raw material is to be scrupulously avoided.

**Sulfur.** It has been suggested that sulfur can be removed from a vacuum melt by the reaction:

$$\underline{2S} \rightarrow S_2 \text{ (gas)}$$

$$S + 2\underline{O} \rightarrow SO_2 \text{ (gas)}$$

$$H2 + \underline{S} \rightarrow H_2S \text{ (gas)}$$

The pressures necessary to force the first two reactions are much lower than those used today, and so little desulfurization can be expected of these. The last reaction, though unaffected by pressure, requires so large a volume of hydrogen gas as to be somewhat impractical. Vacuum melted steels can be effectively desulfurized with cerium additions. Here, however, the melt will be contaminated with cerium and the product of its reaction with sulfur.

Fluoridas[7] has demonstrated that sulfur is removed from high silicon vacuum melts, presumably as a gaseous silicon-sulfide. The silicon needed, however, is frequently in excess of the normal requirements of steel, and the time required for appreciable desulfurization may impose operating (time) penalties on the melter. Although sulfur can be removed by some mechanisms, so far a simple, practical approach has not been found.Until then, careful attention must be given to the sulfur content of the raw materials.

## GAS EVOLUTION EFFECT

A problem of concern to the vacuum melter is the effect of oxygen in the charge of production time. If pumping capacity is low the length of cycle may be extended by the rate of evolution, and even with adequate pumping too high a rate of CO generation may necessitate slowing the cycle by pressure buildup to prevent metal splash. The relationship between time and oxygen is not entirely proportional, and a 0.10 per cent oxygen chromium alloy may lead to no longer a melting cycle than a heat made with 0.05 per cent oxygen. Certainly, however, commercial tests of a 0.50 per cent oxygen electrolytic chromium have demonstrated that so high an oxygen will appreciably lengthen the refining time. The maximum oxygen that can safely be handled will however vary with different melting facilities.

With proper information it is believe it will be possible to determine the length of cycle, and consequently the economic penalty involved when using chromium metals of various oxygen content.

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# STANDARD TIMES DEVELOPMENT FOR JOB SHOP FOUNDRY

by David L. Raphael


## ABSTRACT

The work measurement problems of a job lot foundry are stated. The characteristics of an adequate work measurement tool to solve these problems are discussed. The use of a predetermined motion time system is indicated, and the nature of such a system explained. The procedure for accurate estimation of foundry operation performance times using such a system is outlined. Foundry operations are divided into standard elements. Typical motion sequences and performance times for the performance of these elements are derived. These motion patterns have the characteristic that they adjust sensitively to changes in critical variables, such as weight, which affect the performance time of a standard element. A typical example is discussed, and methods of further simplification of the estimating process using equations and graphs are shown. How such rapid and efficient estimates can be applied in the foundry is then emphasized. These occur in the areas of production standards, production control and scheduling and order cost estimating. The more extensive use of industrial engineering techniques in the foundry is recommended.


## INTRODUCTION

There is one area of the general foundry industry which presents an aggravated problem with respect to cost control and efficient scheduling of production. To a great extent, many foundries are involved in job lot production. This means production runs of a great variety of different items. The particular items will vary in size, shape and order quantity. The result is a heterogeneous chaos of production runs of varying lengths with castings of a wide range of complexity. Such conditions can rapidly lead to dangerous confusion and a high degree of inefficiency in handling and completing orders.

The inherent characteristics of job lot production represent a manufacturing problem of a high order of difficulty. The intense variability possible in pattern complexity, size and in order quantity makes management control far from simple. Often, despite a high volume of orders, the foundry operation may run at a marginal or no return level if not at a loss. Despite the valiant efforts of management to operate the foundry as a profit making enterprise, the result is failure, and most often a failure due to lack of sensitive and effective control. The control mechanisms usually available lack the accuracy and speed of response necessary to adjust adequately to the rapidly changing production problems with which management is faced.

The root of the problem lies principally, though not completely, with the manual labor involved in processing the orders through the foundry. Because of the type, size and quantity variations of orders, it is often extremely difficult if not impossible to estimate the time required to produce a given order or one unit of it. Conventional techniques are often inadequate for the job, due to the time required to use them and the difficulty with which they can be used as predictive tools prior to actual production. The making of a mold cannot be timed with a stop watch if the mold is not actually being made.

The fundamental data starts with the time required for processing orders, including production planning and control, wage payment and bidding and costing activities. Thus, an adequate work measuring technique takes into account these items, and becomes the starting point for successful foundry operation.

## THE WORK MEASUREMENT PROBLEM

One of the major areas of work and, therefore, cost in a job shop foundry is in the making of molds. Performance times for these operations must be estimated if this activity is to be controlled successfully. The large number of different molds which must be made compounds the problem. An adequate, flexible and rapid work measurement tool is needed to supply the required estimates. Only in this way can efficient foundry operation be achieved. It is this problem which is covered herein.

Proper work measurement can be achieved for mold making operations and all their ancillary sub-operations. Techniques and formulas can be developed through the use of some of the newer work measurement tools. This paper proposes to demonstrate how this can be done.

## STANDARD ELEMENTS

The first step to be taken is to reduce the molding operation into a set of standard elements. By standard, a set of elements is meant which includes all the possible steps to be taken in making a mold. Since this set must include all possibilities, it will contain elements which will not necessarily occur in every molding operation. In general, a subset of this set of standard elements can be selected. This subset will cover each required step in the moldmaking operation being analyzed. Such sets of standard elements are shown in Tables 1 and 2. Any person experienced with foundry operations can quickly select an appro-

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priate subset of elements which will describe completely the steps of any given operation.

It is these standard elements which are to be measured. If a reasonably accurate estimate of performance time for any given element can be easily and rapidly produced, it will be possible to generate an efficient and usable estimate of a whole molding operation performance time. Examination of the standard elements will indicate a basic characteristic. In order that the element apply to all similar molding operations, the descriptions are necessarily vague. For example, element J (assemble empty cope to drag) says nothing about cope size, cope weight or the distance of the movement.

All of these factors, as well as others, may affect the time an operator takes in performing this element. Any usable system of estimating element performance times must account for these factors.

The essence of this report is the demonstration of an adequate method of providing sufficiently accurate element time estimates. The problem has often been solved in the past through the use of standard data. These are generally performance times derived from historical time study data collected over a considerable period of time. Such data are usually inadequate for two reasons – (1) the past experience of the foundry does not include all possible variations of molding elements for which future time estimates may be required, and (2) where historical data does not exist, the gap is filled by methods of interpolation and extrapolation. Thus, standard data may often provide incomplete or inaccurate coverage in solving the element time estimating problem.

There is one final recourse. One can set up for purposes of measurement, all standard element variations, which apply to a given foundry and actually take stop watch time studies of each one. This would certainly provide adequate standard data. A moment's thought will quickly indicate the weakness in such a procedure. The number of time studies required for only a reasonably exhaustive program would be extremely large. It would involve several men for a considerable period of time. The expense in both labor cost and lost time would be too great for the usual small or medium size multipurpose foundry.

### *Predetermined Motion-Time Systems*

What then is the solution which can be both efficient and economical? The answer lies in the use of techniques which can synthetically provide reasonably accurate time estimated. They must be capable of simulating sufficiently, the myriad element variations and provide an exhaustive set of element performance time estimates for these variations. Finally, they must do all this at a cost in time and money that is within reason.

Techniques having the foregoing characteristics have been emerging since World War II from the areas of work measurement and industrial engineering. They are generally grouped under the title "Predetermined Motion-Time Systems." Though the basic validity of such systems has been and is a matter of controversy, years of successful application have resulted in at least a partial acceptance of these systems. In the area of estimating and approximating work times even the most rabid critic will admit their empirical validity.

The particular system being used herein is Methods-Time Measurement (MTM). This choice is in no way a recommendation for this system over many of the others. It was militated by the author's personal familiarity with it over a number of years, both in research and application. It, therefore, was a natural

**TABLE 1 — STANDARD ELEMENTS**

| Bench Molding — Loose Pattern, Split |
|---|
| A. Place drag on molding board |
| B. Place pattern (drag section) in drag on molding board |
| C. Riddle sand on pattern |
| D. Shovel sand in drag |
| E. Ram sand |
| F. Pull strike-off bar over drag |
| G. Place bottom board on drag and turn over drag (full) |
| H. Place pattern (cope section) on pattern (drag section) |
| J. Assemble cope (empty) to drag |
| K. Riddle sand on pattern |
| L. Shovel sand in cope |
| M. Ram sand |
| N. Pull strike-off bar over cope |
| P. Make sprue with sprue cutter |
| Q. Draw cope (full) off drag |
| R. Turn over cope (full) and aside |
| S. Clean and slick drag parting surface |
| T. Swab drag mold cavity edge with water |
| U. Tap drawspike in drag pattern and tap drawspike to sides to loosen pattern |
| V. Draw pattern from drag mold |
| W. Gate drag mold |
| Y. Repair and finish drag mold cavity |
| Z. Swab core mold cavity edge with water |
| Aa. Tap drawspike in cope pattern and tap drawspike to sides to loosen pattern |
| Ba. Draw pattern from cope mold |
| Ca. Repair and finish cope mold cavity |
| Da. Set cores in mold cavity |
| Ea. Turn over cope (full) |
| Fa. Close flask (assemble cope on drag) |
| Ga. Weight flask for pouring |

**TABLE 2 — STANDARD ELEMENTS**

| Machine Molding — Matchplate, Jolt-Squeeze Machine |
|---|
| A1. Place cope and drag (assembled, empty) on machine |
| B1. Remove cope from drag and aside |
| C1. Place matchplate on drag |
| D1. Assemble cope to matchplate and drag |
| E1. Turn over flask |
| F1. Riddle sand on matchplate in drag |
| G1. Shovel sand in drag |
| H1. Ram sand (peening) in drag with shovel |
| J1. Smooth off surplus sand with hands |
| K1. Place mold board on drag and turn over flask (drag full) |
| L1. Riddle sand on matchplate in cope |
| M1. Shovel sand in cope |
| N1. Ram sand (peening) in cope with shovel |
| P1. Smooth off surplus sand with hands |
| Q1. Make sprue with sprue cutter |
| R1. Set squeeze board on cope |
| S1. Machine jolts and squeezes flask |
| T1. Draw cope (full) off drag |
| U1. Turn over cope (on end) and aside |
| V1. Draw matchplate (vibrator) off drag |
| W1. Repair and finish drag mold cavity |
| Y1. Set cores in drag mold cavity |
| Z1. Turn over cope (full) |
| A2. Close flask (assemble cope on drag) |
| B2. Remove flask from mold |

choice. Any other system could easily do the job required. Sources for the major accepted predetermined motion time systems are given in the reference.

### *Predetermined Motion Time Systems*

What does a predetermined motion time system do? First, it divides and classifies the motions used by a human being in performing a work task. It further subdivides each classification into subclassifications, determined by the variables which affect the performance of the motion. Second, it assigns to each particular motion variation a normal or average performance time. For example, in MTM a particular motion can be symbolized – M12C8.

Here a basic motion element has been recognized and classified in terms of three variables affecting its performance. These are:

M, Move-an arm movement involving the transportation of an object from one location to another.

12, the distance of the movement in in.

C, indicating a movement to an exact location, i.e., requiring a high degree of control to successfully complete it.

8, lb – the weight of the object being transported.

Application tables for MTM give an estimated normal performance time for this motion of 0.000161 hr.

In addition, each system provides a variety of rules of application so that such things as simultaneous motions can be handled adequately. With these rules, motion classifications and performance time data, most manual industrial operations can be successfully analyzed, and the normal or average performance time of the operation estimated. The procedure is simply:

1) List the sequence of required motions.
2) Determine the performance time for each motion from the tables of times.
3) Sum the individual times to obtain the performance time for the total operation.

This total time is, of course, a normal or average time. Allowances for fatigue and unavoidable delays must be added. These will vary from shop to shop and from type to type of operation. For this reason, they are treated separately and independently.

### *Advantages of MTM*

The advantage of a technique such as this is immediately apparent. No stop watch timing is required. The performance times for individual motions already exist, having been derived through research and analysis of a large number of similar motions performed by operators in industry. All that is required is an accurate determination of the motion sequence required to perform an operation or standard element. This can be done by actual observation if the element is being performed as by simulation where it is not being performed.

For a specific standard molding element, the system automatically compensates for variations in critical variables such as weight. For example, let us assume that element *J* (assemble empty cope to drag) involves a move of 12 in. to an exact location, but with copes with weights of 10, 15 and 20 lb each. The symbols and times for these motions would be:

M12C10 ..... 0.000169 hr
M12C15 ..... 0.000178 hr
M12C20 ..... 0.000185 hr

Further, if the cope weight were the same (say 15 lb) but the distances of the motion were 8, 12 and 16 in., respectively, the symbols and times for the motions would be:

M8C15 ..... 0.000138 hr
M12C15 ..... 0.000178 hr
M16C15 ..... 0.000218 hr

As can be seen by examining these motion times, estimated performance time increases as the weight transported increases and also as the distance of the motion increases. This is in accord with what would be expected. In the same manner, the system sensitively adjusts for all the motions involved to the variations one might expect for different cases of a given standard element.

## MEASURING THE STANDARD ELEMENTS

Let us now consider the manner in which measurement of the performance times of the standard molding operation elements, such as those given in Tables 1 and 2 can be made. Time measurement in this case resolves into two aspects – those elements which are essentially manual and those which are machine controlled with respect to time.

### *Manual Elements*

The manual elements constitute the vast majority of the standard elements to be measured. Table 1 lists the standard elements for bench molding operations. These are completely manual in nature. Table 2 lists the standard elements for typical machine molding operations. Of all the standard elements only element S1, machine jolts and/or squeezes flask, and possibly V1, draw matchplate off drag (vibrator), are machine controlled in determining their performance time. All the rest are manual elements. Thus, the major problem is that of analyzing and measuring these manual elements.

The first step is to regroup these standard manual elements into classes having similar motion characteristics. They can be grouped into the general categories:

1) Assembly of flask and pattern parts.
2) Disassembly (drawing) of flask and pattern parts.
3) Turning of flask and pattern parts.
4) Handling sand.
5) Finishing mold cavity.
6) Setting cores.

Table 3 lists the elements which belong in each classification.

It is beyond the scope of this paper to present typical motion patterns for every standard element. The goal is to demonstrate the procedure one should follow for efficient work measurement in the foundry. Toward this end, analysis shall be made of an ex-

**TABLE 3 — STANDARD ELEMENTS — MOTION PATTERN CLASSIFICATION**

1. Assembly of flask and pattern parts
   a. Bench molding
      A, B, H, J, Fa, Ga
   b. Machine molding
      A1, C1, D1, R1, A2
2. Disassembly (drawing) of flask and pattern parts
   a. Bench molding
      Q, V, Ba
   b. Machine molding
      B1, T1, B2
3. Turn flask or pattern parts
   a. Bench molding
      G, R, Ea
   b. Machine molding
      E1, K1, U1, Z1
4. Handling sand
   a. Bench molding
      C, D, E, F, L, N
   b. Machine molding
      F1, G1, H1, J1, L1, M1, N1, P1
5. Finishing mold cavity
   a. Bench molding
      P, S, T, W, Y, Z, Ca
   b. Machine molding
      Q1, W1
6. Setting cores
   a. Bench molding
      Da
   b. Machine molding
      Y1

ample from one of the categories listed, to demonstrate this method. From category (1) elements J (Table 1) and D1 (Table 2) which can be described "Assemble cope (empty) to drag or drag (empty) to cope" shall be examined.

The elements from category (1) involve the assembly of flask and pattern parts. These are affected by several variables. The most important and critical of these are the weight of the part being assembled, the guide pins which may be involved in terms of their length, fit or clearance with the guide pin holes of the mating part, and possibly their shape and the size of the part being assembled. These variables operate in a basic placing motion pattern which consists of moves of the part by the hand and arm, positions of the part, and final seating moves in that order.

In the case of a cope, drag, flask or pattern plate, both hands usually control the object. In the case of a loose pattern, only one hand may be necessary. The influence of these variables is reflected in the performance time of these motions in this way:

1) *Weight* – by the application of weight factors to move motions, including constant weight times where necessary, the performance time will increase as the weight of the part increases.
2) *Fit of guide pin in holes* – by changing the class of fit of the position motion in terms of the clearance involved, the performance time of the position motion will increase as the clearance decreases.
3) *Size of the part* – performance time of the element will increase by the fact that the move motions will be performed together, separately or replaced by more time consuming body motions as the size of the part increases. In addition, position motions are considered as easy or difficult to handle (thus increasing their time) depending on the bulk of the object.

***Motion Patterns Range***

The specific range of typical motion patterns for elements *J* and *D1* appear in Table. 4. These are presented showing the motions performed by the right hand and the left hand. It should be remembered that both hands are grasping the part. The assembly of an empty cope to the drag or empty drag to the cope can be performed by two general motion patterns, as indicated by Table 4. The particular pattern which will be used is determined by the variable levels which occur for a particular assembly. In addition, the performance time for the particular pattern which applies varies with the distances and part weights involved.

**TABLE 4 — MOTION PATTERNS**

| Assemble Empty Cope to Drag or Empty Drag to Cope | | | | |
|---|---|---|---|---|
| Pattern | A | | B | |
| Flask Length | L _ 15″ | | L _ 15″ | |
| Method | M_C_<br>P1SE<br><br>M_A_ | M_C_<br><br>M2C_<br>P1SE<br>M_A_ | M_B_<br>M2C_<br>P1SE<br><br>M_A_ | M_B_<br><br>M_C_<br>P1SE<br>M_A_ |

Some discussion is necessary for a proper understanding of Table 4. To begin, cope or drag weight, the length of the guide pins and the size of the flask not only affect the times of the specific MTM motion element, but also the method of assembly. Table 4 basically shows the method changes which occur. The additional effect on motion times determined by weight and distance are not shown specifically. The effect is indicated merely by a dash, into which the appropriate weight or distance must be inserted by the MTM motion symbol.

The time for a given motion will be determined from the tables of times by the weight and distance values so inserted. Thus, a particular method in Table 4 does not determine a single standard element performance time, but is merely a framework which can produce a range of times for the conditions under which that method is used. This range of times is produced by the possible weights and distances permissible for that method.

Perhaps a brief description of one of these motion patterns will make their meaning clear. In pattern B in Table 4 it is seen that both hands holding an empty cope of a given weight, move it to the vicinity of the drag. In symbols

M-B- M-B-

The left hand then moves the left hand side of the cope preparatory to positioning the guide pin holes of the cope to the guide pins (M2C-). The cope is then positioned to the guide pins (PISE). The right hand then performs the same task with the right hand side of the cope (M2C-, PISE). Then, together, the

hands lower the cope down to the drag over the guide pins.

| | |
|---|---|
| M-A- | M-A- |

When the weight of the cope and the distances of the movements are known, estimated performance times are automatically determined. When these are summed, the total is an estimated normal time for the element. For example, suppose the cope is 16 in. x 20 in. and weighs 14 lb. Let the distance of movement to the drag be 18 in. and the guide pin distance moved be 2 in. Then, using the MTM application data:

| | | |
|---|---|---|
| M18B7 | 0.000218 | M18B7 |
| M2C7 | 0.000055 | |
| P1SE | 0.000056 | |
| | 0.000055 | M2C7 |
| | 0.000056 | P1SE |
| M2A7 | 0.000038 | M2A7 |
| | 0.000478 | |

The times given are in hours. Note that each hand has a load of 7 lb, since the 14 lb weight of the cope is shared equally by both.

The previous discussion indicates that, using a predetermined motion time system, a standard mold making element can be easily described and rapidly provide an estimated performance time. It can do so with a sensitive adjustment to the factors affecting the elements performance time. If substitution were made to a different cope weight and/or different movement distance, this would result in a different performance time adjusting for the change in variable levels.

## REDUCTION AND COMBINATION OF STANDARD ELEMENTS

By developing motion patterns for all standard elements, in the manner shown in Table 4, a total operation time estimate can easily be derived in a reasonable period for all manual portions of the operations. However, actually selecting a sequence of element motion patterns, inserting the appropriate variable levels in the motion descriptions and finally determining performance times from application tables may still seem too tedious or time consuming. If certain factors can be held constant, a further simplification of the time estimating problem can be made.

For example, if the work place can be standardized to some degree, the general motion pattern for an element can be reduced to a simple equation or graph from which performance times can be estimated even more easily.

Let us suppose an element whose motion pattern is given by Table 4. By standardizing the work place, the distance the empty cope will be moved to the drag can be fixed to average 20 in. In addition, let all flasks have guide pins of a length that the seating moves of the cope on the drag will always be one in. The motion patterns will be:

| | | |
|---|---|---|
| Pattern A | M2OC-<br>P1SE | M2OB- |
| | | MIC-<br>P1SE |
| | MIA- | MIA- |
| Pattern B | M2OB-<br>M2C-<br>P1SE | M2OB- |
| | | M2C-<br>P1SE |
| | M1A- | M1A- |

Note that all pertinent variable levels are inserted in the motion symbols except that of weight. Thus, there are two motion patterns, one for flasks of smaller size and one for flasks of a larger size. Both of these will provide times which will vary with cope weight.

It is possible to derive mathematical equations from these motion patterns. Since the only variable affecting time is now weight, the performance times of the individual motions can be combined into a constant portion and a variable portion related to the weight of the cope. For the two motion patterns these turn out to be:

$$\text{Pattern A} \ldots . T = 0.000392 + 0.00000154W$$
$$\text{Pattern B} \ldots . T = 0.000423 + 0.00000171W$$

where $W$ is the weight of the cope, and the time of the pattern ($T$) is given in hours. We have now reduced the problem of estimating the standard element performance time to selecting a linear equation and substituting a cope weight in it. Similar expressions can be derived for all elements, thus further simplifying the time estimating procedure. If desired, the equations can be expressed in graphical form. In this case, times can be obtained directly from the graph. Even a further simplification may be accomplished. The individual element equations could be combined into a single operation time equation making the estimating procedure a one-step affair.

### *Machine Controlled Elements*

The previous discussion has described the method of handling manual standard elements. The machine controlled elements cannot and need not be handled in this manner. In this case, a stopwatch can be used. This should not, however, cause any undue difficulty. The machine elements are few in number and are relatively stable with respect to time. Standard stopwatch times should be derivable without great difficulty.

With standard element performance time estimates now available from concise analytical expressions or tables of element times, it is possible to develop a general foundry time formula. The methods of producing such a formula are essentially straightforward and available from standard work measurement texts covering this subject. They will not be discussed here. Suffice it to say that such a step would now be possible. With such a formula the unit normal production time can rapidly be obtained whenever such information is required. The stage is now set for placing foundry production under control and into a profit making state.

It is important at this point to reiterate the point that any proved predetermined motion time system will be adequate to solve the problem presented here. What is required, however, is proper and thorough training in the technique chosen. Given a properly

trained person, the goal of good foundry work measurement can be rapidly accomplished. It may be helpful to the interested party to familiarize themselves with the available systems. Individual systems are explained at length by Baily,[1] Karger and Bayha[4] and Geppinger.[3] An actual demonstration of the application of a system to a foundry is discussed by Dobbins.[2] Short but complete resumes of four systems are given in Maynard.[5]

## LIMITATIONS OF THE METHOD

It is necessary to point out certain characteristics of molding operations which may limit the usefulness of the above described procedure. The application of predetermined motion time systems to foundry operations requires a great deal of care in analysis and a full understanding of the operation being studied.

Many elements may be process or machine controlled without such a condition being obvious. Sometimes even minor changes in the parts or materials involved will result in appreciable changes in the performance time of the operation or of a particular operation element.

When setting time standards for such an operation, it may be necessary to check the accuracy of an estimated time with a stopwatch. This is not due of course to the fact that the estimating method is incorrect, but because of qualities inherent in foundry operations themselves. These are such that certain elements of a molding operation may become almost completely process, thus invalidating the standard motion pattern used to describe and estimate performance times. An example of this situation would be the drawing or removal of a pattern plate from a cope or drag.

The pattern being withdrawn from the sand might be so complex or intricate in shape that too rapid removal might result in damage to the walls of the mold cavity formed in the sand. In this case, the operator must necessarily draw the pattern plate with extreme care. The time required to do this would be much larger than would be accounted for with standard element motion patterns. The element is essentially process controlled.

Though such conditions will not often exist, it is important to point out that they may. Only familiarity and experience with foundry operations can provide the knowledge necessary to anticipate and handle such occurrences. The obvious solution, when required, is to treat the element as a machine controlled one and time it with a stopwatch.

## APPLICATION

Though the estimated operation performance times are only approximations to the actual times the operations will take, they will be accurate enough to adequately solve the major problems in achieving efficient foundry operation and higher profits. The rapidity with which they will be available will make possible a degree of control not usually available in a multipurpose job lot foundry.

First, they will make possible the setting of equitable production standards. Since unit times will be available for all orders big and small, the setting of equitable work tasks will be possible. In addition, the performance of a worker on the job can be objectively evaluated in terms of performance and not in terms of subjective criteria. Wage payment can also be placed on a more accurate basis with the existence of a more nearly correct standard performance. What can be expected of a worker, even if he is involved in several different jobs during a work day can be estimated quantitatively and consistently.

This leads to the second possibility available, a workable scheduling and production control system. Quantitative job standards lay the basis for effective production scheduling. When jobs should enter the shop can be determined and scheduled with some hope of success. Estimated operation times yield estimates of processing time for a complete order. Order processing times then make possible realistic work schedules as the order mix wends its way through foundry day by day and week by week. Control systems tied to these schedules can then yield information on order status and meaningful delivery dates.

Finally, a means is now available through production time information to estimate direct labor cost. This makes possible price quotations which can assure a reasonable profit on a given order. The order which loses money because of inaccurate labor cost estimates can become a rarity. A method will be available, with speed and flexibility, to provide production time estimates, and hence, labor costs, synthetically prior to actual production.

## CONCLUSION

The foregoing discussion has demonstrated a method of measuring foundry molding operations. It should be remembered that this was merely a demonstration. It is hoped that it will indicate what can be done with work measurement in the job shop foundry. Certainly the investment in training personnel in a predetermined motion time system, and letting them apply it, could result in a level of quantitative control of production rarely found in the vast number of small and medium size foundries.

It should be also indicative that industrial engineering can contribute to a great degree to efficient foundry management. Industrial engineers have developed, over the years, many techniques such as predetermined motion time systems. It is hoped that they may be useful for the unique problems that prevail in the foundry industry. Certainly, it is the opinion of the author that more complete use of industrial engineering in the job shop foundry is both feasible and practical. Further, the continued existence of the small multi-purpose foundry will depend on its ability to utilize modern work measurement techniques in casting and the control of production.

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# HIGH ALUMINA ROOFS FOR ELECTRIC ARC FURNACES

by R. E. Gray


## ABSTRACT

**The primary consideration of this experimental work concerning furnace roof refractories was one of economics. The study of high-alumina roof practice was conducted utilizing acid and basic lined electric furnaces. Areas investigated were type of service, installation, longevity and cost. Roof cost per ton of steel was reduced by using 70 per cent alumina brick with replacable center sections.**


## INTRODUCTION

In the interest of prolonged roof life under varying conditions of operation (intermittent operations, oxygen lancing and increased operating temperatures), the author's company has used high alumina refractories in furnace roof construction since 1955. During the period 1955 to 1960 various methods of construction were tried, which led to the development of replaceable center sections, two-section skews, corrugated arch brick, drip ring and 80 per cent alumina center section ramming material. These developments have resulted in equalized life for all roof sections and substantial reduction in roof cost per ton of hot metal produced.

The company was interested in obtaining a refractory material for forming the cooling glands around the electrode ports to decrease downtime and associated costs. Several materials and arrangements were tried with little success, and finally it was decided to operate without cooling glands. No detrimental effects have been observed related to this practice during a period of one year.

Realizing that each furnace has its own characteristics, and is operated under different conditions, the results obtained may not be applicable to other shops.

## ROOF CONSTRUCTION

The roof ring is made of segments welded together and water outlets threaded, or welded, in place for circulation of water around the ring. Figure 1 shows layout of roof ring water cooling jacket. Using a jig to position it, the ring is placed on a form (Fig. 2) so that the electrode ports will receive electrodes when the roof is placed on the furnace. The diameter of the electrode port holes is 2 in. greater than the electrodes (A one in. diameter clearance was tried but was rejected because electrodes frequently grounded against the roof).

Skews are made to fit the roof ring on the slanted surface of the water cooled gland, and have the proper angle for the first annular ring of brick. It was observed upon removal of the roofs that the portion of the skew adjacent to the ring was not eroded to the extent of the other refractories of the roof. It was decided to utilize this less affected area of the skew by sectioning the skew longitudinally, so that the eroded portion could be removed and the remaining portion left for rebuilding of the roof. Experience has shown that this peripheral section of the skew could be reused three or four roofs. The author's company now orders the skews manufactured in two pieces.

The annular rings are formed using standard shapes of 9 in. arch, wedge and key brick of 70 per cent alumina. Each brick of the ring is dipped into a high temperature cement mortar prior to placing to provide a bond between each brick. Using high alumina brick, which are virtually nonexpanding, requires that the brick in the ring be tightly wedged together to prevent them from falling out during service. As each annular row of brick is placed in position, three or four key brick, equally spaced, are left protruding approximately 4 in. around the circumference of the ring. The protruding brick are then pounded into place using a wooden block to cushion the blow of the hammer.

The third annular ring of brick is positioned to protrude ¾-in. above the remaining portion of the roof to form a drip ring. This tends to direct the molten roof refractories into the furnace and avoid contact with the roof seal and upper sidewall of the furnace.

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The brick forming the electrode ports are 13½-in. arch brick. The back face of the brick (opposite the port face) is corrugated (Fig. 2) to prevent slippage. The contour of the roof form is designed to allow the port brick to protrude equal distance on the top and under side of the roof. The life of the port brick is increased by the number of heats required to burn away the portion of the brick protruding below the roof contour. It is not essential to have protrusion on the top side in cases where cooling glands are not used. However, this requires a change in roof form or brick size which has not been tried to date.

Two partial rows of brick are placed against the annular ring and between the electrode port brick allowing ample space for ramming of the remainder of the roof. Ramming material of 80 per cent alumina is inserted and ramming started with a tapered malleable iron butt. The face of the rammer butt contains welded beads to help prevent lamination. Material must be constantly added while ramming is being done; if not, lamination will occur and roof life will be greatly reduced. The last step in constructing the roof is to wash or brush the roof with a thin coat of high temperature cement.

## ROOF LIFE — ACID PRACTICE

The roofs on a P.T. furnace (2500 kva transformer) converted to top charge have attained as many as 477 heats, with one center section replacement after 226 heats. The average roof life with this practice is 440 heats, and the center section life is about one-half the total life or 220 heats. The average refractory cost of the center section is $154, or 33 per cent of initial roof cost. The heat size is 7850 lb; oxygen lance 4 min at 90 psi; tap at 3070 F

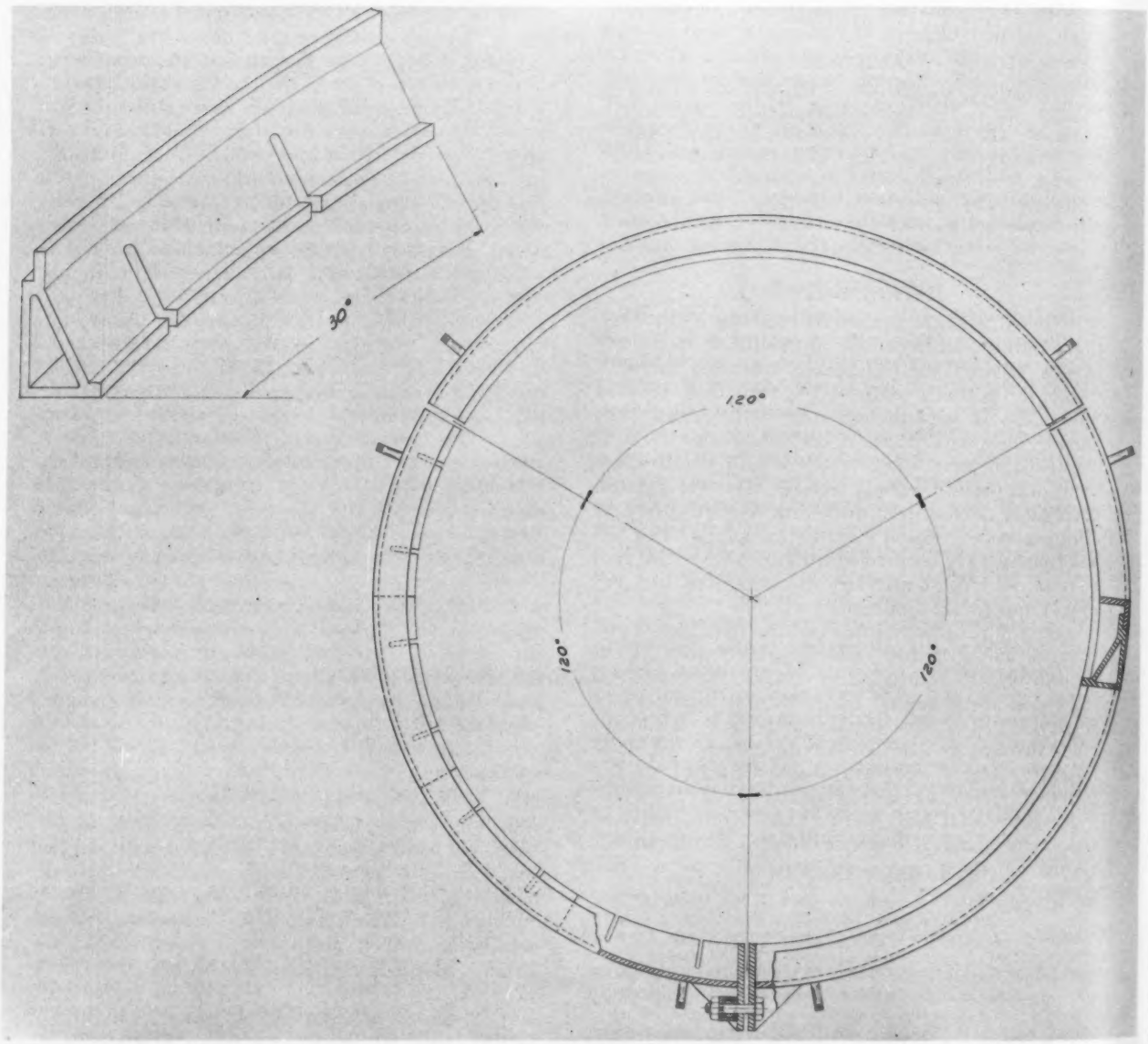



Fig. 1 — Water cooled roof ring.

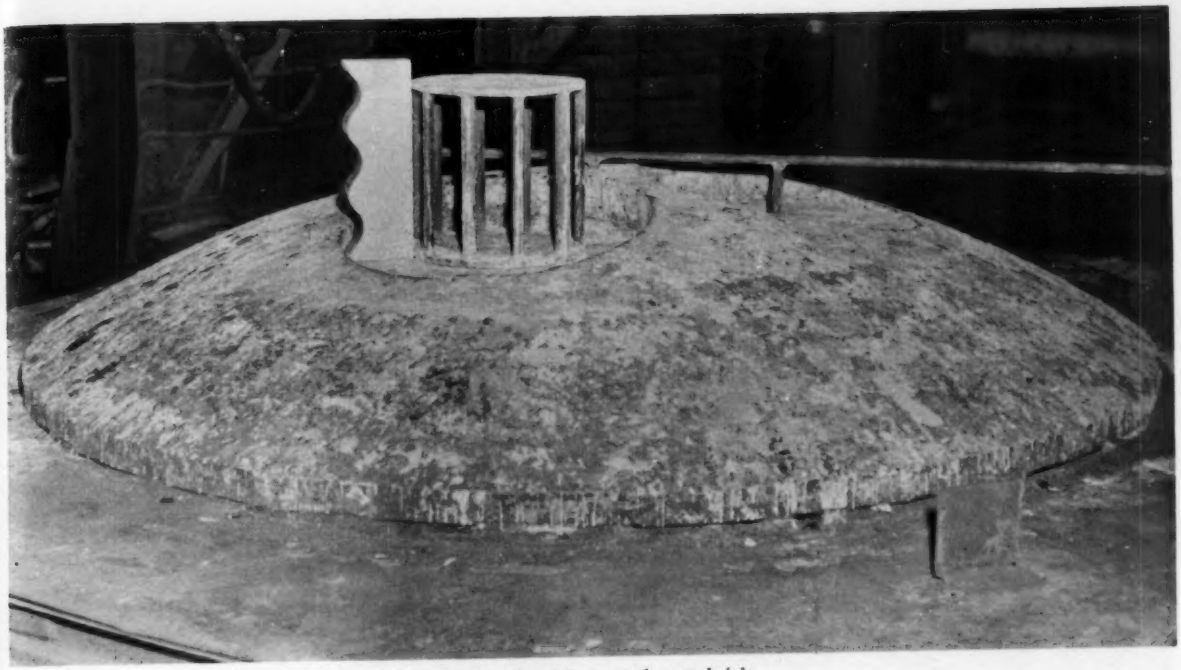


Fig. 2 — Roof form and port brick.

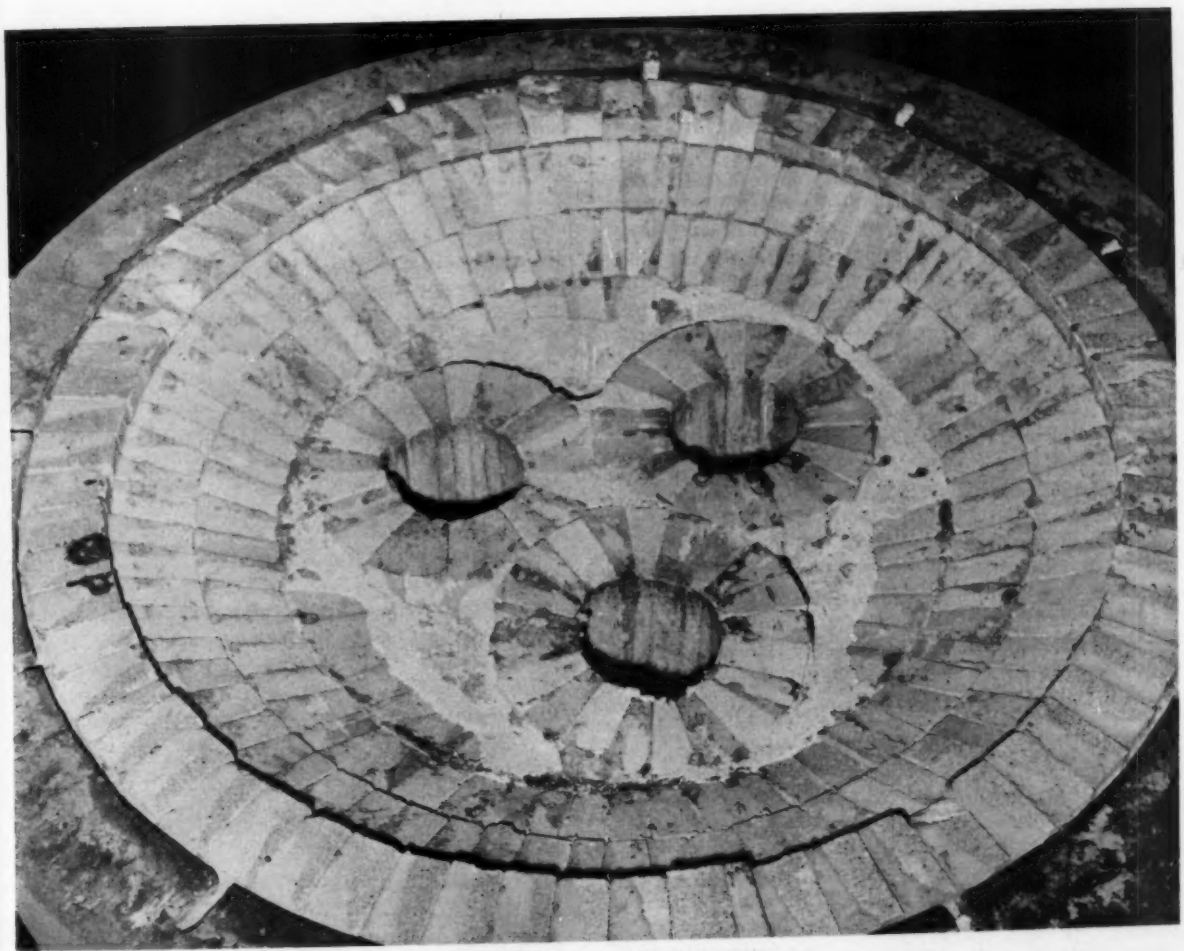


Fig. 3 — Partial cast roof center section.



Fig. 4 — Section of roof showing center section failure and directional erosion.

(1688 C) (immersion pyrometer); 70 min tap to tap; 60 per cent low alloy and 40 per cent Grade B metal.

In the ingot division on a 20 TT-furnace (7500 kva transformer), put in service in June, 1958, the roof life has varied from a maximum of 344 to a minimum of 188 heats, and has averaged 230 heats. The author's company has attempted to equalize the life of the various sections of this roof. The initial roof had one annular ring, two partial rows in the center section and the port hole brick of 70 per cent alumina with the remainder of the brick being 45 per cent alumina.

As the number of rows of 70 per cent alumina brick were increased, the cost increased about 5 per cent per annular ring. The fifth row increased cost by 5 per cent; fourth row, 9 per cent; third row, 14.77 per cent; second row, 21.34 per cent. Using all 70 per cent alumina brick, with the exception of the skew, the cost was 27.6 per cent more than the initial roof. The 70 per cent alumina brick roof has increased roof life by only 11.4 per cent, but has enabled the rebuilding of the center section at a cost of $384. This resulted in an additional 178 heats or a total roof life of 420 heats.

Roof cost per heat has been reduced by 13.1 per cent as a result of this development work.

## ROOF LIFE — BASIC PRACTICE

On a 4 M.T. furnace (1800 kva transformer), the roof has been made of high alumina refractories since initial startup in 1955. The furnace is used exclusively for melting stainless steel heats averaging 6000 lb. The average roof life using 70 per cent alumina brick with an 80 per cent alumina rammed center section is 46 heats.

The author's company has experimented using three roofs of high alumina castable material. Two of the roofs were cast complete, and one roof had the center section cast (Fig. 3) which is normally rammed. The material was supplied by three separate manufacturers, and was cast in place using a vibrator with 8000-10,000 cycles/min.

The cast roofs lasted 26 and 23 heats, and the partial cast roof was in service for 24 heats. In each case, the roof center section was the cause of failure (Fig. 4) by burning out of the electrode port area. The castable material was only 11.5 per cent less in cost than the regular roof.

## REFRACTORY COOLING GLANDS

For many years, cast iron cooling glands were used around the electrodes. To reduce the cost associated with the glands, the writer's company has experimented with silica carbide, castable materials and ramming materials, but has not had satisfactory results from any of these; average of 52 heats with a maximum of 87 heats and a minimum of 21 heats. All furnaces are now operating without cooling glands.

These data concerning cooling glands are presented because of their generally accepted relationship to furnace roof life.

## SUMMARY

Reviewing the information obtained from observations of high alumina roof practice in the author's company, several pertinent and interesting facts were noted. Varying the composition of the refractory material and the method of construction resulted in (1) increased roof life, (2) reduced down time and (3) reduced cost per ton of steel. This was accomplished by utilizing 70 per cent alumina brick with replaceable center sections.

# MEETING THE CHALLENGE OF HIGH RELIABILITY CASTINGS

by W. Babington, E. E. Miller and W. R. Clitherow


## ABSTRACT

The needs of the authors' companies for high quality castings for use in military applications, the problems encountered in obtaining the required reliability in castings of all categories and the combined approach employed in reaching a solution are described. A dual program entered into jointly by the two organizations is described, covering (1) efforts aimed at pin pointing engineering and manufacturing requirements and improving the line of communication between the designers of one company and the engineering and purchasing departments of the second company and the producing foundries and (2) foundry surveys in all areas of casting effort to evaluate the capabilities of commercial foundries in order to differentiate between producers, as regards to their capacity to achieve and maintain all levels of quality required.

The manner in which these two phases of the program has been conducted is described. The results obtained to date, the quality improvements accomplished and the goals set for future work are discussed.


## INTRODUCTION

As man reaches for higher objectives and the space age becomes a reality, permissible deviations and error in technical areas are rapidly diminishing. The term reliability is heard on all sides, and its significance is emphasized in relation to all aspects of design and production in the missile field. Permissible variations have decreased in some areas from one in 1000 to one in 300,000 and even to one in one million, and no end is as yet in sight.

Requirements for high reliability castings for use in missile components are no exception to this trend. Castings having thin wall sections and close dimensional tolerances that a few years ago would have been deemed impossible are today being produced in large quantities. Castings relatively free from structural and metallurgical defects are being guaranteed even in extremely difficult sections. Great strides have been made in the development of new casting alloys by means of which many unique properties are becoming attainable.

This bespeaks a highly fluid state of the art of casting as related to the foundry industry. Since many of the requirements for such high reliability castings are space-age demands, they are frequently further complicated by being coupled to crash programs, the urgency of which usually does not permit normal development time for the achievement of the desired objectives.

## MISSILE COMPONENT DEVELOPMENT PROCESS

The authors' companies' team presently engaged in the design, development, testing, procurement and production of missile components has for many years had many problems in attempting to consistently obtain the quality of castings required for many of these complicated, sensitive and precise mechanisms. In the normal sequence of events in the development of missile components by this team, these procedures usually prevail:

1. The missile components, ground control equipments and related mechanisms are initially designed by one author's company's laboratories. These equipments usually include the use of castings of various types frequently requiring a high degree of dimensional precision, high reliability with regard to strength and soundness and often the use of unusual alloys capable of providing properties unique to the design involved.

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2. Following the design phase, prototype models of the apparatus must be prepared and placed under test in order to determine the adequacy of the design and the degree of reliability of the apparatus. It is standard practice in the development of prototypes to employ commercial production methods wherever practicable.

In many instances the production of development models will be undertaken by an author's company, taking advantage of their production and assembly facilities, and providing an opportunity for the company to gain manufacturing invaluable experience when and if production quantities of the specific equipment are required by the military. In addition to the manufacturing experience obtained by this practice, the company thus becomes more cognizant of the system design requirements.

This phase of development also affords an opportunity to determine the feasibility of the casting designs, and to establish satisfactory sources of supply for castings having the necessary quality level demanded by the specific applications.

3. As early as practicable, during the prototype development phase, manufacturing information concerning the unit under study is transmitted to the company, in order to expedite the necessary production planning. In all phases, the exchange of engineering information between the design organizations of one company and the manufacturing engineering groups of the other company is made as complete as is consistent with the extent and progress of the developmental studies. Every effort is made to effect a transition from the prototype and development phases to the production stage as smoothly and trouble free as is possible.

## CASTING PROCESSES

It is obvious that the problems discussed here relating to the procurement and use of high quality castings are common to both companies. The wide variety of applications for castings of all types in such equipments involves all the conventional processes employed for casting production. This includes sand, shell, permanent mold, plaster mold, investment and die casting. The range of materials covers the full scope of recognized engineering alloys, including aluminum, magnesium, copper-base, nickel-base, carbon and alloy steels, cast iron and nodular iron, stainless steels and limited quantities of high temperature materials. However, a large majority of the high reliability castings required for the missile applications herein described are produced in the light metals.

The figure shows a representative group of magnesium alloy castings produced by a number of methods, all of the high degree of reliability and quality required for use in a missile guidance system. These castings are typical of the types of design necessary for the somewhat unique missile applications. They cover a variety of magnesium alloys, including conventional aluminum-zinc-magnesium alloy AZ92, rare earth-zinc-magnesium alloy EZ33 and aluminum-manganese-magnesium alloy AM20. The latter alloy, while somewhat difficult to cast, has been found extremely useful due to its high damping and thermal conducting properties, as also has the zirconium-magnesium alloy K1 in other applications.

## PROBLEMS AND MEANS OF SOLUTION

The difficulties encountered in the past in obtaining the high quality castings for the applications already described were due to a number of causes, including inadequate communications, procurement and crash programs.

### *Inadequate Communication*

Communications via the casting drawings, between the two companies of the authors and the casting producers, involve the specific casting attributes required and means of checking these requirements both at source and at the point of receival. The qualities to be pin pointed included dimensional requirements, surface smoothness; soundness as determined by radiographs, magnetic particle and dye penetrants; mechanical properties; heat treatment; and chemical composition.

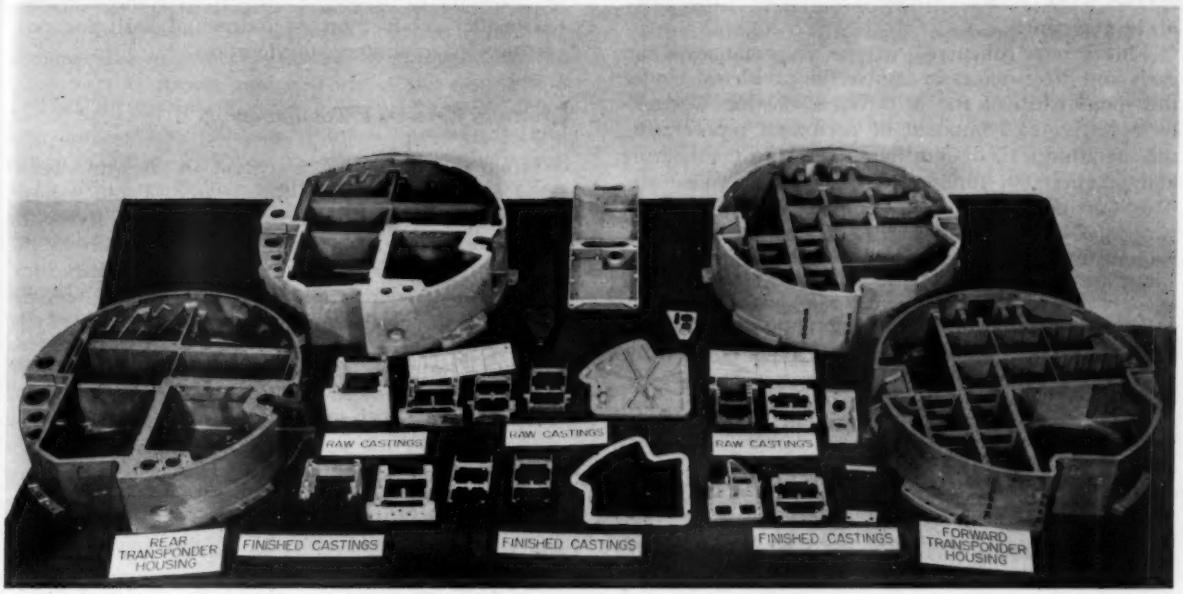
It was recognized that many of the casting specifications being employed were too general in nature and failed to specify means by which many of the important attributes should be measured. Supplementary notes frequently employed on casting drawings, in an effort to further control the required quality, were generally vague and difficult to interpret in terms of inspection procedures.

### *The Procurement Problem*

This problem of major concern in the field of high quality castings is closely associated with the relative capabilities of the various foundry producers. Operating under required and conventional purchasing procedures for military equipments, both of the companies must obtain multiple quotations for all lots of castings required.

In the past, the purchasing departments have, without complete information as to the relative capabilities of the various foundries being used, obtained quotations on given lots of castings from multiple sources. Orders were placed based on the most attractive quotations and the best delivery promises thus obtained, on sources which may or may not have equal capabilities.

It is apparent that foundries which operate without adequate quality control facilities, and are not burdened with the overhead that accrues therefrom, are frequently able to underbid those sources which do maintain satisfactory control and ship only castings capable of meeting high standards of quality. This practice inevitably results in the receival of castings of substandard quality, delayed delivery schedules and wasted machining effort on worthless castings. Loss of time from already overly tight schedules and the




**Typical high quality castings for missile applications.**

necessity for replacement of rejected parts is another result.

The responsibility for this situation cannot be placed solely with the foundry sources. As outlined before, it is recognized that an urgent need exists for the early exchange of information between the casting supplier and customer. This is in order that specific understandings are reached as to the exact requirements for the castings needed, together with a close scrutiny of the facilities available to the proposed vendor in order to guarantee the delivery of the exact casting quality desired prior to awarding a purchase order.

An intelligent approach to the procurement of high quality castings requires that detailed first-hand information, as to the capabilities of the various potential producers, be obtained. It also requires that only foundries having approximately equivalent production and quality control facilities shall be requested to quote on such work.

It is obvious that the shift in accent in the direction of improved quality and delivery requires that procurement procedures be revised. The emphasis in obtaining high quality castings for critical applications shall be (1) quality, (2) delivery, (3) cost, in the order named. It is believed to be significant that this need for a shift in emphasis, and resulting change in procurement procedures is concurred in by purchasing, engineering and design organizations alike at both the authors' companies.

### *Crash Programs*

These programs, associated with obtaining either limited prototype quantities or extensive production quantities of superior quality castings under a rush program as frequently exists in the military effort, present additional difficulties. Such schedules impose additional hardships on foundry sources, and certainly can be successful only if close and early liaison is established between the design engineers, manufacturing engineers and foundrymen. Thus, the producer has at an early date all the specific quality requirements, and in turn takes full advantage of the suggestions frequently offered by the foundrymen which may improve castability and reduce foundry delays.

### *Quality Levels*

Economically high reliability castings when required are relatively inexpensive compared with the overall costs of completed systems. It is obvious that all the casting designs required in a given missile system, or in any other equipment for military use, need not be of the highest quality obtainable. This fact further emphasizes the need for careful scrutiny of all casting drawings by those competent to determine the quality level necessary to meet the end-point requirements of the component in which the castings will be used.

Such a study of the design and early discussions of the potential casting with proposed producers will serve to categorize the casting as to the quality level required. Thus, procurement of adequate castings will be expedited for the operating conditions required of the finished product.

In view of the several problems outlined, and recognizing the urgent need for improvement in casting quality, a conference was called of those concerned with the quality and reliability of military equipment in both the authors' companies. All phases of the casting difficulties were discussed, and a recommended approach agreed upon. As a result of this discussion, the matter was then presented

to the joint Military Design Standards Committee of the companies.

This group concurred in the importance of an early and direct effort to resolve the problems. Under the sponsorship of the M.D.S.C. a Casting Committee was created composed of personnel representing the metallurgical, design, manufacturing engineering and procurement functions from both organizations. The Casting Standards Committee at once named two relatively small task groups charged with the responsibility of investigating and recommending solutions for the two major problems:

1) Improvement in the means of communicating specific casting quality requirements via casting drawings.
2) Appraisal of potential casting producers and evaluation of the relative capabilities of suppliers.

## DESIGN DOCUMENTATION TASK GROUP

This group was charged with the responsibility for ascertaining the adequacy of existing military casting specifications, recommending appropriate means for specifying the necessary quality requirements and means for checking these requirements, and for suitable screening procedures for all casting designs. The ultimate goal is that each casting drawing should be a complete document, delineating all practical dimensional tolerances and quality requirements commensurate with the end use of the casting under consideration.

This group first undertook a detailed study of all existing military specifications for castings, made by all commercial processes and in a wide range of alloys, in order to determine those most suitable for pin pointing the attributes considered essential in quality control. This investigation was made on an alloy basis, first covering aluminum and magnesium alloys since these materials are of prime importance in most missile components.

### *Specifications Study*

This study showed that there are adequate means for the control of high quality aluminum and magnesium castings by making use of selected sections of existing military specifications. Since this procedure requires referring to selected sections of current specifications in order to pin point the specific attributes considered essential, a series of tabulations were prepared for use by the design and drafting organizations. Each requirement to the appropriate specification or section thereof was referenced.

The study further revealed that casting soundness, considered to be the least common denominator of quality control, could be controlled with a high degree of accuracy by reference to radiographic standards made available by the American Society for Testing Materials. These standards are currently available for aluminum and magnesium alloy castings, and have been recently revised to provide even closer control of quality level. For materials other than those covered by such available standards, it will be necessary to develop radiographic standards from satisfactory product on an individual basis for each casting design under consideration.

### *Multiple Reference Tabulation*

From the tabulation a system of multiple references for use on drawings was devised. Table 1 shows an example of such a series of references. Recognizing the existence of a considerable gradation in the quality of castings required for various applications, an arbitrary classification of casting quality was devised. These classifications are:

Class A. Highest quality castings; capable of meeting the soundness requirements of Grade B or better per Spec. MIL-C-21180A or MIL-C-6021D.
Class B. Aircraft quality castings.
Class C. Commerical quality castings.
Class D. Run of the mill castings.

**TABLE 1 — TYPICAL METHOD OF SPECIFYING QUALITY CONTROL OF HIGH STRENGTH ALUMINUM ALLOY SAND CASTINGS**

| Requirement | Specification |
|---|---|
| Chemical Composition | MIL-C-21180<br>ALLOY A-356 or C-355 |
| Temper | — |
| Mechanical Properties | MIL-C-21180<br>CLASS XII |
| Penetrant Inspection | MIL-C-21180 |
| Radiographic | MIL-C-6021<br>GRADE B |
| Sampling for Penetrant and Radiographic Inspection | MIL-C-21180 |
| Workmanship and Repair of Casting | QQ-A-601 |

Cross-reference tabulations of requiremens, and suitable existing specifications or sections thereof, have been prepared covering aluminum alloys, magnesium alloys, stainless steels and nickel-copper-silicon alloys. These specifications have been distributed to all military design and drafting organizations for use in specifying materials requirements for casting drawings. This work will be continued to include suitable reference tables to cover appropriate specifications for the commonly used copper-base alloys, carbon and alloy steels, cast iron, nodular iron and high temperature materials. It is obviously intended that the nature of the materials "call-out" on the casting drawing shall be such as will indicate, to the procurement organizations and to the potential producer, the quality of casting required.

As a natural sequel to the system of employing multiple specification references for high quality castings, appropriate steps have been taken to encourage all drafting and design organizations to submit all casting designs for examination and consultation with the Metallurgical Engineering Group. By this means recommendations are made to the designers.

These recommendations concern the selection of alloys to meet the environmental, physical and mechanical requirements of the part under design, sug-

gestions as to the most practical and economical casting process to be employed, comments on the dimensional tolerances specified in the light of the alloy, casting process and end use of the casting and a check on the specifications referenced on the drawing for composition, mechanical properties and quality control. This practice, it is believed, will provide drawings which will carry all information pertinent to the degree of dimensional and quality control required.

In addition to these consultations, the designers, manufacturing engineers and purchasing personnel are being encouraged to discuss the potential casting designs at as early a stage as is feasible with competent representatives from the potential foundry. This is in order to resolve the design and production problems before pattern or die equipment are produced.

This program has been in use for several months, and its worth has been well established. Enthusiastic cooperation has been obtained from the designers, manufacturing engineers, purchasing representatives and casting producers alike, and marked improvement in quality of castings received and in the delivery schedules achieved, together with reductions in overall costs has been achieved.

## FOUNDRY APPRAISAL TASK GROUP

The foundry appraisal group was charged with undertaking a survey of as many as possible of the commercial foundries engaged in producing castings of the type employed in the electronic and missile field. The purpose of such surveys was to obtain first hand knowledge of the comparative capabilities of casting suppliers to produce the high quality castings required for this work. These surveys covered foundries producing under all the commonly employed casting processes—sand, shell, permanent mold, plaster, investment and die castings, and a wide range of alloys—aluminum, magnesium, copper-base, nickel-base, stainless steel, carbon and alloy steel, cast iron and nodular iron.

The group for this study was composed of one representative each from the Metallurgical departments, Engineering department and the Purchasing organization of the authors' companies. Initial plans were made by preparing a list of foundries which should be surveyed. For obvious reasons, the initial list included a large number of suppliers of aluminum and magnesium alloy castings and a substantial number of those sources from which both companies had obtained castings in the past. As the surveys progressed additional foundries were added to the tentative survey list, and interested foundries were encouraged to request a visit from the appraisal group.

In the interest of uniformity of results, and as a means for summarizing the information obtained from the surveys, an appraisal form was prepared listing the items to be documented during each foundry visitation. A code system was also devised to provide a means for classifying and analyzing the information obtained. Some of the features included in the survey form, pertaining both to the physical equipment available at the foundry and the information obtained through discussions with the key personnel of each supplier, are:

1. Casting processes employed.
2. Alloys commonly cast and those on which some experience had been obtained.
3. Engineering and quality control organizations.
4. Pattern and die shop facilities.
5. Foundry facilities for applicable processes.
6. Melting equipment and temperature controls.
7. Heat treating facilities and temperature controls.
8. Scrap segregation practices.
9. Sampling procedures.
10. Degassing techniques – means of checking.
11. Casting cleanup practices.
12. Facilities for (a) chemical analysis, (b) x-ray, (c) mechanical testing, (d) magnetic particle and surface penetrant inspection.
13. Inspection personnel and techniques.
14. Layout facilities (dimensional checks on product).
15. Approximate plant capabilities.

In addition to the listed items relative to the physical facilities, informal discussions were held with the foundry management and key personnel to determine the general attitude toward participation in programs frequently requiring:

1. Substantial research and development effort prior to production.
2. Furnishing small quantities of high quality castings for prototype development with future production dependent on the evaluation obtained in the development phase.
3. Relatively high quantity production of superior quality castings.

Space was also provided in the survey form for general comments regarding the conclusions reached by the survey group as to the apparent capabilities of the plant surveyed.

### *Appraisal and Classification System*

The code system developed for appraising and classifying the capabilities of each foundry is shown in Table 2. As indicated in this table, letter and numeral codes were employed to designate the alloys covered, the casting processes employed, the apparent potential quality appraisal and the capabilities of the foundry in terms of R&D work and/or production effort. A typical coding is also included in the table to demonstrate the manner in which the classifications were made.

The foundry surveys were started in Feb. 1960, and are continuing as rapidly as time and available personnel permits. The appraisal form is completed immediately after each survey. The information obtained from each such visitation is kept in complete confidence and only discussed with the interested foundry at their request. The tabulation of quality ratings is kept up to date, and forwarded only to a restricted list of purchasing and engineering personnel of the authors' companies.

A check on the accuracy of each appraisal is obtained through monthly inspection reports covering

TABLE 2 — CODE SYSTEM USED IN FOUNDRY APPRAISAL

| Casting Quality | Metals Cast |
|---|---|
| A — Capable of producing Grade B or better per Spec. MIL-C-21180A or MIL-C-6021D | A — Aluminum and magnesium |
| B — Aircraft Quality | $A_A$ — Aluminum only |
| C — Commercial Quality | $A_m$ — Magnesium only |
| D — "Run-of-Mill" Quality | B — Copper-Base Alloys |
| 1 — Research & Development Only | C — Nickel alloys, stainless steels, high temperature alloys |
| 2 — Quantity Production Only | D — Carbon and alloy steels, cast iron, nodular iron |
| 3 — R&D and Production | $Z_n$ — Zinc alloys |

| Types of Castings | Typical Appraisal Rating |
|---|---|
| s — sand cast | A-A-3-s-pl-pm |
| dc — die cast | This rating indicates a foundry capable of carrying on R&D effort as well as quantity production in aluminum and magnesium castings of highest quality in the sand, plaster and permanent mold processes. |
| pm — permanent mold | |
| ps — precision sand | |
| im — investment mold | |
| sh — shell | |
| pl — plaster mold | |
| cfg — centrifugal | |

the receival of all castings, the source of the castings, the number of rejections for all causes and the adherence to promised schedules. By this means, the appraisal ratings may be kept up to date in the light of capability based on actual experience. Those foundries given ratings lower than Class A on the initial survey, may at any time, request a reappraisal, based on indicated improvements in facilities and/or quality control personnel, which in the opinion of the survey group would merit such a re-examination.

## RESULTS AND CONCLUSIONS

The documentation group has completed and distributed the necessary specification references for high reliability castings covering the commonly used aluminum-base alloys, magnesium-base alloys, stainless steels and monels. Work is continuing on similar tabulations for copper-base alloys, carbon and alloy steels and for cast and nodular iron. Revisions in the drafting practices for conveying the necessary information on the casting drawings have been placed in effect.

All groups in the military areas interested in the design of high reliability castings are being urged to discuss their proposed designs and requirements with the laboratories' metallurgists, and excellent cooperation in this matter is being demonstrated. As a result of this total effort, a marked improvement has been observed in the communication of information via the casting drawings from designers to manufacturing engineers to purchasing personnel to foundrymen.

The work of the foundry appraisal group has been highly productive, although somewhat disappointing as regards to the number of foundries visited which were felt to be capable of Class A effort. Up to March 1, 1961, 68 foundries were carefully surveyed and classified. In this group 16 foundries were reported as capable of producing Class A quality castings in one or more categories of casting processes and metals, and 6 foundries found capable of producing Class A quality castings in aluminum and magnesium alloys. The light metal production and capabilities are obviously of major interest in view of the current activity in missile applications.

As previously indicated, the appraisal rating information has been made available to the authors' companies' purchasing organizations as rapidly as possible. Indications of the effectiveness of this program, and the use of classified producers exclusively for high reliability applications are demonstrated from these facts:

> During the first quarter of 1960, at which time this program was hardly underway, castings were being received at the Greensboro, N.C. plant of the author's company, where practically all the missile guidance castings are being machined, at a rate of approximately 2000 castings per month, but with a rejection rate as high as 50 per cent during certain periods. In comparison, receival of castings at the same plant during the fourth quarter of 1960 was approximately 8000 castings per month, with an average rejection rate of slightly under one per cent. During Jan. 1961, receivals at this same plant were 11,249 castings, with but 54 rejections (approximately 0.5 per cent).

### *Program Aims*

One of the important aims of this program is that drawings covering castings of all types and gradations in quality shall be carefully reviewed in the light of their ultimate usage. All important properties shall be indicated with exactness, including dimensional tolerances, composition, mechanical strength, soundness and surface quality, and these requirements shall be realistic and meaningful. It is also necessary that the importance of these attributes be conveyed in the strongest terms to casting suppliers, and that the foundryman be made aware that these drawings reflect minimum requirements, deviations from which must be negotiated before the fact.

An indirect, but important advantage, which will accrue from the successful administration of this program will be that a reduction in the number of problems relating to casting quality will result in a substantial saving in engineering time required to resolve such difficulties and will free essential manpower for important basic development work.

Based on the results obtained to date, there is every reason to believe that future benefits of this coordinated program will continue to mount. It is realized that it is somewhat revolutionary to carry on a procurement program with the emphasis on quality, delivery and price in the order named, and to have the complete cooperation of the purchasing, engineering, design and research organizations. This is now an accomplished fact.

It is further felt that the evident success of such a program should be a powerful incentive to foundrymen in all categories to meet the ever growing challenge of supplying to those engaged in military communication and missile programs the high reliability castings needed for such effort.

# PRESS FORGING OF HIGH ALLOY STEEL CASTINGS

by T. Watmough, J. T. Berry and P. R. Gouwens


## ABSTRACT

The press forging of 18-8 stainless steel, 17-4 precipitation hardening stainless steel and S.A.E. 4340 steel sand castings under carefully controlled conditions of temperature and deformation has been investigated. Considerable increases in strength are obtainable under optimum conditions. The most suitable alloys for the technique are low carbon, high alloy types of steel, where the martensite can sustain deformation. In the case of 18-8 stainless steel, strength increases of 400-600 per cent in yield strength and 200-250 per cent in ultimate tensile strength were procured by press forging at −100 F with 40 per cent deformation, followed by tempering at 650 F.

In the case of the 17-4 PH steel, 10 per cent increases in both yield and ultimate tensile strength were obtained by press forging the predominantly martensitic structure 10 per cent at room temperature, followed by aging at 860 F. S.A.E. 4340 is responsive to strengthening by warm working prior to transformation, and at the 240,000 psi strength level the yield point can be increased by 15 per cent and the ultimate tensile strength by 10 per cent. The composition of 4340 is not entirely suited to these techniques.

The process was applied to a specific air frame component and comparisons made between conventional castings, forgings and press forged castings produced in 18-8 stainless steel and 17-4 PH stainless steel. The most encouraging results in the press forging of castings were in the 18-8 stainless steel, deformed 30 per cent at −100 F. Utilization of the process with the low carbon, high alloy steels is feasible, but not without certain difficulties. These involve the design of closed dies of adequate strength to press forge at the lower temperatures. Selective press forging of critical areas in castings by semi-open die techniques, is perfectly feasible.


## INTRODUCTION

Although the use of steel castings in aircraft has been opposed, to some extent by designers, their applications offer many economic advantages. This opposition was based upon the incidence of internal flaws and variable metallurgical quality found in many commercially produced castings. However, against this must be set the cost reduction and manufacturing speed associated with the latter. The improvement in the mechanical properties and concurrent overall increase in casting quality would render castings more acceptable. Cold or warm working constitutes a possible means of approaching this goal. The specific advantages of a duplex process would be numerous:

1. Higher tensile strengths, even up to ultra-high strength levels, when compared to normal castings of the same composition.
2. Corresponding increases in yield and proof strength without the same degree of ductility decrease usually associated with strengthening by heat treatment only.
3. Relative freedom from the marked directional properties encountered in most forgings.
4. Reduced lead time and costs when compared to forgings for short-run or prototype parts.

A research program was undertaken to determine the effect of various press forging and heat treatment conditions on the properties of steel castings. The investigation was conceived as having two separate phases. First, a test program was undertaken to cast test bars and to establish the following forging conditions to obtain optimum properties—1) per cent deformation, 2) forging temperature and 3) heat treatment prior to and after forging.

The above conditions were to be established for an alloyed carbon steel—S.A.E. 4340, a nonheat treatable austenitic stainless steel—CF8C and an age-hardenable austenitic stainless steel—17-4 PH.

The second phase was to select a production forging for redesign and revised processing as a press forged casting. This part was to be cast, press forged and tested to demonstrate the improvement in

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strength gained by press forging, as well as to compare the mechanical characteristics with the original forging.

A reference[1] describes in detail the results of the first phase of this investigation, particularly for the 18-8 stainless steel and the 17-4 PH precipitation-hardening steel. With the S.A.E. 4340 material, processing difficulties along with the general unsuitability of the alloy led to only small amounts of property improvement and contrasted with the promising results obtained with the two high alloy steels.[2]

Consequently, investigation of property improvement in a production aircraft casting was pursued as:

1. 18-8 stainless steel (to CF8C spec) to be press forged to secure 30 per cent deformation at −100 F.
2. 17-4 PH stainless steel to be press forged to secure 10 per cent deformation at room temperature.

The properties obtained would then be compared with commercial forgings and castings.

## SELECTION, DESIGN AND PRODUCTION OF CASTINGS

An airframe part, strut-engine rear-mount trunnion, was selected for subjection to press forging and subsequent evaluation. Shown in Figs. 1 and 2 are the rough forging and finish machine drawings of this part.

To procure the deformation necessary in each of the selected materials it was necessary to design cast forging blanks which, after deformation, would conform to rough forging dimensions, as shown in Fig. 1. The drawings for the cast blanks are shown in Figs. 3 and 4 for the CF8C steel and 17-4 PH steel respectively.

This schedule details the production of castings:

10 cast forging blanks in CF8C, as in Fig. 3.
10 cast forging blanks in 17-4 PH, as in Fig. 4.
10 castings in CF8C as in Fig. 1.
10 castings in 17-4 PH, as in Fig. 1.

A dry sand molding technique was employed for the castings. Metal was melted in a 600 lb high frequency induction furnace. Mean pouring temperatures of the castings involved were in the range of 2850-2900 F (1565-1593 C). The maximum exothermic type powder was added to the riser tops after pouring. Figure 5 shows a view of a casting with runners and risers attached.

## PRESS FORGING DIES PRODUCTION

To press forge the cast blanks to the desired final dimensions, shown in Fig. 1, a three-stage semi-open press forging sequence was employed. Two sets of dies were required for these operations—one set of flat dies for the end portions of the blanks, and one set for the center rib portion. The dies were cast in H13 composition high strength die steel, and a ceramic molding technique was employed to obtain the necessary dimensional control. Shown in Fig. 6 is one half of the press forging die for the center rib portion of the castings.

The dies were heat treated to a Rockwell hardness figure of approximately Rc50. The upper and lower faces of the flat dies to be utilized for press forging the end portions of the castings were ground flat and parallel. The lower face only, i.e., the face which rests on the press, was ground on the pair of dies for the center portion of the castings.

## PRESS FORGING OF CAST BLANKS

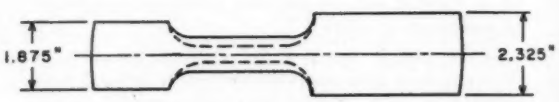
### *CF8C (18-8 Stainless) Steel Castings*

Prior to working, all castings for press forging were homogenized at 2050 F (1127 C) for 2 hr and water quenched. Using the open dies mentioned previously, 18-8 stainless steel castings were press forged at −100 F. A 1000 ton press was utilized for the work. The press forging of the strut-engine rear-mount trunnion was done in three stages:

1. Large end portion between two flat dies.
2. Small end portion between two flat dies.
3. Center portion between dies illustrated in Fig. 6.

Detailed in Table 1 are the amounts of deformation procured in each section as determined by differences in thickness dimensions over the parts press forged.

TABLE 1 — PRESS FORGING DETAILS OF CF8C (18-8) CASTINGS




| Casting No. | Deformation, % | | |
|---|---|---|---|
| | Small End | Middle | Large End |
| 1 F | 30.0 | 50.0 | 29.0 |
| 2 E | 30.0 | 47.0 | 29.0 |
| 3 E | 32.0 | 54.0 | 31.0 |
| 4 E | 34.0 | 48.0 | 32.0 |
| 5 F | 30.0 | 45.0 | 30.0 |
| 6 E | 30.0 | 52.0 | 28.0 |
| 7 F | 30.5 | 40.0 | 30.0 |
| 9 E | 32.0 | 45.0 | 30.0 |
| 10 F | 29.6 | 47.0 | 28.5 |
| 11 E | 32.5 | 41.0 | 27.0 |
| 12 F | 34.0 | 47.0 | 29.0 |

### *17-4 PH Stainless Steel Castings*

Prior to the forging operation the castings for press forging were given this heat treatment:

1. Heat to 2025 F (1107 C)−2 hr−air cool.
2. Heat to 1950 F (1066 C)−2 hr−oil quench.
3. Subzero treatment at −100 F for minimum of one hr.

A similar technique to that adopted for press forging the 18-8 stainless steel castings was employed for the 17-4 PH castings.

The press forging was conducted at room temperature. Detailed in Table 2 are the amounts of

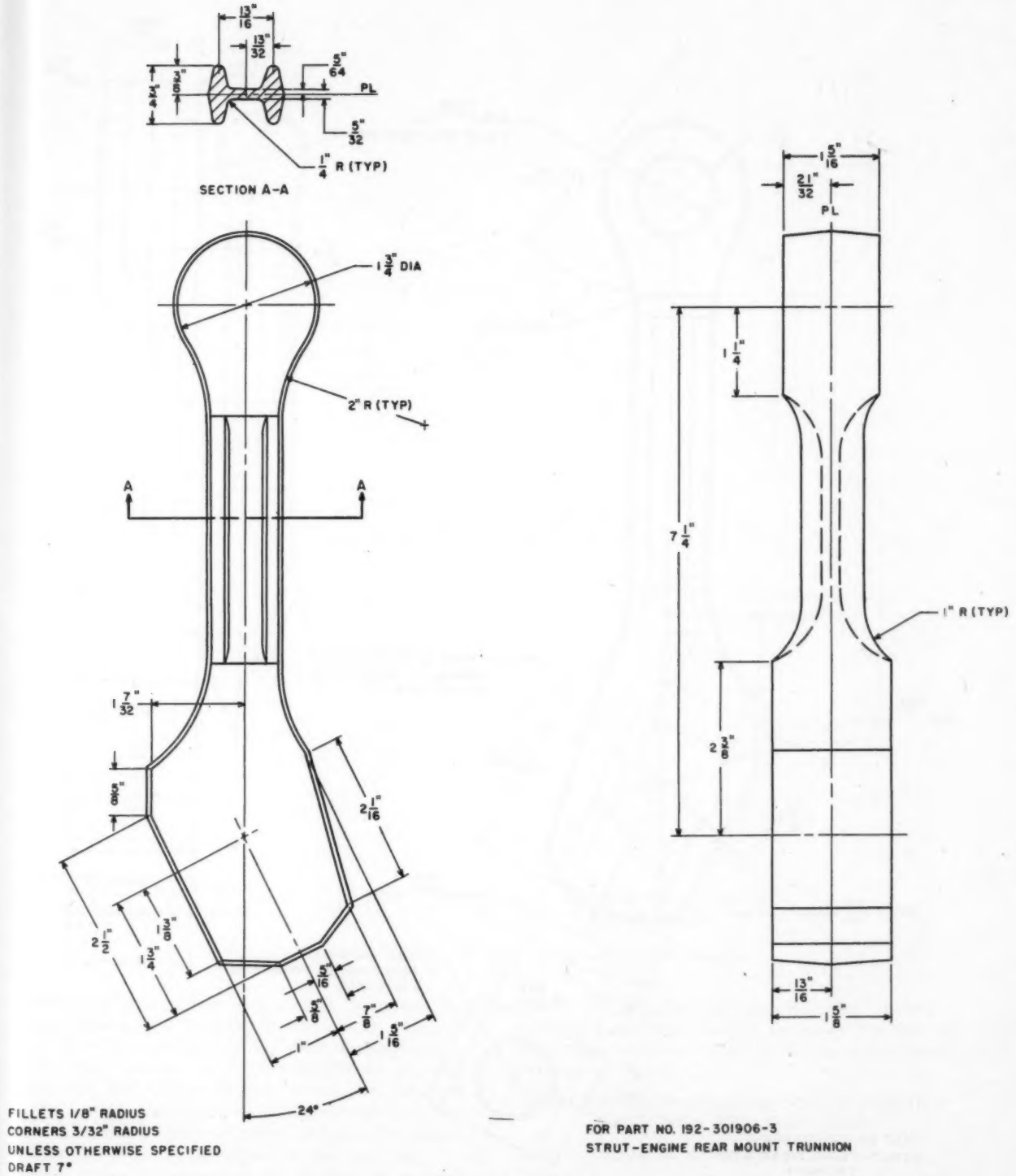



Fig. 1 — Rough forging for rear mount trunnion strut.

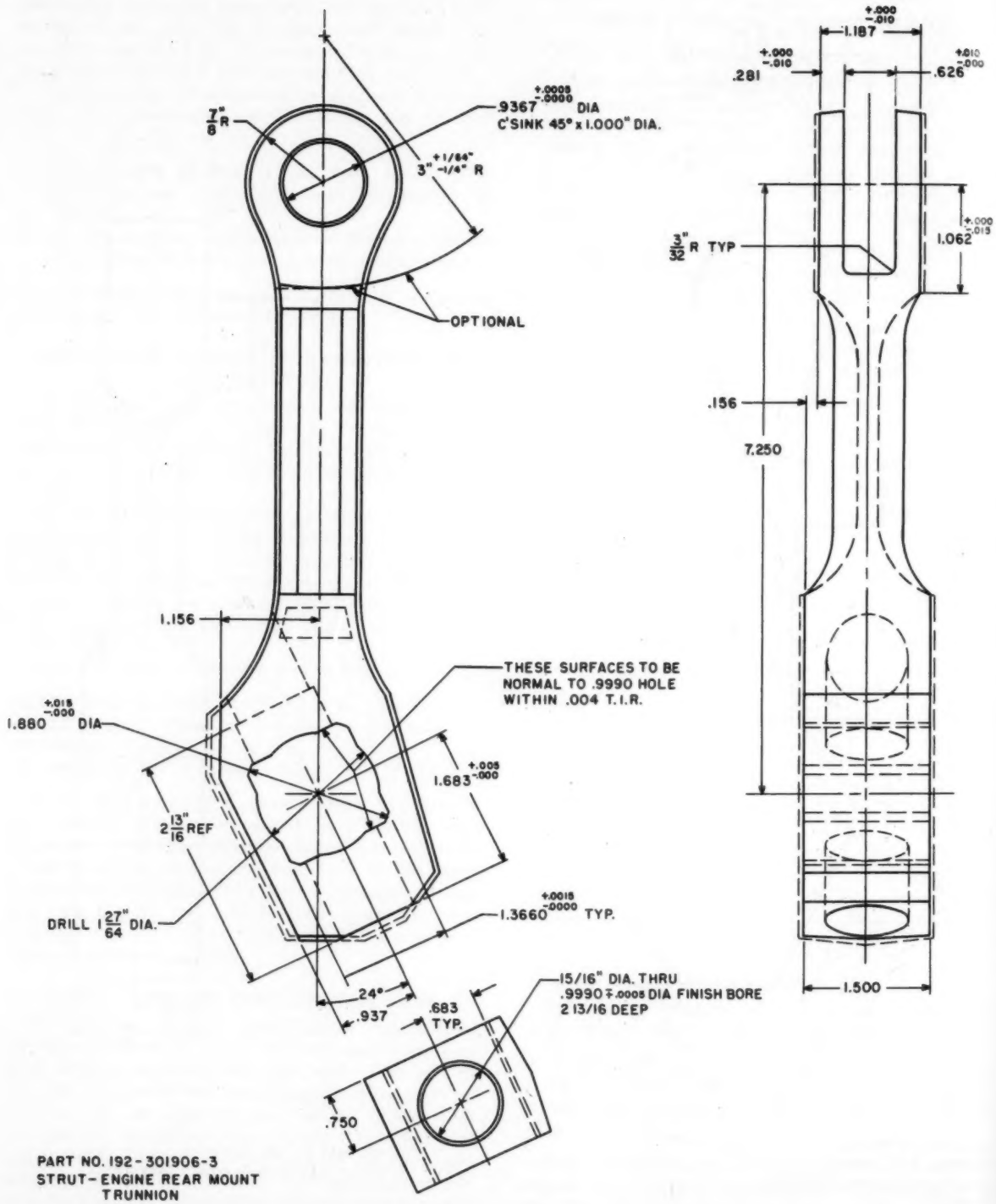



Fig. 2 — Finish machine drawing for rear mount trunnion strut.

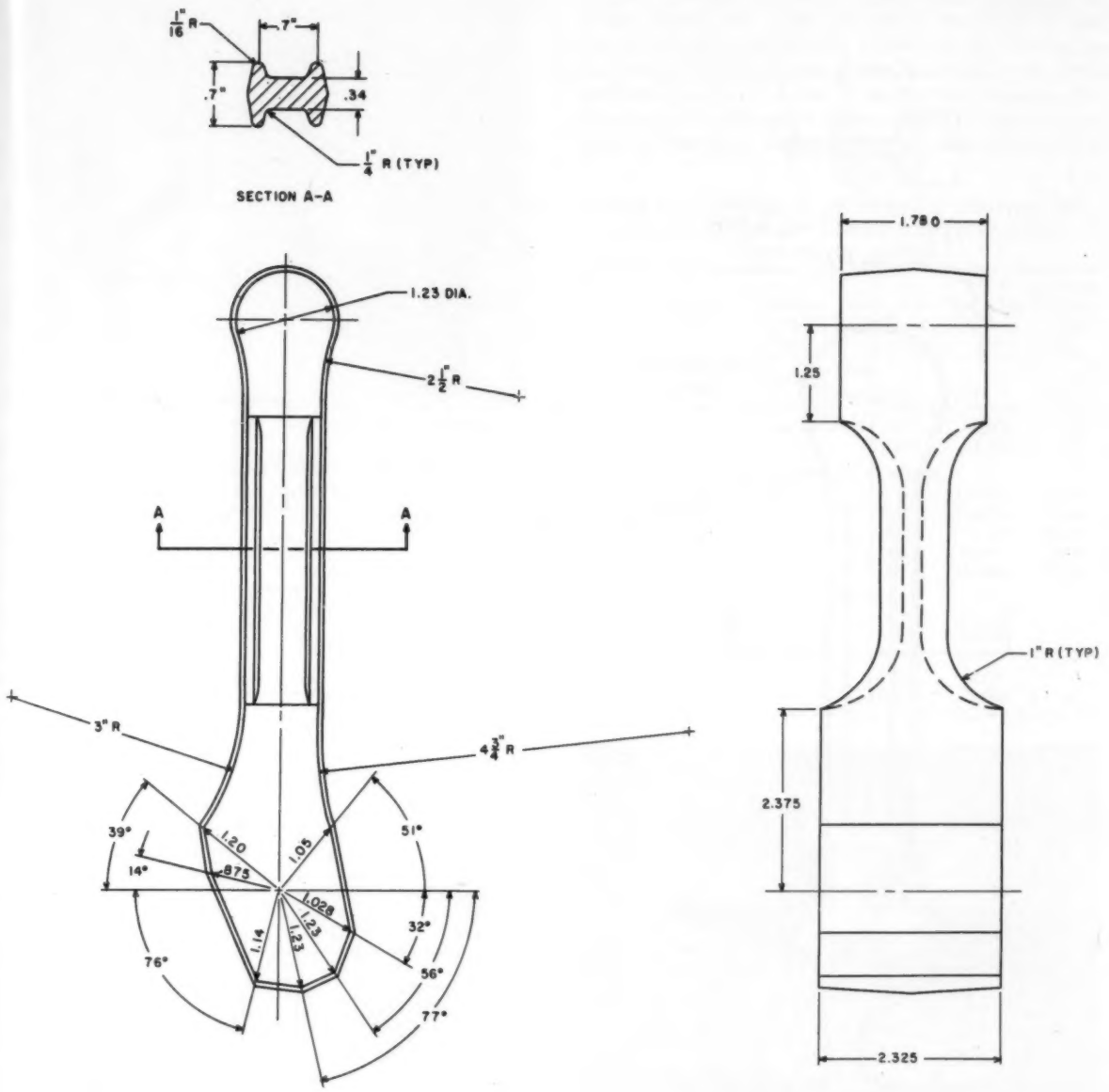



**Fig. 3 — AMS5363A cast forging blank to be deformed 30 per cent.**

deformation procured in each section. In several instances cracks were initiated in the center web portion of these castings during press forging operations. This was due to the biaxial stresses imposed upon this web during forging. When this occurred in the first few castings, the die was modified slightly to reduce the severity of these stresses. The modification reduced the incidence of the cracking.

This system of semi-open die forging of selected areas which in total covered the cast blanks was considered successful.

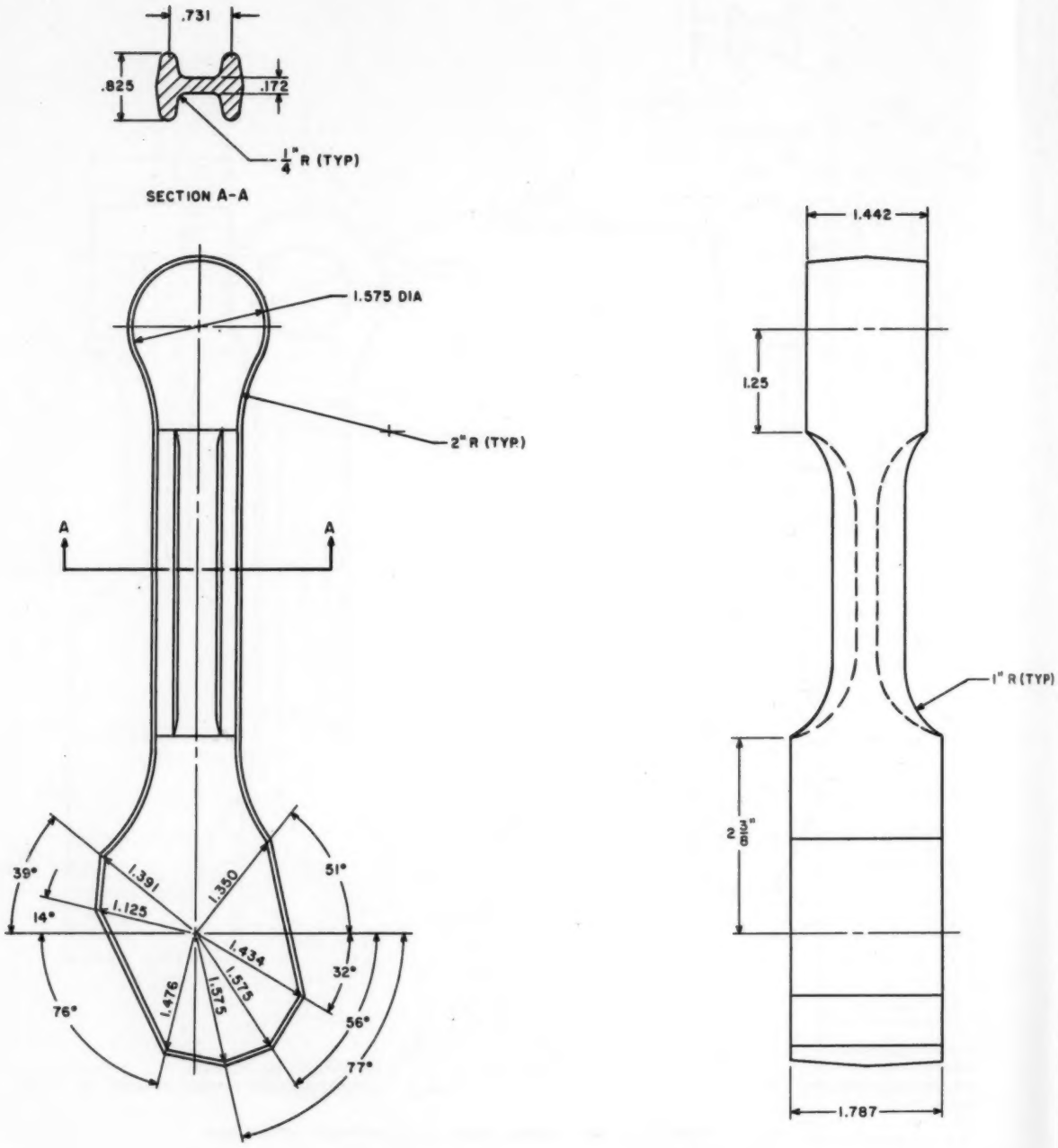
After the press forging operation, five of each of the press forged cast blanks, the press forgings and the castings were machined to the principal dimensions shown in Fig. 2.

## TESTING OF FORGINGS, PRESS FORGED CASTINGS AND CASTINGS

### *Tensile Loading of Actual Strut Part*

Figure 7 illustrates the typical test rig for tensile loading to destruction of the subject parts.

**CF8C (18-8) Stainless Steel.** In the case of the press




FILLETS 1/8" RADIUS
CORNERS 3/32" RADIUS
UNLESS OTHERWISE SPECIFIED
DRAFT 7°

MAT'L : 17-4 PH STAINLESS STEEL
FOR PART NO. 192-301906-3
STRUT-ENGINE REAR MOUNT TRUNNION

Fig. 4 — 17-4 PH cast forging blank to be deformed 10 per cent.

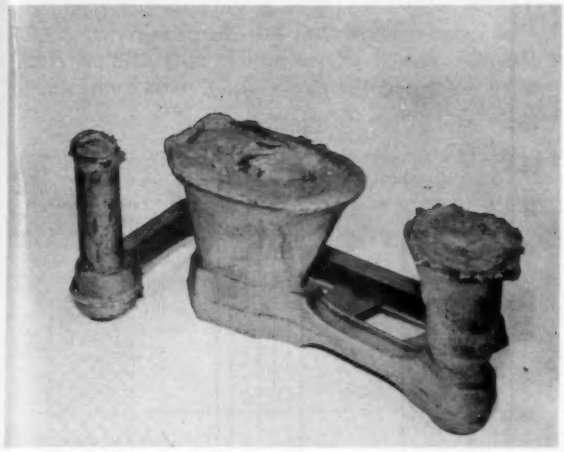
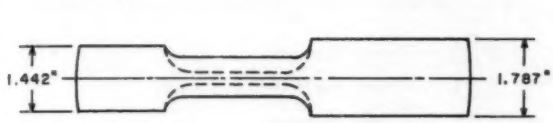


Fig. 5 — Strut engine rear mount trunnion castings.

TABLE 2 — PRESS FORGING DETAILS OF 17-4 PH CASTINGS




| Casting No. | Deformation, % | | |
|---|---|---|---|
| | Small End | Middle | Large End |
| 1 H | 13.3 | 10.6 | 11.5 |
| 2 H | 12.5 | 23.0 | 13.5 |
| 3 H | 10.2 | 19.5 | 8.7 |
| 4 G | 9.4 | cracked | 10.2 |
| 5 G | 5.5 | cracked | 9.4 |
| 6 G | 12.4 | cracked | 11.6 |
| 7 G | 9.5 | cracked | 10.2 |
| 8 G | 9.5 | cracked | 13.0 |
| 9 G | 9.0 | cracked | 12.0 |
| 10 H | 13.0 | 18.2 | 11.0 |

(TYP)

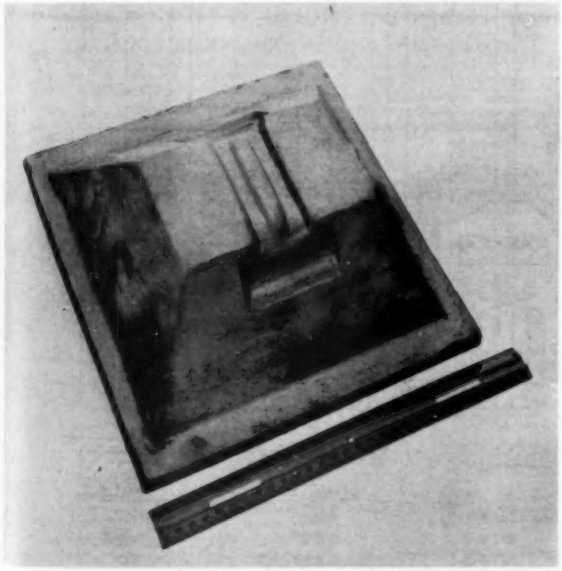


Fig. 6 — Press forging die for center portion of castings.

forged castings the strut parts were tempered at 650 F (344 C) for 2 hr. Exhibited in Table 3 and Fig. 8 are the comparative results of the testing to destruction of the similar parts processed in the three different manners. Table 3 shows the apparent deformations experienced in the center of the selected part of the press forged castings. This deformation

TABLE 3 — RESULTS OF TENSILE TESTING TO DESTRUCTION OF STRUT PARTS IN CF8C STAINLESS STEEL

| No. | Breaking Load, lb | Location of Break | Area at Site of Break, sq in. | Mean Breaking Stress, psi | Apparent Deformation, % |
|---|---|---|---|---|---|
| **Forgings** | | | | | |
| 18-1 | 29,500 | Small Ends | 0.370 | 79,700 | — |
| 18-2 | 31,100 | Small Ends | 0.364 | 85,500 | — |
| 18-3 | 31,000 | Small Ends | 0.361 | 86,000 | — |
| 18-4 | 31,000 | Small Ends | 0.367 | 84,500 | — |
| **Press Forged Castings** | | | | | |
| 2 E | 49,500 | Center | 0.42 | 118,000 | 47.00 |
| 3 E | 44,000 | Center | 0.42 | 105,000 | 54.00 |
| 4 E | 46,800 | Center | 0.42 | 111,000 | 48.00 |
| 9 E | 44,200 | Center | 0.42 | 105,500 | 45.00 |
| 12 F | 43,700 | Center | 0.42 | 104,000 | 47.00 |
| **Castings** | | | | | |
| 1 A | 29,300 | Center | 0.42 | 70,000 | — |
| 2 A | 27,000 | Center | 0.42 | 64,500 | — |
| 3 A | 24,600 | Center | 0.42 | 59,000 | — |
| 4 A | 27,000 | Center | 0.42 | 64,500 | — |
| 5 A | 30,000 | Center | 0.42 | 72,000 | — |

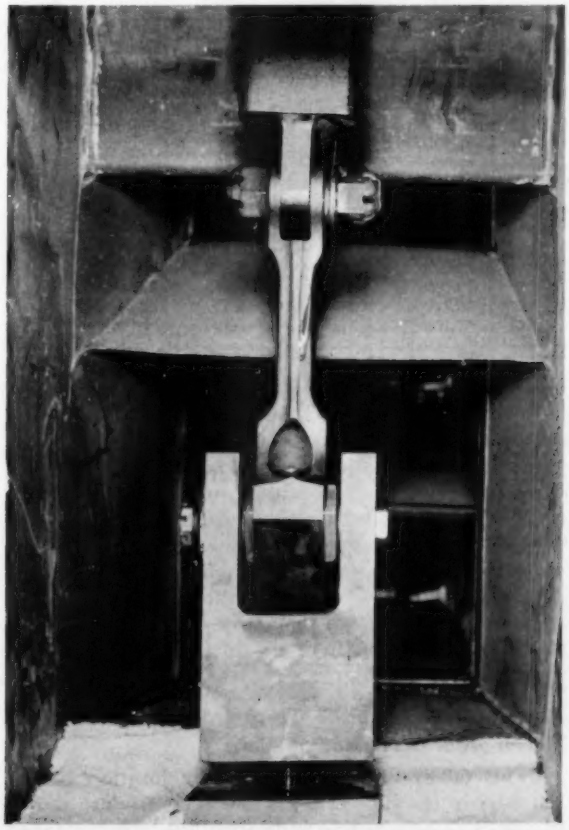


Fig. 7 — Typical test rig for tensile testing of strut part.

was computed by measuring the thickness change in the center of the web section after deformation, but because of the complex shape of the cross-section the figure quoted does not represent the true overall mean deformation.

The true deformation (as indicated in Fig. 1 of reference 1), throughout the whole cross-section, was of the order of 15 per cent. It can be seen, however, that in spite of this limitation the breaking load and mean breaking stress in the press forged castings are appreciably higher than in both castings and forgings.

**17-4 PH Stainless Steel.** The parts were aged, prior to testing, at these temperatures:

Forgings ..... 925 F (491 C) – 2 hr
Press Forged Castings ..... 860 F (455 C) – 2 hr
Castings ..... 925 F (491 C) – 2 hr

Shown in Table 4 are the results of tensile testing to destruction of strut parts in 17-4 PH stainless steel.

TABLE 4 — RESULTS OF TENSILE TESTING TO DESTRUCTION OF STRUT PARTS IN 17-4 PH STAINLESS STEEL

| No. | Breaking Load, lb | Location of Break | Area at Site of Break, sq in. | Mean Breaking Stress, psi | Apparent Deformation, % |
|---|---|---|---|---|---|
| **Forgings** | | | | | |
| 17-1 | 72,100 | Small Ends | 0.364 | 198,000 | — |
| 17-2 | 72,700 | Small Ends | 0.364 | 200,000 | — |
| 17-3 | 70,500 | Small Ends | 0.367 | 191,000 | — |
| 17-4 | 72,500 | Small Ends | 0.370 | 196,000 | — |
| 17-5 | 72,100 | Small Ends | 0.370 | 195,000 | — |
| **Press Forged Castings** | | | | | |
| 1 H | 67,900 | Small Ends | 0.370 | 184,000 | 13.3 |
| 2 H | 50,400 | Center | 0.42 | 120,000 | 23.0 |
| 3 H | 61,500 | Center | 0.42 | 146,000 | 19.5 |
| 10 H | 37,000 | Center | 0.42 | 88,000 | 18.2 |
| 5 G | 29,100 | Center | 0.42 | 69,500 | cracked |
| **Castings** | | | | | |
| 1 C | 43,700 | Small End | 0.354 | 124,000 | — |
| 2 C | 58,800 | Small End | 0.366 | 160,000 | — |
| 3 C | 38,400 | Small Ends | 0.375 | 102,500 | — |
| 4 C | 63,700 | Center | 0.42 | 152,000 | — |
| 5 C | 61,100 | Center | 0.42 | 146,000 | — |

The results given in Table 4 for the press forged castings and the castings were disappointing. This particular casting alloy appears to be extremely sensitive to minor surface and subcuticular defects, since in fact all fractures involved such features. As will be seen, when these extremely minor defects are removed, the castings and press forged casting compare extremely favorably with the press forgings. In addition to the effect of the above stress concentrators, mention must be made of the effect of part geometry upon biaxiality of stress.

### *Machining and Tensile Testing of Specimens Taken from Strut Engine Parts*

**CF8C (18-8) Stainless Steel.** The location of the test bars in the subject parts and their size are shown in Fig. 9. The schedule indicating the number taken from each site is:

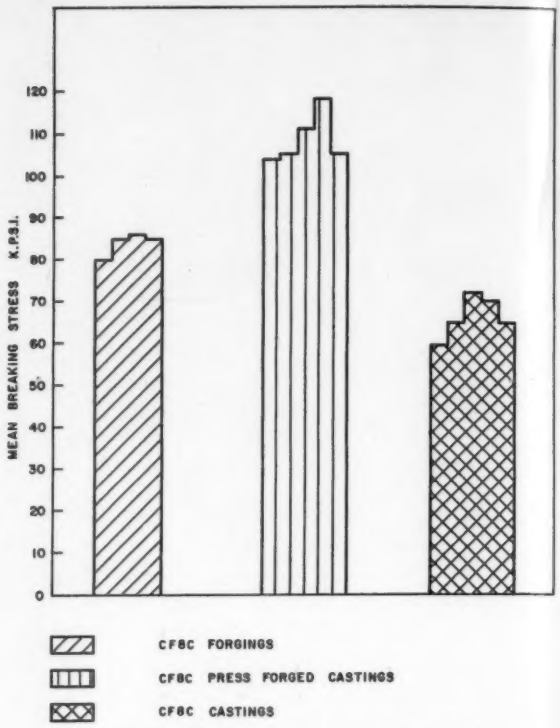
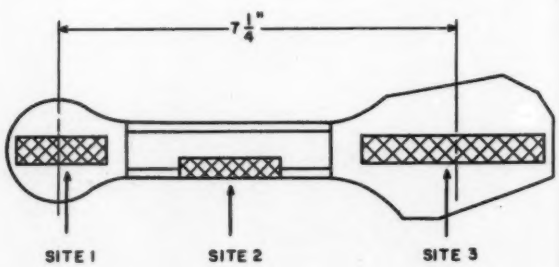



Fig. 8 — Comparison of mean breaking stress of CF8C stainless steel forgings, press forged castings and castings.

| | Site 1 | Site 2 | Site 3 |
|---|---|---|---|
| Forgings ..... 2 | 2 | 2 | 2 |
| Press forged castings ..... 2 | 2 | 2 | 3 |
| Castings ..... 2 | 2 | 2 | 2 |

In the case of the press forged castings, the test bars were tempered at 650 F (343 C). Shown in Tables 5, 6 and 7 and in Fig. 10 are the results of these tensile tests carried out upon the forgings, press forged castings and castings.

The data presented in these three tables are par-




TEST BAR SIZES

| | DIAMETER | LENGTH |
|---|---|---|
| SITE 1 | 0.160" | 1.5" |
| SITE 2 | 0.126" | 0.96" |
| SITE 3 | 0.252" | 3.0" |

Fig. 9 — Location and size of test bars machined from castings, press forged castings and press forgings.

ticularly encouraging insofar as the press forged castings are concerned. In certain instances, e.g., test bars at Site 1 in Table 6, the tensile strength has been more than doubled. It should again be noted that the deformations associated with the center portion of the part are only apparent and not true values.

**17-4 PH Stainless Steel.** The location of the test bars in the subject parts and the size of the test bars at each location are shown in Fig. 9. The schedule indicating the number of bars taken from each site and the aging treatments given is:

| | No. of Test Bars, Sites 1, 2, 3 | Aging Treatment |
|---|---|---|
| Forgings | 2 each | 925 F (491 C) — 2 hr |
| Press forged castings | 2 each | 860 F (454 C) — 2 hr |
| Castings | 2 each | 925 F (491 C) — 2 hr |

Shown in Tables 8, 9 and 10 and Fig. 11 are results of the tensile tests carried out upon the forgings, press forged castings and castings.

The results presented in these tables are again encouraging in that, where deformation of the section of the castings could be procured—namely, at the ends

**TABLE 5 — TENSILE TEST RESULTS FROM CF8C (18-8) STAINLESS STEEL FORGINGS**

| Site | Test Bar Diameter, in. | 0.2 % Yield Stress, psi | U.T.S., psi | Elong., % | R.A., % |
|---|---|---|---|---|---|
| 1 | 0.160 | 44,600 | 88,910 | 72 | 71 |
| 1 | 0.160 | 51,500 | 96,080 | 50 | 70 |
| 2 | 0.126 | 40,300 | 88,300 | 86 | 72 |
| 2 | 0.126 | 40,300 | 86,700 | 83 | 70 |
| 3 | 0.252 | 32,000 | 79,000 | 60 | 72 |
| 3 | 0.252 | 30,000 | 80,400 | 61 | 72 |

**TABLE 6 — TENSILE TEST RESULTS FROM CF8C (18-8) STAINLESS STEEL PRESS FORGED CASTINGS**

| Site | Test Bar Diameter, in. | Deformation at-Site, % | 0.2% Yield Stress, psi | U.T.S., psi | Elong., % | R.A., % |
|---|---|---|---|---|---|---|
| 1 | 0.160 | 30.5 | 172,000 | 174,000 | 9 | 30 |
| 1 | 0.160 | 29.6 | 184,000 | 188,000 | 5 | 15 |
| 2 | 0.126 | 38.0 | 102,000 | 114,700 | 27 | 50 |
| 2 | 0.126 | 41.0 | 98,000 | 112,000 | 37 | 51 |
| 3 | 0.252 | 30.0 | 147,600 | 154,400 | 6 | 32 |
| 3 | 0.252 | 28.0 | 150,000 | 156,200 | 5 | 8* |
| 3 | 0.252 | 29.0 | 150,000 | 168,000 | 3 | 22 |

*Broke outside gage length.

**TABLE 7 — TENSILE TEST RESULTS FROM CF8C (18-8) STAINLESS STEEL CASTINGS**

| Site | Test Bar Diameter, in. | 0.2 % Yield Stress, psi | U.T.S., psi | Elong., % | R.A., % |
|---|---|---|---|---|---|
| 1 | 0.160 | 36,500 | 95,650 | 58 | 60 |
| 1 | 0.160 | 36,500 | 96,850 | 58 | 58 |
| 2 | 0.126 | 31,400 | 94,980 | 76 | 62 |
| 2 | 0.126 | 35,800 | 98,500 | 69 | 67 |
| 3 | 0.252 | 29,000 | 81,000 | 58 | 66 |
| 3 | 0.226 | 34,500 | 81,500 | 57 | 61 |

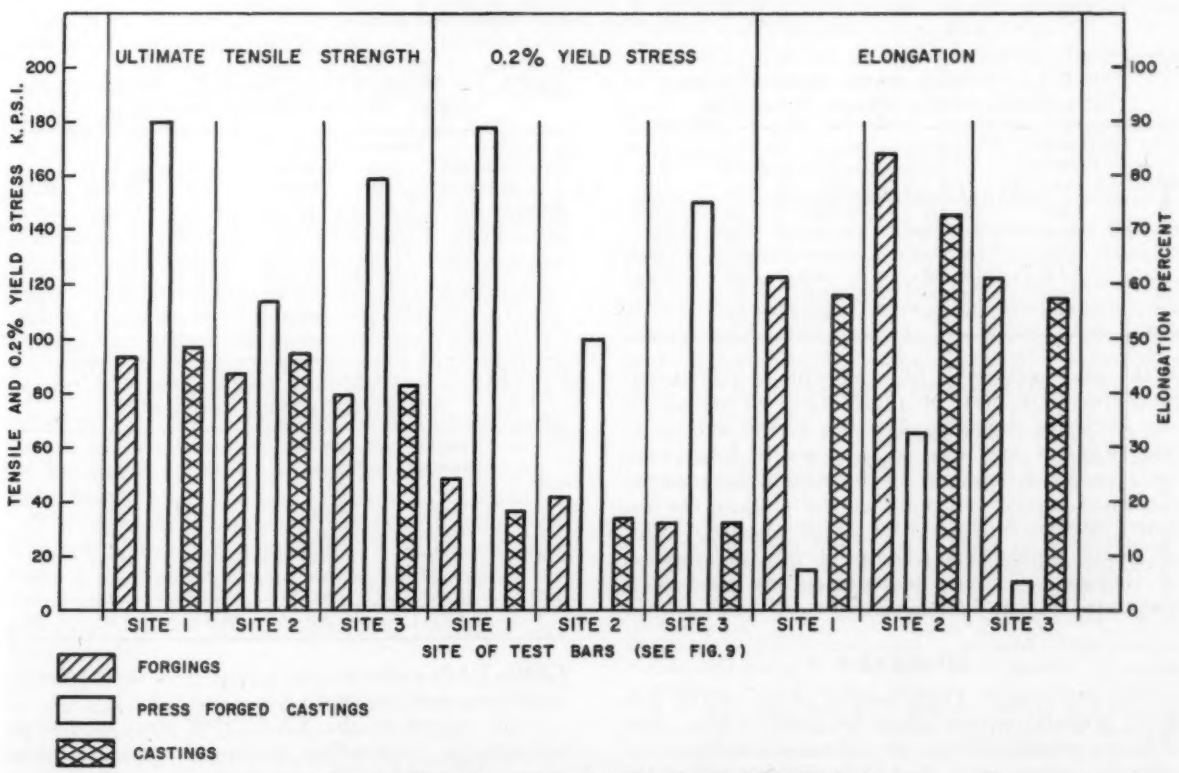



Fig. 10 — Comparison of mechanical properties of test bars from CF8C forgings, press forged castings and castings.

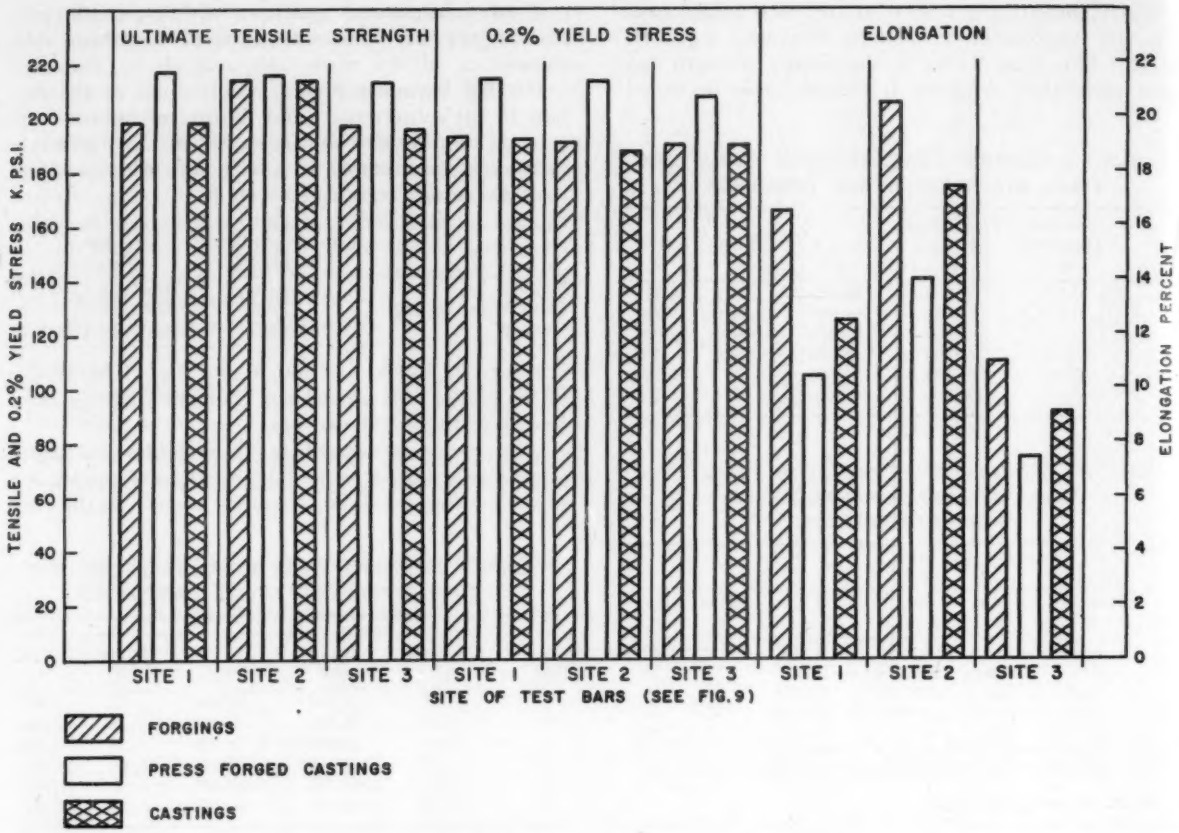



Fig. 11 — Comparison of mechanical properties of test bars 17-4 PH forgings, press forged castings and castings.

TABLE 8 — TENSILE BARS FROM 17-4 PH STAINLESS STEEL PRESS FORGINGS

| Site | Test Bar Diameter, in. | 0.2 % Yield Stress, psi | U.T.S., psi | Elong., % | R.A., % |
|---|---|---|---|---|---|
| 1 | 0.160 | 195,000 | 200,200 | 17 | 51 |
| 1 | 0.160 | 191,000 | 196,500 | 16 | 58 |
| 2 | 0.126 | 190,500 | 215,000 | 21 | 55 |
| 2 | 0.126 | 190,500 | 215,800 | 20 | 55 |
| 3 | 0.252 | 190,000 | 197,400 | 11 | 53 |
| 3 | 0.252 | 190,000 | 198,400 | 11 | 52 |

TABLE 9 — TENSILE BARS FROM 17-4 PH STAINLESS STEEL PRESS FORGED CASTINGS

| Site | Test Bar Diameter, in. | Deformation at-Site, % | 0.2% Yield Stress, psi | U.T.S., psi | Elong., % | R.A., % |
|---|---|---|---|---|---|---|
| 1 | 0.160 | 9.0 | 214,000 | 216,700 | 10 | 32 |
| 1 | 0.160 | 12.4 | 215,000 | 217,200 | 11 | 35 |
| 2 | 0.126 | cracked | 215,000 | 217,700 | 11 | 37 |
| 2 | 0.126 | cracked | 211,000 | 214,100 | 17 | 40 |
| 3 | 0.252 | 10.2 | 207,000 | 214,000 | 7 | 33 |
| 3 | 0.252 | 13.0 | 208,000 | 212,800 | 8 | 35 |

TABLE 10 — TENSILE BARS FROM 17-4 PH STAINLESS STEEL CASTINGS

| Site | Test Bar Diameter, in. | 0.2 % Yield Stress, psi | U.T.S., psi | Elong., % | R.A., % |
|---|---|---|---|---|---|
| 1 | 0.160 | 192,000 | 197,200 | 13 | 36 |
| 1 | 0.160 | 192,000 | 197,200 | 12 | 37 |
| 2 | 0.126 | 188,000 | 213,300 | 16 | 38 |
| 2 | 0.126 | 188,000 | 214,700 | 19 | 40 |
| 3 | 0.252 | 190,000 | 195,000 | 9 | 36 |
| 3 | 0.252 | 190,000 | 196,000 | 9 | 30 |

of the parts—noticeable improvements in mechanical properties have been obtained. Two further points are worthy of comment. The size of the section, in both forgings and castings, appears to have some effect upon the mechanical properties of the material contained therein. In spite of the cracking in the center section of the press forged castings, which effectively limited the amount of true deformation of the section, some improvement in mechanical properties was still forthcoming.

## SUMMARY

The preliminary experimental phase of the program investigated the effects of press forging three different casting alloys, in different conditions, at different temperatures and varying amounts of deformation. The alloys involved were S.A.E. 4340, CF8C (18-8 stainless steel) and 17-4 precipitation hardening stainless steel.[1,2]

With regard to the S.A.E. 4340 steel, a definite though not outstanding increase in physical properties is obtainable. The conditions that appear necessary are principally a large amount of deformation,

greater than 35 per cent, and any of a small number of part and die temperature combinations. The continuous cooling transformation characteristics, under stress, of this steel do not make it particularly amenable to an ausforming type strengthening operation in larger section sizes.

Deformation of low carbon austenites and/or martensites occurring in the CF8C and 17-4 PH stainless steels do indicate considerable promise. Certain reservations have to be made, however, as regards deformation at room temperature of the martensite of the 17-4 PH steel, which is, of course, in the company of delta ferrite. This particular structure or combination of structures appears to be sensitive to minor surface and subcuticular defects or stress raisers during the straining occurring in processing or testing. The elimination of these extremely minor defects—for example, by vacuum melting and casting into high quality molds—must be accomplished before full advantage can be had of the process. A combination of processes involving separate ausforming and martenforming operations would appear to be worthy of investigation.

The results obtained when the CF8C (18-8 stainless) steel was deformed at −100 F were extremely gratifying insofar as mechanical property improvement is concerned. Several percentages of deformation were applied at this temperature and, provided austenite grain size and composition are fixed, a consistent degree of improvement is to be expected. In certain instances when 40 per cent deformation is applied, increases in yield stress and ultimate tensile stress of 400 to 600 per cent and 200 to 250 per cent, respectively, were obtained. The accompanying elongations and reduction of area also appear to be highly favorable. The relative insensitivity of the material to surface defects, stress raisers, etc., must also be noted in this context.

### *Industrial Application*

As regards the industrial application of this process, difficulties can be foreseen with closed die forging of these two materials at the relatively low temperatures involved. These problems are not considered insurmountable. Careful design of the dies—for example, the employment of massive rings around the periphery of the dies—would assist in this aspect. In this research work, forging was done to three distinct areas in the part. Under proper conditions of die design, part configuration and die constraint it should be perfectly feasible to accomplish this in a single operation.

The method of semi-open die press forging techniques is immediately applicable. The selective press forging of highly stressed areas could readily be accomplished upon suitably designed parts. The advantages to be gained from this technique are considerable.

Aside from the property enhancement accomplished by the special forging of casting procedures outlined in this paper, it is also proper to point out that the conventional forging methods and temperatures could be used successfully for final shaping of a precast part. It is conceivable that such methods would simplify forging of complex shapes, particularly of the exotic alloys. Thus shapes impossible to achieve, such as deep and thin flanges, could be first cast and then upset to the correct dimensions.

Another potential advantage of forging a cast blank is that the anisotropy of the forging mechanical properties would not be as great. Therefore, a part subjected to complex stresses, involving transverse as well as longitudinal loads, could perform in a superior manner to a forging made from bar stock, or similar wrought product exhibiting marked directionality of properties even before the forging sequence has begun. Since the transverse properties decrease as the amount of working increases, it is obvious that in many instances the wrought-forged product would be inferior to a cast-forged product. Validation of this was given by Sims,[3] and German use of that method during World War II further substantiates it. The guiding principle in selection of such a cast-forged process should be the complexity of stress distribution.

The fundamental metallurgical aspects of the current work, together with recently reported work[4,5,6] upon straining and subzero rolling of high nickel transformable austenitic alloys, are worthy of further investigation. Presented in a reference[4] are data which show that tensile strengths of 310,000 psi with 10 per cent elongation have been readily obtained. Three phenomena appear to be responsible for these outstanding properties, and may also be valid for the effects observed currently for CF8C and 17-4 PH steels:

1. Precipitation prior to martensite transformation.
2. Formation of "martensite of deformation" as opposed to "martensite of cooling."
3. Effect of deformation of martensite upon subsequent precipitation processes.

Combinations of the above processes must also be encompassed in any future work.

## ACKNOWLEDGMENT

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# IRON—THE PROBLEMATIC FACTOR IN QUALITY OF ALUMINUM ALLOY DIE CASTINGS

by Walter Bonsack


## ABSTRACT

**Iron is an inherent impurity of aluminum and its alloys, and since iron is further carried into the alloys by casting processes and improper remelting its presence increases with the re-use of aluminum alloys. All casting methods, except die casting, try to avoid high iron contents because of its detrimental influence on mechanical and physical properties. Yet the die casting industry specifies minimum iron contents, which are the maximum for or way beyond those used in the other processes.**

**The fact whether these iron contents are actually desirable and necessary for the quality of a die casting is questioned and, if necessary, the application as often used at present is detrimental to good machining, and they certainly do not enhance mechanical properties. Die casting, rightfully, is a proud process of mechanical engineering and science. Unfortunately, the basic science of metal casting has not held step for one reason or another. Discussing just one factor of the basic science of casting (metallurgy), i.e., the phase of an iron content and its influence on the quality of a casting, may give food for thought whether other metallurgical factors when as rigidly controlled as the mechanical part of the die casting process would not advance die casting still further.**


## DIE CASTING STATUS

According to present day statistics, permanent molds and die castings have enjoyed a tremendous increase in consumption. It is shown that in the last 10 years the use of permanent mold castings increased 267 per cent and of die castings 418 per cent, while the sand casting industry was not quite as fortunate, increasing only 133 per cent.[1]

There must be a reason for this percentage of increase. Because we are used to shopping to buy the best for the least amount of dollars, these figures mean, in general, that mass consumption of die casting is the least expensive, permanent mold next and sand casting last, demand being equal. This, however, is a relative statement and an assumption. This alignment of desirability and adaptation is interesting to a metallurgist because the three processes are fundamentally the same. They are all casting processes. However, they result in vastly different products economically and technologically.

There are quite a number of well recognized and acknowledged facts about these three processes. The physical properties of alloys increase with the amount of chill they obtain in the mold or the rapidity in which they solidify.[2]

### *Sand Casting*

On this basis sand castings should give the lowest property, and they do, because the chilling media (sand) has low heat transfer. Heat treating and grain refining has helped sand casting along with proper gating, risering and chilling. The quality sand castings of today are far superior to those castings of yesteryear.[3,4,5,6]

### *Permanent Mold Casting*

The next more effective chilling media is metal, metal molds and dies. Between these two mentioned above is one more step which must not be overlooked. While we readily jump from sand casting to permanent mold casting there is one step between — semi-permanent mold casting. It has an important place in the production of castings and must by no means be overlooked, because many an automotive or other production casting is only possible by use of sand cores in permanent or metal molds. Due respect should be given to them, and when the process is properly executed these castings do approach permanent mold castings.

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They are, of course, only used when the configuration of a casting does not permit a metal core. Since semi-permanent molds demand a rather expensive dispensable sand core (contrary to a rather permanent metal core), design must be directed to avoid the more expensive sand core. Even metallurgically this would be more desirable. This, then, brings the full permanent mold in as the candidate for the highest properties, i.e., basically closer tolerances, less machining, higher mechanical properties and, connected with its higher production rate, more economy.

Again, when these castings are heat treated their properties improve considerably and in many cases approach, equal or even surpass forging or wrought alloy products. At present, a sound, heat treated permanent mold casting produced under metallurgical care is seldom surpassed in mechanical properties by any other casting.

The only limiting factor is size. If its size becomes too large, it reverts to semipermanent mold or even sand casting in property respect. But this, of course, is true for anything in life; anything large becomes problematic.

### *Die Casting*

The most rapid solidification and highest chill is produced in the die casting process. By all proper consideration it should produce the acme of mechanical properties, and it does in quite a few cases. Die casting introduces the metal into the die or mold under pressure while the other two processes depend solely on gravity. The use of pressures, and today they are high pressures, allow detailed reproductions of unprecedented dimensional accuracy. This, in most cases, obliterates excessive machining and its cost.

Due to the fact that high pressures are available today to push the metal into the most difficult and thinnest configurations by mechanical means, the die caster is able to obtain the highest chill on a metal, and with such prospects produce the highest mechanical properties of any casting process, in the author's opinion. However, the die casting industry is not quite at this ideal state. Having a powerful car, it is quite easy to pass the speed limits and to do things which are not within the law. In the case of die casting, this law is the law of nature. It is a bit unfortunate that quite often one can escape man made laws, but laws of nature are stern and uncompromising.

It is the author's intention to discuss some of these laws of nature which seemingly plague the die casting industry. Laws which can be easily obeyed with little effort, as long as they are recognized as physical laws which cannot be changed and must be honored to make every die casting what it has a right to be.

## ALLOYS AND ALLOYING ELEMENTS

There are only a few die casting alloys used today, although this is a limitation imposed by man not by nature. Any alloy that is worth casting can be die cast, if the effort is made to cast it and it is paid for. The American Society for Testing Materials (A.S.T.M.), which is our foremost body of writing standard specifications of engineering materials and methods, records the alloys given in Table 1[7] as the most useful and desirable die casting alloys today.

These compositional limits were established with the aid of producers of the alloys giving serious consideration to consumers and producers of die castings. They, therefore, represent the cross-section of the technological knowledge of die casting. However, since industry is not sterile, attempts are being made to expand this field of alloys, and experimentation and actual use of alloys beyond this field is constantly going on.

To discuss the A.S.T.M. alloys first, there are basically six alloys from which to choose. Tables 2 and 3 give additional information as to their mechanical, physical and other properties or advantages. All these tables have to be fully understood and appreciated. For a member of the committee writing these specifications they are, of course, self-explanatory. But let us take the extreme outside case, the designing engineer, looking for the alloy and its properties he has in mind.

A design engineer would look at G8A alloy as the best alloy, because it gives excellent strength coupled with high ductility, it is corrosion resistant and has a high luster when polished to a silvery color after anodizing. This would be, to him, the ideal alloy for the job. Little does he conceive the foundry factors also mentioned in Table 3, nor their significance as far as foundry effort and price is concerned. So let it be said, generally, that those extra attractive properties mentioned mean extra care, experience and, of course, cost. To go one step further, a styling en-

TABLE 1 — CHEMICAL REQUIREMENTS

| Alloy | Al, % | Cu, % | Fe, max, % | Si, % | Mn, max, % | Mg, % | Zn, max, % | Ni, max, % | Sn, max, % | Other Constituents, Except Al, max, (Total) % |
|---|---|---|---|---|---|---|---|---|---|---|
| G8A | rem | 0.25 max | 1.8 | 0.35 max | 0.35 | 7.5 to 8.5 | 0.15 | 0.15 | 0.15 | 0.25 |
| S5C | rem | 0.6 max | 2.0 | 4.5 to 6.0 | 0.35 | 0.10 max | 0.50 | 0.50 | 0.15 | 0.25 |
| S12A | rem | 0.6 max | 1.3 | 11.0 to 13.0 | 0.35 | 0.10 max | 0.50 | 0.50 | 0.15 | 0.25 |
| S12B | rem | 0.6 max | 2.0 | 11.0 to 13.0 | 0.35 | 0.10 max | 0.50 | 0.50 | 0.15 | 0.25 |
| SC84A | rem | 3.0 to 4.0 | 1.3 | 7.5 to 9.5 | 0.50 | 0.10 max | 1.0 | 0.50 | 0.35 | 0.50 |
| SC84B | rem | 3.0 to 4.0 | 2.0 | 7.5 to 9.5 | 0.50 | 0.10 max | 1.0 | 0.50 | 0.35 | 0.50 |
| SC114A | rem | 3.0 to 4.5 | 1.3 | 10.5 to 12.0 | 0.50 | 0.10 max | 1.0 | 0.50 | 0.35 | 0.50 |
| SG100A | rem | 0.6 max | 1.3 | 9.0 to 10.0 | 0.35 | 0.40 to 0.6 | 0.50 | 0.50 | 0.15 | 0.25 |
| SG100B | rem | 0.6 max | 2.0 | 9.0 to 10.0 | 0.35 | 0.40 to 0.6 | 0.50 | 0.50 | 0.15 | 0.25 |

TABLE 2 — DIE CASTING AND OTHER CHARACTERISTICS

| Alloy | Approximate Melting Range, F | Die Casting Characteristics: Resistance to Hot Cracking | Pressure Tightness | Die Filling Capacity | Other Characteristics: Resistance to Corrosion | Machining | Polishing | Electroplating | Anodizing Appearance | Chemical Oxide Coating (Protection) | Strength at Elevated Temperature | Suitable for Welding | Suitable for Brazing |
|---|---|---|---|---|---|---|---|---|---|---|---|---|---|
| G8A | 995 to 1150 | 5 | 5 | 5 | 1 | 1 | 1 | 5 | 1 | 1 | 4 | No | No |
| S5C | 1065 to 1170 | 2 | 3 | 3 | 2 | 5 | 4 | 2 | 4 | 3 | 5 | No | No |
| S12A | 1065 to 1080 | 1 | 2 | 1 | 3 | 4 | 5 | 3 | 5 | 3 | 3 | No | No |
| S12B | 1065 to 1080 | 1 | 2 | 1 | 3 | 4 | 5 | 3 | 5 | 3 | 3 | No | No |
| SC84A | 1000 to 1100 | 2 | 2 | 2 | 5 | 3 | 3 | 1 | 3 | 5 | 2 | No | No |
| SC84B | 1000 to 1100 | 2 | 2 | 2 | 5 | 3 | 3 | 1 | 3 | 5 | 2 | No | No |
| SC114A | 960 to 1080 | 2 | 2 | 1 | 5 | 3 | 3 | 2 | 4 | 4 | 2 | No | No |
| SG100A | 1035 to 1105 | 1 | 1 | 1 | 3 | 3 | 3 | 1 | 3 | 3 | 2 | No | No |
| SG100B | 1035 to 1105 | 1 | 1 | 1 | 3 | 3 | 3 | 1 | 3 | 3 | 1 | No | No |

TABLE 3 — TYPICAL MECHANICAL PROPERTIES OF TEST SPECIMENS

| Alloy | Tensile Strength, psi | Yield Strength (0.2 per cent Offset), psi | Elongation in 2 in., per cent | Shear Strength, psi | Fatigue Strength (R. R. Moore Specimen), 500,000,000 Cycles, psi |
|---|---|---|---|---|---|
| G8A | — | — | — | — | — |
| S5C | 33,000 | 14,000 | 9.0 | 19,000 | 17,000 |
| S12A | 42,000 | 19,000 | 3.5 | 25,000 | 19,000 |
| S12B | 43,000 | 21,000 | 2.5 | 25,000 | 19,000 |
| SC84A | 47,000 | 23,000 | 3.5 | 27,000 | 20,000 |
| SC84B | 46,000 | 23,000 | 2.5 | 28,000 | 20,000 |
| SC114A | 48,000 | 24,000 | 2.5 | 29,000 | 20,000 |
| SG100A | 46,000 | 24,000 | 3.5 | 26,000 | 18,000 |
| SG100B | 44,000 | 25,000 | 2.5 | 28,000 | 20,000 |

gineer may be quite dissatisfied with the high luster of this alloy, because he may expect nothing less than a chrome plate luster.

### *Aluminum Alloys*

This then will eliminate, at present, the high magnesium alloys from the all around alloys. This leaves the aluminum-silicon, aluminum-silicon-magnesium and aluminum-copper-silicon alloys as desirable alloys production wise. Straight aluminum-silicon alloys are the 13 per cent silicon alloy, known as 13 alloy, or shown in Table 1 as S12 alloy. It is used primarily where corrosion resistance is of greatest importance. It is not as corrosion resistant or easily machined as G8 alloy (218) but it will cast much easier, i.e., fills thinner sections and does not cause cracking because this alloy is not as hotshort as the G8 alloy.

Tables 2 and 3 give all the pertinent property factors, and from them the advantages and disadvantages will be learned. This alloy, because of the property factors and cost, is only used where its application is called for. To provide higher mechanical properties, alloy 360 or SG100A is provided. Its 0.5 per cent magnesium content increases yield and tensile strength, but reduces elongation and impact strength. Magnesium also improves the machineability of the alloy without reducing its inherent corrosion resistance. In order to obtain all these advantages, the magnesium content has to be controlled between the limits shown in the specification. This adds to the effort of the produced casting.

## MOST PROMINENT ALLOYS

The remaining alloys shown as SC84, or known as 380 or S9 alloy, are the most commonly used alloys. Four per cent copper and 8 per cent silicon is their basic composition. This combination produces an alloy which not only casts well, but also has a combination of all around properties (except corrosion resistance and its allied fields like anodizing, etc.) which makes them the universal alloy of today. There are two different specifications for the same alloy. One is used in the "gooseneck" process of die casting and admits higher iron limits (and lower properties), and the other demanding lower iron for the high pressure cold chamber process of die casting.

This alloy, being the most generally used alloy, should be minutely examined for its virtues, its shortcomings, its uses and abuses. Since this alloy has few close relations, SC64 (319) and SC114 (384), these two will also enter the picture because they will help to explain the metallurgical framework of this discussion. There is another interesting aspect to this overall picture, and since we are directly discussing permanent mold and die casting, the much discussed S.A.E.331, 333 and D132 alloys must by nature be brought into the frame of this picture.

TABLE 4 — DIE CASTING ALLOYS

| | Composition, % | | | | | |
|---|---|---|---|---|---|---|
| | Cu | Fe | Si | Mn | Mg | Zn |
| SC64 (319) | 3.5 | 1.0 | 6.5 | 0.8 | 0.5 | 1.0 |
| SC84 (380,S9) | 3.5 | 1.3 | 8.5 | 0.50 | 0.10 | 1.0 |
| SC114 (384) | 3.8 | 1.3 | 12.0 | 0.50 | 0.10 | 1.0 |
| SC103A (D132) | 3.5 | 1.3 | 9.0 | 0.50 | 0.8 | 1.0 |
| 333 S.A.E.331 | 3.8 | 1.2 | 9.0 | 0.8 | 0.5 | 1.0 |

Table 4 repeats the die casting alloys, but brings also SC64 and the 333 or S.A.E.331 alloy to show their basic relation. Figure 1 shows the same, graphically, in a ternary, three dimensional manner. It is quite plain in either picture that they are closely

related; the only difference being the silicon, magnesium and, iron content. In all of them the copper content is practically the same.

It then seems to convey that a relatively large spread of copper produces the same castings and mechanical properties throughout. This, however, is not quite true, because 3 per cent copper produces less of a ternary eutectic than 4.5 per cent copper. Therefore, freezing and feeding characteristics will be different. These may not be as noticeable in die casting as in permanent mold casting, because die casting has the tremendous pressure behind every alloy while permanent mold casting depends solely on nature's feeding characteristics (gravity and capillary suction). Also in die casting, large variations in ternary or even cinquinary eutectics may cause varying results in the quality of the castings. But an evaluation of this kind is still in the future.

### *Silicon Content*

Silicon, as stated in the specification, varies with the alloys under discussion from 5-13 per cent. S5 alloy, 43 or 95/5 has no copper content, neither has 360 (SG100A) or 13 alloy (S13). However, in all of them silicon assumes an important role, and this is the role of castability and fluidity. From 5-13 per cent these two properties increase rapidly. Therefore, the more complex the casting, the higher a silicon content is of advantage. This has to be recognized and well understood. A copper content may add to these properties at the expense of ductility. Magnesium contents do not add to castability but to mechanical properties. Therefore, in this respect the alloys are well presented and defined.

The other alloying elements are of a much minor role than attributed to them generally. They are called usually impurities and often rightly so, because they are carried in gradually by unwise use of scrap. The influence of most of these elements like zinc, iron, manganese, nickel, chromium, tin, lead, and so on, have been studied, discussed.[8] They have been limited to their harmless status in the A.S.T.M. specifications with the exception of a few, i.e., iron, magnesium, chromium and their relatives in the die casting specifications.

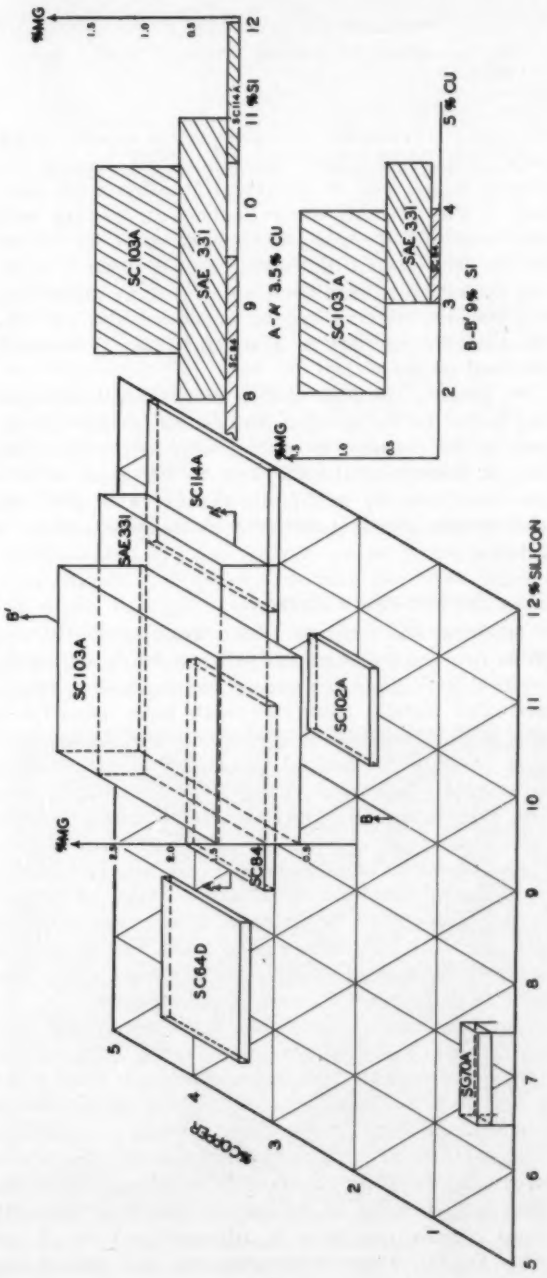
## IRON — A DESIRED UNWANTED ELEMENT

In 1943, in the author's papers cited above, it was plainly stated that, generally, the only really undesirable element in almost all aluminum alloys is iron. Silicon in magnesium alloys is mostly undesired, and copper in corrosion resistant alloys, but in all alloys iron is not especially wanted. The present trend of super high purity alloys confirms this statement made about 20 years ago. The absence of iron allows fuller use of other alloying elements to obtain much higher properties.

However, in die casting we have apparently run into a dead end street in which it seems hard to turn around. All die casting alloys not only permit high iron contents, but demand them, and the producers of super high purity low iron aluminum alloys become accommodating to advocate a high iron minimum specification for die casting alloys. This, of course, is nothing new, since most die casters had, and have had for many years, a top and bottom limit of iron on their alloys when they bought them for their suppliers.

This status is the most curious one, because on the one hand all effort is expended to lower the iron content in aluminum alloys for all other methods of casting except for die casting. In all other methods

**Fig. 1 — Die casting alloys, including SC64 and 333 (S.A.E. 331) alloys, as shown in Table 4.**



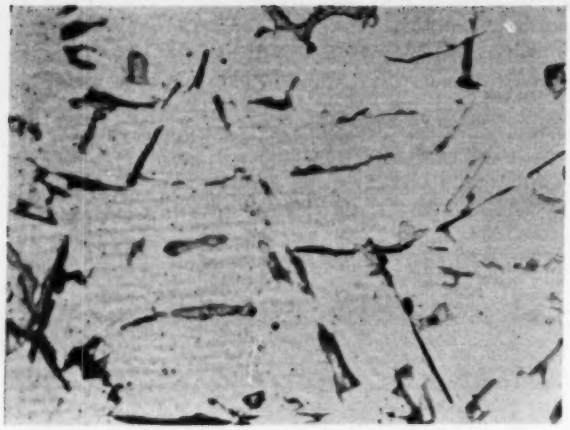


Fig. 2 — AlCuFeSi needles (black). $H_2SO_4$ etch. 400 ×.

of casting it is claimed and proved that superior properties can be obtained, but not in die casting. Die casting needs iron to prevent the alloys from sticking to the die. All other methods of casting must take care to prevent alloys from sticking to the molds; why cannot die casting do the same? It is being done,[1] and it has been stated that by oxide coating the die, sticking can be prevented and any aluminum alloy can be cast advantageously. This sounds like real progress.

At present, the iron content and its misunderstanding is one of the greatest hindrances in the full success of die castings. Iron may help to prevent sticking of castings to the die, but, at the same time, it interferes severely in a costly manner with good machineability which is inherent in aluminum alloys in general.

### *Poor Machinability Reasons*

Studying die castings which machine badly, one finds two main reasons for this condition: excess porosity due to improper gating, venting and injection, and also usually connected with large particles of precipitated iron-manganese-silicon-aluminum complex compounds. Now the geometry of a die design is a special field and changes from casting to casting, but the metallurgy of the alloys poured into the die is universal and can be discussed with finality. This is because metallurgy is physics in the broadest sense and, therefore, governed by laws of nature. These laws cannot be changed as we can change the gate on a casting when we may want to, but must be learned, recognized as such and followed to the outmost to make as perfect a casting as possible.

In the case of iron it must be realized that it is never present in aluminum as such, but in the form of simple or complex compounds. First a few which may be present in die casting alloys will be mentioned. Some die casters cast pure aluminum, and in order to prevent hotshortness the silicon content must be lower than the iron content. Here this iron is present as $Al_3Fe$ (Fig. 2) which is in needle form crystals, but there is still another form of needles—$Fe_2Al_7$. When other elements like silicon, copper, manganese, chromium are present, and the combinations of the composition and temperature time and speed of solidification are right, many compounds can be formed.

Mandolfo,[9] in his book of 1943, lists 18 different aluminum-iron complex compounds which are found in complex aluminum alloys. Just to talk about die casting alloys aluminum-silicon alloy with iron may contain $\alpha$ (AlFeSi), $FeSiAl_6$, $\delta$ (AlFeSi), depending on the equilibrium conditions which in die casting depends, of course, on temperature, and as time is long or short (this phase will be explained later on) one never is too sure what will be found in the structure of the casting.

When copper and manganese are present in an aluminum-silicon alloy, and in alloys like 319, 380, 384 and the like, then we must add $\alpha$ (AlMnSi), $\delta$ (AlMnSi), AlCuFeSi, $\alpha$ (AlFeMnSi) and $\delta$ (AlFeMnSi) as possibilities. If we add chromium to this list, it may even get more complicated. This is not the end, however. There are different crystal shapes and sizes of the same compound, depending under which conditions they are formed. And the form they appear in the castings may make them desirable or disastrous. It should not be surprising that nature will do this to us, because we all know that moisture in the air, for example, can precipitate when cooled "properly" as a gentle snowflake or, improperly, as a destructive hailstone. The size of this hailstone depends on how long it has been kept in suspension in the clouds and allowed to grow.

### *Iron and Manganese Content*

Coming out of the clouds and into the furnace, we have similar conditions to consider. In the aluminum-silicon-copper alloy we are discussing, we have also iron as desired and manganese, as shown in the specifications. We then have to contend with the aluminum - iron - manganese - silicon complex compounds. These precipitate in various shapes, depending on the pouring temperature and the time they have available to grow into the shape the physical laws tell them to assume.

In sand casting and permanent mold alloys, they usually assume the shape of a snowflake, known as Chinese script (Fig. 3), because these alloys are usually cast above 1250 F (677 C), preferably 1350 F (788 C) and even higher. When poured from these temperatures into a cold sand mold or a fast chilling permanent mold, they become the snowflake like, desirable, strengthening, easy to machine crystals of the casting. However, die castings are most often cast at temperatures of 1160 F (682 C) plus, minus or mostly minus (This is an indicting statement which must be justified). At these temperatures, the iron complex compounds grow into hailstones or hardspots.

How hard this condition is to prevent can best be shown when large ingots of die casting alloys of the above composition are microscopically analyzed. When the ingot molds are cold, all complex iron will crystalize as snowflake like crystals. As the molds get hotter and the alloy cools slower, these feathery crystals will grow into a fatter hailstone shape, and the growing aluminum crystals will push them ahead

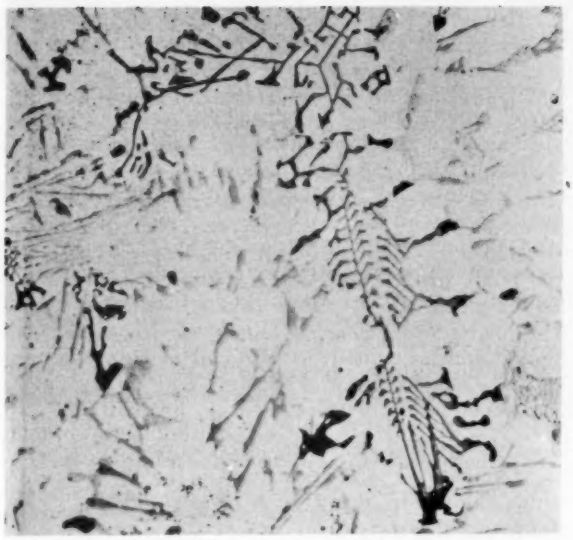


Fig. 3 — AlFeMnSi as feathery Chinese script. $H_2SO_4$ etch. 400 ×.

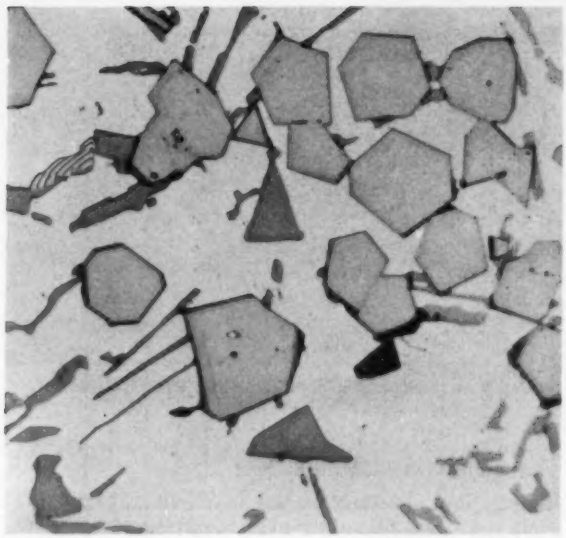


Fig. 4 — Same as Fig. 3, but grown into hardspots by segregation in slowly cooled ingot. $H_2SO_4$ etch. 400 ×.

of their growth, like snow shovels push snow, and concentrate them in the center or shrink area of the ingot. They may grow to such a size that they may be picked up by the naked eye on a machined ingot slice (Fig. 4).

This indicates how fast these crystals grow into this shape and size, and that this shape must be their equilibrium form in this temperature range. This too then ardently insinuates that temperature is more important than time—temperature the major factor and time, the minor.

Most die castings with machining difficulties show this formation of hailstone compact, complex compound which under an uncontrolled condition can take on "devilish" proportions (Fig. 5). These hardspots were found in a poorly machining die casting. Figure 6 shows a colony of the complex compound crystals at much lower magnification, but the size is still enormous. Since Table 2 states that all these alloys are liquid above 1100 F (593 C), why then should this hailstone, the hardspot, be formed because die castings are poured at controlled temperatures of 1160 F (682 C). There are many reasons.

## CAUSES OF DIFFICULTIES WITH IRON

To start with, one must discuss temperatures and temperature control. The latter will be discussed first, because it is the easiest. Temperature is mostly controlled by thermocouples and pyrometers. These relatively thin wires of the thermocouple are easily eaten up by the aluminum-silicon alloys, therefore, they must be protected by "protection tubes." Temperature changes must be transmitted fast to the control instruments. Therefore, the protection tube must have high heat conductivity. Metal tubes have the highest conductivity, then follows graphite, then silicon carbide.

However, here is trouble again. Metal dissolves in aluminum fast unless protected by nonmetallic coatings, which is the only way to do. Watching these and renewing the coating is a must. Graphite oxidizes and disappears, silicon carbide is fragile and breaks. Therefore, the "next best solution" is to provide a heavy coating which is not attacked by aluminum alloys and does not need to be replaced daily. Little is it recognized that these protection tubes will control the temperature at 10:00 a.m. to 2:00 p.m. or, in other words, fool the alloy to the detriment of the quality of the casting. This certainly is false economy. Effective temperature control means control every minute.

What happens when controlled temperature is uncontrolled? Only one answer can be given; anything can happen and usually the least desirable happens. In many, many die castings this is the growth of the iron complex compounds into the hailstone form, the hardspot. Even low frequency electric furnace cannot prevent it.

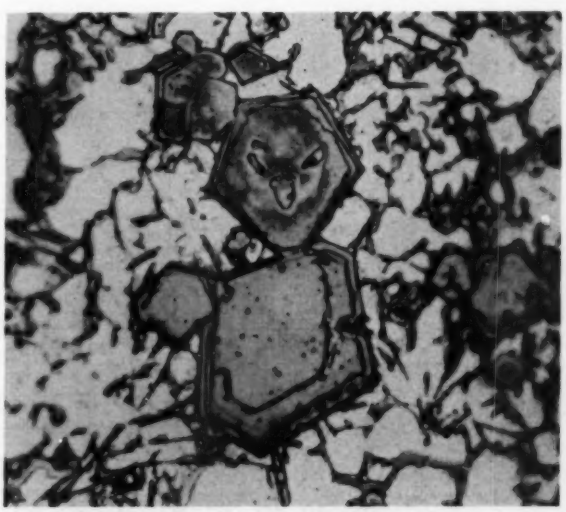


Fig. 5 — The trouble maker. Hailstone (hardspot) complex compound. Keller's etch. 400 ×.

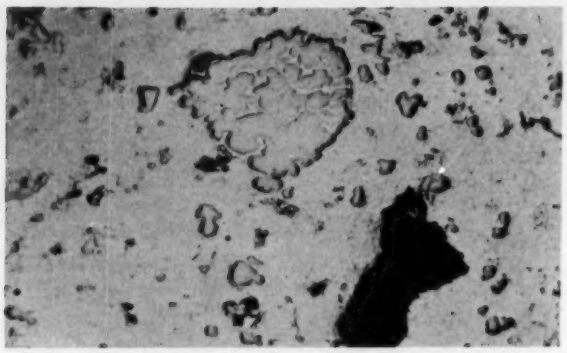


Fig. 6 — Hardspots in a die casting. Unetched. 75 ×.

### *Aluminum Alloy Solidification*

Spear and Gardner in an excellent experimental study on "Solidification of Aluminum Casting Alloys"[10] bring, among other alloys, the study on the solidification of 319 alloy (SC64). Since this alloy is a brother in constitution to SC84 and SC114, the findings on this alloy will illustrate adequately what happens when the latter two alloys are solidified in the manner SC64 was treated in their experiments. It also fits nicely into the picture of this discussion that their experimental procedure (quite incidentally) simulates what happens in die casting, except for spraying of semi-liquid metal as in die casting instead of casting liquid metal and for a starting temperature of about 1100 F (593 C).

They melted the alloy at 1400 F (760 C), and cooled it slowly to various temperatures. These temperatures were pertinent ones because before or after these temperatures were reached, certain elements or compounds had to be formed and be present in solid form because the freezing (solidification) curve of the alloy indicated a crystallization. At these points the alloys were quenched in water which froze the metal as quick as a die casting would solidify. The diagram (taken from this paper) Fig. 7, shows in cooling this alloy, various steps occur.

At 1118 F (603 C), a break occurs and aluminum dendrites had crystallized. The second break, just above 1100 F (593 C), produced an aluminum-iron-silicon complex compound shown in Chinese script or snowflake form. As mentioned before, this shape of the compound is good and desirable, and it is mostly found in this form in permanent mold and sand castings which are poured at high temperatures and which usually have a somewhat lower iron content compared to die castings.

### *Metal Temperature Effect*

Most die castings are said to be poured at around 1140-1180 F (616-637 C). In many instances cold or hardly warm metal is added to the metal at frequent intervals in the form of ingots or scrap returns, causing fluctuations of temperatures to less than the liquidus, shown in the Fig. 7 as 1100 F (593 C). To dramatize the menace of low temperature pouring, Fig. 8 illustrates the heat content of aluminum at its melting point. It takes only 280 Btu's to bring one lb of aluminum up to the melting temperature, but it takes another 170 Btu's to put it into a liquid metal. This same relation is true for die casting alloys (Fig. 9).[11]

It shows that 170-175 Btu's are necessary to change an alloy of this type from solid to liquid. If the casting temperature is close to the solidification or melting temperature, then any addition of solid metal will change the pouring condition without changing the temperatures indicated by controls. In the case of aluminum, shown in the Fig. 8, about 60 lb of aluminum could be added to 100 lb molten aluminum without any change in temperature, except that the state of the metal changes from liquid to mushy to solid.

In this long span of immobility of temperature, compounds of higher melting temperature grow unrestricted in dimensions. Certainly in die casting, with all its pushing power, it does not make too much

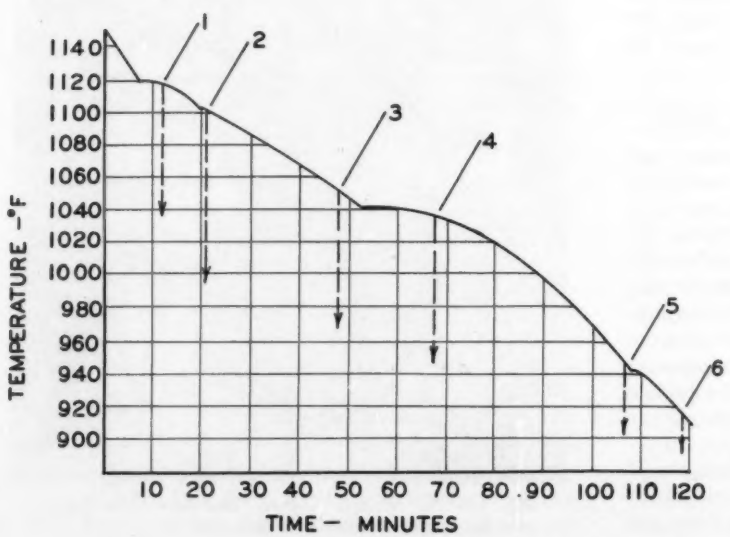



Fig. 7 — Cooling curve from Spear and Gardner.[7]

difference how mushy the metal is. If it is forced into the proper place it will give a desired shape reproduction, accurate but disappointing in properties. Disappointment because, a) the hardspots are formed and well fed and b) entrapped air and die lube gasses cannot escape the near solid mush of metal. Machining will be the problem besides poor properties.

If die castings are made at a temperature at which only 300 Btu's/lb are present instead of 450 or 500, less heat has to be extracted than in a permanent mold and sand casting, and the die will stand up longer because the heat shock is not as high and more shots can be made per hour. But what does this saving mean in terms of quality castings? High scrap figures, high rejection, low tool life and ill will and distrust of consumers is more expensive than lower die life and less shots per hour.

Many a die casting is halfway between casting and hot pressing and, therefore, neither fish nor fowl especially with all the debris mixed in.

### *Temperature and Iron Content Effect*

Going back to the effect of temperatures and iron content, from the microscopic investigations of most die castings in trouble, the actual furnace (conditioned) temperature hovers around 1060-1080 F (571-582 C). In this range, the iron complex compound feeds freely on the fat of the land and grows into larger and larger particles. When they become bigger, like hailstones, they settle out of circulation and drop to the bottom of the furnace forming the sludge.[12] When agitated in the least, they float up like sugar will in cold water.

Low frequency induction furnaces with their constant agitation will not change this condition, but will prevent their growing fat so they become smaller but plentiful. Nature's laws cannot be changed. The ladle will pick the metal up and it goes into the casting. This ladle was cold too, bringing temperatures still lower. We have most probably reached point 3 or 4 of the temperature curve, because the plunger well will cool it still more. The metal is pushed into the die by force and solidifies instantly because of the cold die. Instead of finding the iron complex compound in these nice feathery crystals, they have become the ruinous hardspots.

The cooling curve diagram gives the time in minutes. The good feathery iron compound will be a bulky hardspot after 2 hr at this temperature (Fig. 6). The fact that SC84 has 2 per cent more silicon and SC114 5 per cent more silicon than SC64, aggravates this situation so much more. Because, as shown in the constitution diagram (Fig. 10) of aluminum-silicon-copper alloys, with the increase of silicon the solidification temperature is lowered and the temptation of lower pouring temperature increases to the detriment of the quality of the casting. However, the solidification temperature of the iron complex remains the same, and difference in temperature gives it more chances to grow bigger.

In Fig. 4, the same compact hardspots were shown as segregation in ingot material. Their presence was explained as slow cooling, or a relatively long time at low temperature just above the solidification point.

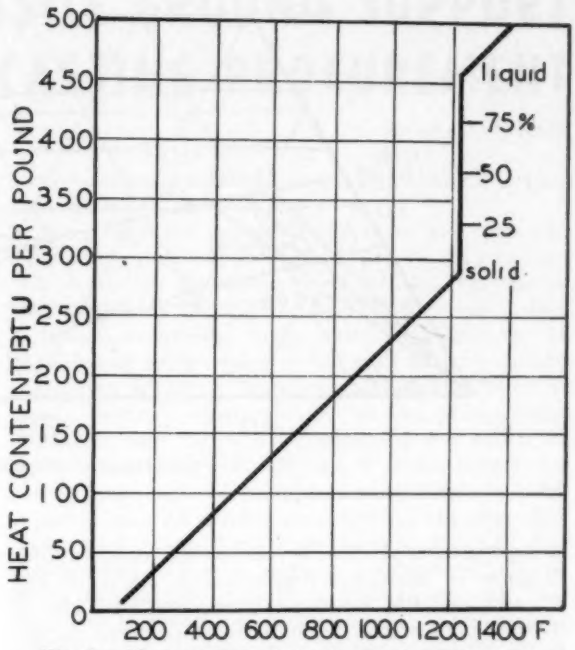



**Fig. 8 — Heat content of aluminum at various temperatures.**

If ingots are fed into a furnace from which die castings are made, and which are operated at low temperatures to begin with, these iron crystals will not dissolve, but settle out and form nuclei for more precipitation. This is equally true for die casting scrap returned to the furnace at the machine.

At these low temperatures they cannot dissolve because the solution at those temperatures, as shown,

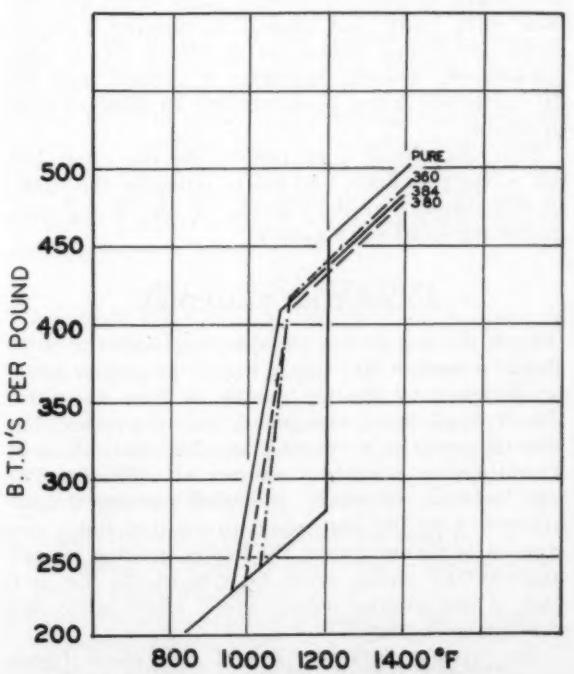



**Fig. 9— Heat content of aluminum die casting alloys.**

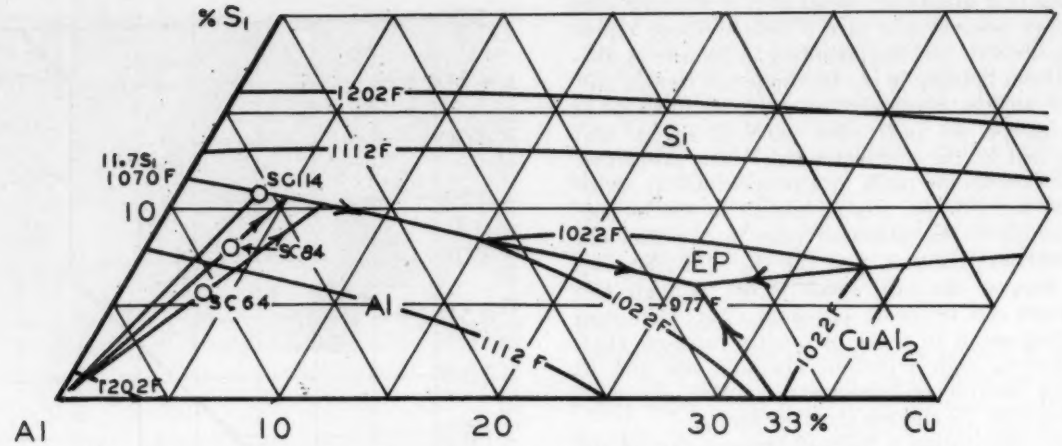



**Fig. 10 — Constitutional diagram of aluminum-silicon-copper alloys.**

is super-saturated. The corner of the constitutional diagram of aluminum-copper-silicon alloys in Fig. 10, and also the melting ranges given in Table 2, indicate plainly that the liquidus temperature changes down to 1070 F (577 C) or less but the solidification temperature of the iron complex compound is not taken into consideration. This is because at these temperatures it has already crystallized to quite some extent, and is present like undissolved sugar in ice tea.

Undissolved sugar in ice tea will still taste sweet in the tea drinker's mouth, because the body warm saliva will dissolve it. However, the low die temperature can never dissolve the high melting iron complex compact crystals. Their function as an useful alloying element becomes problematic. The only explanation might be that the already pasty metal becomes even more pasty, and instead of pouring a liquid metal which may attack an uncoated die, a high temperature concrete aggregate is pressed into the die cavity which has no power left anymore to weld to the die.

This, then, is of advantage to the die caster but not to the consumer, who has to pulverize the stones in this concrete at the expense of his cutting tool, machining speed and accuracy.

## SUGGESTED REMEDIES

Since this is a serious problem, and added to it we should consider that much higher properties could be obtained in the die casting if these inert, extremely hard, brittle compounds were not present but were dispersed in a desirable manner, there must be a solution or possibly a number of solutions. The only solution, of course, is higher pouring temperatures to keep the compounds in solution and a new concept of taking care of the higher pouring temperature in the casting cycle. Coating of the die, it is said, is one solution which would allow lower iron contents to be poured.

Die coating, however, should also allow higher pouring temperatures which will prevent the formation of iron complex hardspots and retain them as feathery snowflakes, which add to the properties and do not interfere with machining. Proper balance of pouring and die temperatures to produce this superstrong casting is possible, as it is in permanent mold casting. It may not be as economical, but is it economical to produce many castings fast and lose many of them because of their defects and their difficulty in machining? Certainly there is enough ingenuity in producing difficult machines and dies; there must be equal metallurgical ingenuity to find the proper die condition and heat balance between metal and die to produce the casting that the die casting is destined to be.

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# MISSILE GROUND SUPPORT EQUIPMENT CASTING PROCUREMENT

by Loren W. Smith


## ABSTRACT

Attention is focused on the nonairborne steel castings such as are found in the missile ground support field. The tendency by aircraft designers in the missile field to design and specify airborne quality for ground type castings needs to be reviewed in light of the foundry hardships to meet aircraft quality and the high costs involved in doing so for applications where lower quality standards will suffice. Pertinent factors involved the procurement of nonairborne steel castings are presented, which may influence aircraft casting engineers who are concerned with the design of ground support type of casting applications.

Orientation of casting advantages are presented. Factors influencing choice of molding processes together with cost of molding are noted. A chart of general comparison of casting methods is shown. These include sand, $CO_2$, shell, ceramic and investment castings. The important factor in the production and procurement of a steel casting which is covered by the term, quality, is discussed. Understanding and mutual agreement on what constitutes quality is stressed. The general design criteria for nonairborne applications should not be the same as airborne castings. The use of high integrity, close tolerance steel castings, heat treated to strength levels of 180,000 psi or more, which require little or no machining is not indicated for the ground support equipment type application.

A proposed specification for ground application castings, drafted by the Army Rocket and Guided Missile Agency at the Redstone Arsenal, is viewed as the direction that most steel foundries believe ground support castings should be directed toward with respect to procurement specifications. Specification requirements such as surface defects, magnetic particle defects, radiographic inspection are noted with their effect on interpretation and costs of inspection. An illustration of how the cost factor is affected when magnetic particle and x-ray inspection standards are specified is shown.


## INTRODUCTION

Over the past decade, considerable effort has been expended by both the aircraft and steel casting industries toward satisfactory ferrous castings for aircraft use. The desired objective of having steel castings engineered into aircraft frame design as structural components has not been realized to the extent that was thought possible. The production and application problems involved in both industries for aircraft quality castings have proved to be complex, and with the apparent decline in airplane manufacture the need for such steel castings will probably decrease. The advent of missiles and their structural requirements will continue the endeavor to utilize the advantages that cast-to-shape metal processing offers, but emphasis will be in the refractory metal field with some possible applications in the high strength ferrous casting area.

Along with the missile development has come the need for ground support equipment (G.S.E.) to provide means for the launching of missiles. The use of steel castings in this type of design, where applicable, is good economics, hence good engineering. If realistic end use prospective and steel foundry casting limitations are properly incorporated in the design of G.S.E. castings, there should be no procurement difficulties such as have accompanied the efforts to provide aircraft quality castings. It is the purpose of this paper to provide orientation to this end, in order to avoid aircraft quality complexities entering into design and procurement requirements of G.S.E. missile castings. Ground support equipment ("Kiwi bird") castings are difficult to consistently cast and expensive to inspect to aircraft quality specifications.

Since G.S.E. castings are "grounded," they are a type that the steel foundry industry has been producing for years with outstanding service records. From this type of experience, the foundryman is in a good position to urge the casting designer to use design criteria and specify quality requirements that will suffice for the intended service. Needless critical load design coupled with unnecessary casting quality can lead to trying or almost incompatible relationships for those people who are directly involved in the procurement and supply of steel castings.

It is disheartening to the foundry engineer to learn that a casting over which he has sweated to produce the highest possible quality and surface finish as specified by an aircraft quality casting specification is merely used to carry a low static loading on some ground application and be well covered over with several layers of paint. It is somewhat similar to the feeling of frustration that an aircraft engineer would have who, after achieving the design and manufacture of a new fighter aircraft, finds out upon delivery to the military customer that it is used for airport ground training for new pilots.

## CASTING ORIENTATION

A general knowledge of casting metal to shape is basic for the design engineer. A listing of the advantages that castings have over other forms of metal fabrication, and the general classification of molding processes involved in casting steel to shape, are available. The foundry industry notes at least six advantages that the casting process has over other forms of metal fabrication. These are:

1. Design freedom – metal where you want it.
2. Reduction of stress concentrations – no sharp corners or residual stresses.
3. Low end cost – quality product at lowest cost – eliminates indirect labor.
4. Uniform strength – multi-directional properties.

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GENERAL COMPARISON OF CASTING METHODS

| Casting Requirements | Sand Castings | $CO_2$ Castings | Shell Castings | Ceramic Castings | Castings Investment |
|---|---|---|---|---|---|
| Micro-inch Surface smoothness | 200-500 | 150-400 | 100-200 | 80-150 | 40-125 |
| Thinness of metal section, in. | 3/8-5/16 | 3/8-5/16 | Good | Very good | Excellent |
| Base tolerances, in./in. | ± 0.030 | ± 0.020 | ± 0.008 | ± 0.006 | ± 0.005 |
| Added tolerance across parting or shift, in./in. | ± 0.020 | ± 0.020 | ± 0.010 | ± 0.010 | No Parting |
| Intricacy | Fair | Good | Very good | Extra good | Excellent |
| Metal limitations | None | None | Some low carbon steel | None | None |
| Finish allowances | Most | Most | Average | Least | Least |
| Size adaptability | No Limit | Slight Limit | Small, up to 100 lb | Small, up to 100 lb | small, up to 40 lb |
| Pattern costs | Low | Low | High | Average | High |
| Lead time | Shortest | Shortest | Long | Short | Longest |
| Prototype rating | Not applicable | Not applicable | Generally not applicable | Low cost | High cost |

5. Good appearance – streamlining possible.
6. Choice of properties – metals developed to resist corrosion, heat shock, abrasion, creep, alloys that cannot be wrought.

Considerable space could be used in discussing the details and citing examples to show how the casting process is of great economic importance to its users, but this can be gained by reading the metal casting literature. A recommended reference for this purpose is the *Steel Castings Handbook**. It also contains a glossary of foundry terms, engineering tables and steel casting specifications.

When reference is made to metal casting one normally thinks in terms of producing castings in sand, as practiced by foundries since the beginning of the casting art. However, casting metal to shape is an art which is rapidly becoming a science. Modern technology has come into the foundry and considerable progress has been made. New molding methods, employing sands or similar refractory materials appropriately bonded to form the mold and cores, are production proved and offer specialized improvements in the casting. Various factors influence the choice of these molding processes.

Design, weight, length, width, thickness, height, shape and contour will determine the best choice of the casting process to use. Thin castings will demand investment, special ceramic or shell in preference to sand or $CO_2$ in order of their listing with respect to increasing metal section. On the other hand, the upper limitations of the weight, length, width and height throw a definite restriction on the investment or shell process. The special ceramic process, other than investment, offers a happy medium but with restriction. Tolerance and surface finish obtainable with various processes are in decreasing order – investment casting, special ceramic, shell molding, $CO_2$ process and sand casting.

Costs involved in the molding processes are also in decreasing order as shown. Sand casting is generally the lowest. However, overall costs in the use of the casting is the method to evaluate the choice of the molding process to be used.

A detailed comparison of steel and alloy casting methods with their typical characteristics is shown in the table.

## QUALITY CONSIDERATIONS

Probably the most pertinent factor involved in the procurement of a steel casting is the term quality. This term can cover a wide range of interpretations and subjective standards, almost up to the point of wishful thinking. While in the past, this term may have been the basis for more misunderstandings and mistrust between the steel foundry and its customers than any other factor, it has been satisfactorily resolved for many industries. The cast valve and transportation industries are good examples. Complete and mutual understanding has been reached on what is quality for these industries in regard to steel castings.

The aircraft industry has been trying to establish what can be classified and delivered as aircraft quality for steel castings. Considerable progress has been made in this direction, as witnessed by the growing number of parts produced as aircraft castings. While these are expensive (in some cases up to $30.00/lb), it does show aircraft quality can be realized. Objective standards in the form of MIL specifications are now noted on aircraft casting prints. Some of these standards are so high in quality requirements that most steel foundries refuse to consider even bidding on the inquiries.

The initiation of an understanding on what quality requirements are desired and practical for castings to be applied to G.S.E. missile use is opportune, so that procurement practices can become standard for this growing industry. Standardization will allow desired quality along with lower costs and promote production scheduling.

Design load factors can readily be incorporated into G.S.E. castings for missile support applications, that will allow maximum dependability in service and at the same time be compatible with the quality

*Steel Founders' Society of America, 3rd Edition (1960).

of the individual steel casting that a first class foundry can consistently produce and deliver on an agreed upon delivery schedule. The use of high integrity, close dimensional tolerance steel castings, heat treated to strength levels of 180,000 psi or more, is not indicated to be required for most G.S.E. casting applications. Requirements for special alloys with resistance to corrosion, erosion and thermal shock and high hot strength are not apparent.

The absence of such specialized material requirements removes the need for close dimensional castings to alleviate machining difficulties and costs. Well proved constructional steel casting alloys can be specified to useful strength levels, that will also give satisfactory ductility and toughness properties in addition to being readily weldable with no loss in physical properties. A proposed specification for the procurement of high impact strength low alloy steel castings, drafted by the Army Rocket and Guided Missile Agency at the Redstone Arsenal, is in the direction in which the steel foundries feel that the missile industry should specify the G.S.E. casting. This specification calls out three classes of steel castings based on a minimum ultimate strength and impact properties. These are:

| Class | Tensile Strength, psi | Impact Resistance Charpy Keyhole, ft-lb |
|---|---|---|
| Class 1 | 105,000 to 120,000 | 15 |
| Class 2 | 125,000 to 150,000 | 12 |
| Class 3 | 150,000 to 180,000 | 10 |

The chemical composition of the alloy to provide the above mechanical properties after suitable heat treament is:

| Element | Range, % |
|---|---|
| Carbon | 0.25-0.30 |
| Manganese | 0.80-1.00 |
| Phosphorus | 0.025 Max. |
| Sulfur | 0.025 Max. |
| Silicon | 0.20-0.60 |
| Nickel | 1.65-2.0 |
| Chromium | 0.75-0.95 |
| Molybdenum | 0.50-1.00 |

An existing specification that provides suitable mechanical properties for cast steel components is Federal Specification QQ-S-681b. It is being used by a number of designers for G.S.E. missile castings at the present time. This specification covers a number of classes based on minimum mechanical properties. These properties range from tensile properties of 60,000 to 150,000 psi and yield points from 30,000 to 125,000 psi. Elongation and reduction of area factors vary accordingly with the tensile strength range.

The chemical analysis range is usually in the plain carbon low alloy class unless the cross-section reaches a point where heat treat hardenability becomes a factor, and then alloys such as chromium and molybdenum are added. Steel to this specification is procurable from most steel foundries with either electric or basic open hearth melting facilities. The choice of the chemical composition is left to the foundry. Mechanical property test bar accompanying the castings is the control factor by the customer.

Requirements on surface finish and surface defects and the determination of the level of quality in this regard can present serious procurement problems, as those who are in the field of supplying or using steel castings are well aware. Specifications on these points should contain definite and measurable limits of acceptance and rejection. Lack of a means of positive measurement has caused misunderstandings between customers and foundries, and, in many cases, the rejection of acceptable castings. Vague or severe specifications on casting finish make difficult the foundries task of accurately pricing new parts, and can result in unnecessary cost to the customer.

Most casting specifications are specific for visual and magnetic particle surface defects such as the positive surface type, which is excess metal caused by metal penetration into the mold material and the negative surface type which can be caused by slag or oxide inclusions, gas, shrinkage, oxidation pitting or cracks. The positive surface defects, generally referred to as excess metal are usually acceptable provided tolerance requirements can be met. There are two types of negative surface irregularities, propagating and nonpropagating.

Propagating defects are those which could conceivably increase in length or width under loading resulting in failure of the part. In this category would fall cracks, hot tears, segregation, pits or holes so aligned as to be propagating in nature. Nonpropagating defects are usually rounded impressions such as gas holes, holes resulting from air entrapment, dross holes, inclusions and pits. Usually such defects can be removed without affecting the serviceability of the part by welding. The extent of such defect correction is determined by the selection of the grade of quality desired in the casting by the customer.

## RADIOGRAPHIC INSPECTION

Internal quality of the casting is checked by determination of subsurface imperfections by means of radiography. A.S.T.M. Radiographic standards have been set up so that various classes of internal quality can be maintained as specified by the customer. In general, Class I x-ray standard required almost a perfectly sound casting, i.e., one free of any propagating imperfections such as cracks, hot tears, dendritic shrinkage, cold shuts and oxide folds. Small imperfections up to $\frac{1}{32}$-in. in diameter of nonpropagating imperfections can be allowed. These could be blow holes, pin hole porosity, globular gas porosity, globular shrinkage cavities and inclusions. Class II standards are more relaxing as to type of allowable imperfections together with size and distribution.

The customer should recognize that the application of the more relaxed standards, which is quite pertinent in the case of the G.S.E. type of casting, will result in a less expensive part and will allow the foundry to concentrate on the more critical areas and thus improve the overall quality of the part. For reasons of economy, the design engineer should furnish a drawing of the part indicating the areas where the various class standards will be in effect. These areas should be defined by dimensional limits to eliminate borderline areas of interpretation.

Along with these possible degrees of quality in a casting, it is well to point out how the determination

of the degree of quality specified for a steel casting can affect its cost to the customer. Quality takes time, materials and effort to make sure that it is there as specified. In many instances, inspection procedure to establish quality specified is more expensive than total casting production cost.

An illustration of how the cost factor is affected when magnetic particle and x-ray inspection standards are specified in the procurement of steel castings is shown by a typical foundry's table of pricing factors for cost estimates.

**Radiographic Examination**

| | |
|---|---|
| *Class I .... II | Add 100-200% depending on casting configuration to casting cost — plus cost of x-ray and magnetic particle inspection, if specified. |
| Class II .... | With magnetic particle add 35% to casting cost — plus cost of x-ray and magnetic particle inspection. |
| Class II .... | Without magnetic particle add 30% to casting cost — plus cost of x-ray. |
| Class III.... | With magnetic particle add 25% to casting cost — plus cost of x-ray and magnetic particle inspection. |
| Class III.... | Without magnetic particle add 17½% to casting cost — plus cost of x-ray. |

**Magnetic Particle only**

Add 22½% to casting cost — plus cost of magnetic particle operation.

**X-ray Exposures**

| | X-ray Class or Standard | | |
|---|---|---|---|
| | I | II | III |
| Single Exposure, $..... | 18.00 | 12.00 | 6.00 |
| **Double Exposure, $..... | 24.00 | 16.00 | 8.00 |

*Orders on Class I castings will be accepted only after approval of each individual design.
**Two films exposed in the same cassette.

The pricing of a casting is determined by the total of the costs of producing the molten metal for the casting, the mold and necessary cores for the shape of the casting and the finishing operations to clean and heat treat the casting for delivery to the customer plus the hazards involved due to quality requirements, plus the cost of the number of x-ray shots specified to assure this quality attainment.

From a realization of these pricing factors, the casting designer can see that if only the critical areas of the steel casting are specified for maximum quality inspection, and if these areas can be relaxed to class three as far as radiographic inspection is concerned and the actual number of x-ray exposures are kept to a minimum, the cost of the casting will be lowered considerably. In addition, the adoption of a standard sampling plan for the radiographic inspection will lower the total cost when a large quantity of castings are involved.

An illustration can be made at this time to show how the aircraft quality complex has extended into the G.S.E. missile design of castings. Several designed prints for G.S.E. castings have been issued via subcontractors for inquiries to steel foundries. The quality inspection standards are noted on these prints to be to MIL-C-6021D, "Classification and Inspection of Castings (for Aeronautical Applications)." This specification states under requirements:

3. REQUIREMENTS

3.1 *Classification* — Castings shall be classified as follows based on function and design margin of safety.

3.1.1 *Class I* — A Class I casting is defined as a casting, the single failure of which during any operating condition would cause loss of the aircraft or one of its major components, loss of control, unintentional release of or inability to release any armament store, failure of gun installation components, or which may cause significant injury to occupants of the aircraft. Target drone castings need not be included in class I. Class I castings shall be further classified in accordance with 3.1.1.1 and 3.1.1.2.

3.1.1.1 *Class IA* — A Class IA casting is a class I casting the single failure of which would result in the loss of the aircraft.

3.1.1.2 *Class IB* — Class IB includes all class I castings not included in Class IA.

3.1.2 *Class II* — Class II castings are all castings not classified as class I. Class II castings shall be further classified in accordance with 3.1.2.1 and 3.1.2.2.

3.1.2.1 *Class IIA* — Class IIA castings are those having a margin of safety of 200 per cent or less.

3.1.2.2 *Class IIB* — Class IIB castings are those having a margin of safety greater than 200 per cent, or those for which no stress analysis is required. All target drone castings may be classed as IIB.

## SUMMARY

Classification of G.S.E. castings in terms of airborne castings would seem to be over extending the quality requirements for grounded castings. Furthermore, the classification of such castings as Class IA or Class IB would appear to be highly questionable, when the Class IIB requirements for the safety factor can be incorporated in the original design. It would seem reasonable that a margin of safety greater than 200 per cent, probably in the order of 300 per cent, could be used in the design of a casting that is intended for G.S.E. service. This design concept would remove the severe radiographic inspection requirement.

Relaxation in radiographic inspection procedures does not mean that the nonfunctional areas of the castings will be ignored. The foundry in its determination of the optimum pattern design will check out the complete soundness of the castings and, provided that time is available, request approval for the attained quality of the casting from the customer before proceeding to full production. Sufficient lead time should be given to the foundry to check out the pattern equipment and correct for desired quality.

For the user of G.S.E. castings to procure satisfactory and economically produced castings, it is a must to design and to specify the minimum quality for the application. Following this, a first class steel foundry can consistently deliver this specified quality. An understanding on this quality concept by missile design engneers is necessary if G.S.E. quality castings are to be readily procurable and adequately serviceable at a price consistent with the practical aspects of the national defense budget. This quality concept should be formalized in terms of acceptable procurement specifications. This can be accomplished by the Aircraft Standards Group of the Aircraft Industries Association, working closely with steel casting suppliers.

# NICKEL ALLOYED AUSTENITIC DUCTILE IRON GRAPHITE STRUCTURES

by I. Karsay and R. D. Schelleng


## ABSTRACT

The occurrence of flake-like chunks of graphite in heavy section castings of nickel containing austenitic ductile iron has been found to be dependent upon the composition of the iron and the freezing rate of the casting. A spheroidal graphite structure is obtained in castings freezing in 40 min or less if the carbon, silicon and nickel contents are adjusted according to the formula:

$$TC\ \% + 0.2\ Si\ \% + 0.06\ Ni\ \% \leq 4.4$$

The iron must be inoculated with calcium bearing ferrosilicon. Cerium is detrimental in austenitic ductile iron.


## INTRODUCTION

An abnormal type of graphite may occur in nickel austenitic ductile iron castings which freeze slowly. This type of graphite forms as fine flake-like chunks in the most slowly cooled portions of the casting. The presence of chunk graphite is detrimental to the mechanical properties of austenitic ductile iron. Typical values are:

| Graphite Structure | Casting Size, in. | Tensile Strength, psi | Yield Strength 0.2% Offset, psi | Elong., % | Bhn | Melt No. |
|---|---|---|---|---|---|---|
| Spheroidal | 1 in. Keel | 55-60,000 | 25-30,000 | 20-30 | 125-150 | — |
| Spheroidal | 5 x 5 x 10 in. | 56,000 | 27,200 | 20½ | 119 | 1 |
| 40% Chunk | 5 x 5 x 10 in. | 45,300 | 28,000 | 10 | 117 | 24 |
| 95% Chunk | 5 x 5 x 10 in. | 42,300 | 28,000 | 9½ | 111 | 27 |

When chunk graphite occurs as compared with spheroidal graphite the tensile strength is reduced by about 20-25 per cent, and the elongation is reduced by about 50 per cent. The yield strength is unaffected. This report describes an investigation which was undertaken to determine the cause of this abnormal graphite structure and to develop methods for the prevention of its occurrence.

## PROCEDURE

The alloys tested were melted in an indirect arc furnace from a charge of low phosphorus pig iron, steel scrap, electrolytic nickel, ferrosilicon, ferromanganese and, in the case of chromium bearing alloys, ferrochromium. The liquid metal was tapped at 2900 F, cooled in the ladle to 2750 F, treated with nickel magnesium alloy, post inoculated with 0.5 per cent ferro-85-silicon (calcium bearing) and poured. Exceptions are indicated in the table. The test casting employed was a 65 lb, 5 x 5 x 10 in. block requiring 40 min to freeze.

The examination of the castings consisted of chemical analysis, microscopic examination along the central vertical axis, and occasional mechanical tests.

## RESULTS AND DISCUSSION

### *Chunk Graphite*

The presence of chunk graphite can be detected in many cases without microscopic examination. A machined surface of austenitic ductile iron develops different textures in areas containing spheroidal and chunk graphite, as shown in Fig. 1. The areas containing chunk graphite, the dull spots, are located in the central part of the casting, while the more rapidly cooled parts contain spheroidal graphite. This distribution indicates the importance of cooling rate in the formation of chunk graphite.

Under the microscope, chunk graphite is seen to occur in globular cells. Spheroidal graphite or car-

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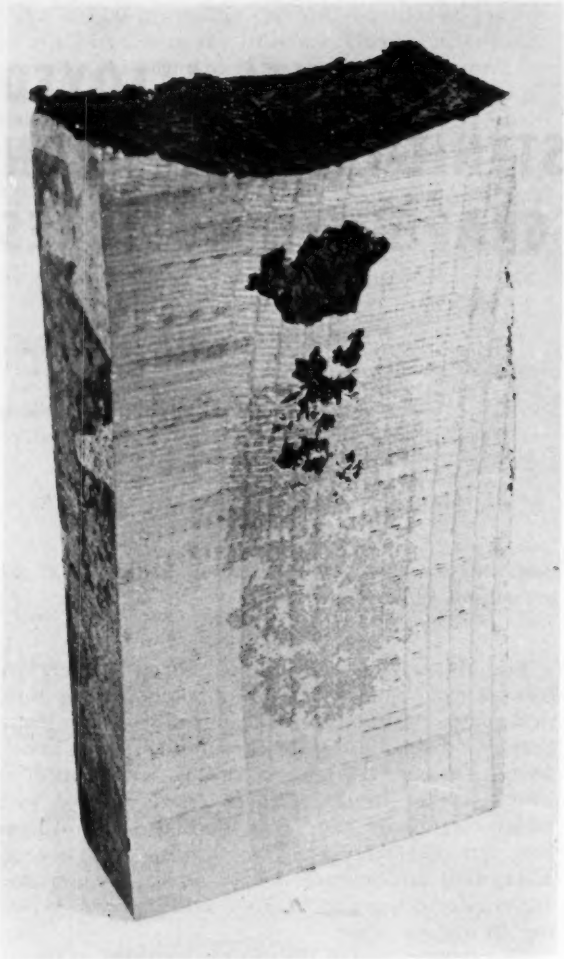


Fig. 1 — Gray spots on the saw cut surface of a 5 x 5 x 10 in. austenitic ductile iron block (melt 28), indicating the presence of chunk graphite.

bide, depending upon the graphitizing tendency of the particular alloy, forms in the intercellular regions. Within the cells the distribution of the graphite chunks is interdendritic, i.e., a skeleton of graphite-free primary austenite dendrites is imposed over the cell structure. Figures 2 and 3 illustrate these characteristics of this type of graphite.

The appearance of chunk graphite in the photographs seen so far resembles interdendritic, A.S.T.M. Type D flake graphite. However, examination at a higher magnification shows that this is not the case, and that chunk graphite is different from any other type studied in the literature. The shape of the individual particle is not that of flakes but of cones or pyramids occurring either individually or in irregular groupings, as shown in Fig. 4.

The polarized light photograph (Fig. 5) shows the crystalline structure of the chunks to be substantially different from that of graphite flakes. There is some resemblance to the inner structure of graphite spheroids. It is possible, that graphite chunks are the building blocks of spheroidal graphite, but are scattered instead of collected radially around one center. Figure 6 seems to support this explanation.

### *Carbon, Silicon, Nickel and Magnesium Content Effect*

It was found that chunk graphite can be eliminated by lowering the carbon or silicon contents. A decrease of the silicon content by one per cent was determined to be equivalent to a 0.2 per cent decrease in carbon content for nickel levels of 20 to 40 per cent. The maximum permissible combination of carbon and silicon levels is a function of the nickel content. Lower values are required at higher nickel levels. These relationships permit all the data obtained to be presented on one plot of per cent nickel versus per cent carbon plus 0.2 (per cent silicon).

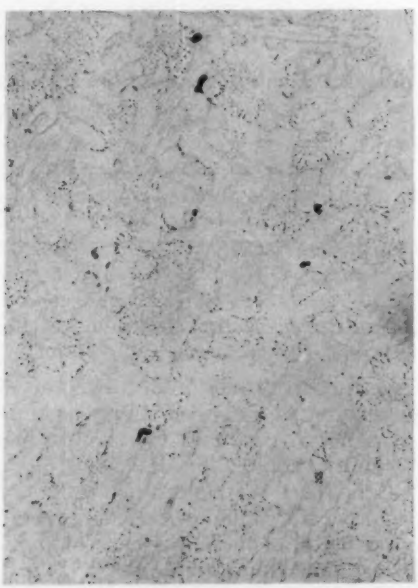


Fig. 2 — Chunk graphite with ledeburite in casting 26. 3 per cent nital etch. 30 X.

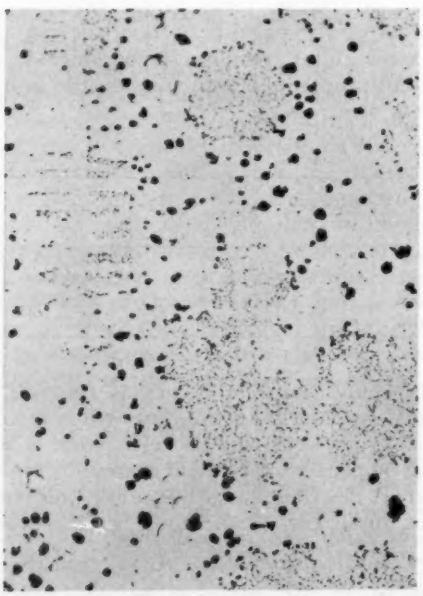


Fig. 3 — Chunk graphite cells surrounded by spheroidal graphite in casting 41. Unetched. 30 X.

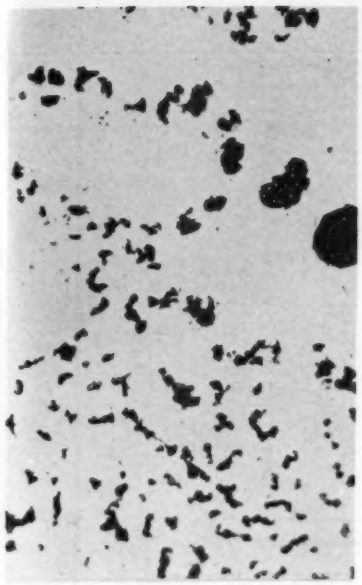


Fig. 4 — Chunk graphite in casting 41. Unetched. 200 X.

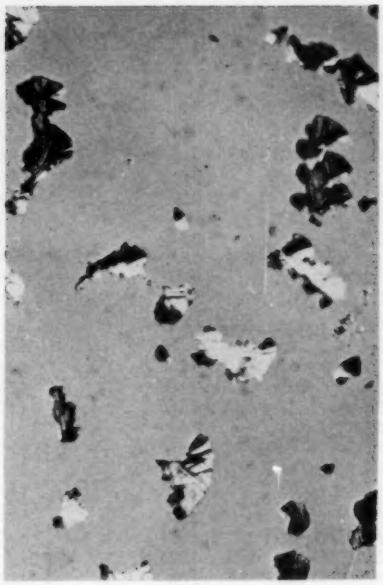


Fig. 5 — Radial structure in chunk graphite. Unetched. Polarized light. 750 X.



Fig. 6 — Imperfectly formed graphite spheroid associated with chunk graphite. Unetched. 500 X.

It can be seen in Fig. 7 that the points representing spheroidal and chunk graphite structures on such a plot are separated by a straight line. This line is described by the equation:

$$\% \text{ TC} + 0.2\ (\% \text{ Si}) + 0.06\ (\% \text{ Ni}) = \text{A Constant}$$

The sum of the left side of the above equation will be called saturation number (S.N.) to avoid confusion with the eutectic carbon equivalent.* Detailed data of the chemical compositions are given in the table for a selected group of experimental alloys with compositions close to the division line in Fig. 7, and those which contribute to the other conclusions of the present work. Expanding the table by the inclusion of all of the alloys plotted in Fig. 7 was judged to be unnecessary.

Figure 8 is derived from Fig. 7, and shows the maximum permissible combination of carbon and silicon contents for the common types of nickel austenitic ductile irons. The satisfactory spheroidal graphite structure is a heavy casting from melt 47, shown in Fig. 9, is typical of the structure which can be obtained when the composition is controlled within the proper limits. The magnesium contents of the melts upon which the above plots are based were maintained between 0.074 and 0.110 per cent and the casting solidified in about 40 min. Under these conditions the permissible saturation number is 4.5.

The magnesium content slightly influences the permissible saturation number. At a magnesium content of 0.04 per cent a satisfactory graphite structure was achieved at a maximum saturation number of 4.4, while at about 0.09 per cent magnesium a saturation number of 4.5 can be tolerated as shown in melts 3, 4, 34 and 37.

*The formula for the carbon equivalent is: CE = % TC + 0.33 (% Si) + 0.047 (% Ni) − 0.055 (% Ni) % Si (from a reference[1]).

Fully spheroidal graphite structures were sometimes obtained at saturation numbers higher than those mentioned, but no chunk graphite structure was ever observed at lower saturation numbers. The magnitude of the scatter in the data corresponds approximately to the standard error range of the chemical analyses.

### *Inoculation Effect*

The tendency of irons of high saturation number to contain chunk graphite in heavy sections has been found to be an abnormal effect of inoculation. In the uninoculated condition coarse irregular graphite structures occur as shown in Fig. 10. Such structures yield poor properties. Omission of inoculation in melt 51 resulted in a tensile strength of only 44,000 psi with 10 per cent elongation.

When an iron with a saturation number less than 4.4 is properly inoculated with calcium bearing ferrosilicon, a regular spheroidal graphite structure and good properties are developed in heavy sections. Irons with saturation numbers greater than 4.4 develop chunk graphite when inoculated and no improvement in properties is achieved.

### *Cerium and Other Alloying Elements Effect*

Cerium is known to counteract the harmful effects of certain impurities in ductile iron. However, cerium contents of 0.0086 and 0.003 per cent were found to be extremely detrimental in austenitic ductile iron, causing chunk graphite even in one and 2 in. keels of heats 65 and 66. The saturation numbers of these heats were 4.14 and 4.34, respectively, much lower than the critical 4.4 level found in much heavier castings where no cerium was present. It is evident that the amount of cerium resulting from the use of the commercially available, magnesium additives which contain one per cent rare earths cannot be tolerated in these ductile irons. In the presence of lead and

## CHEMICAL COMPOSITIONS AND GRAPHITE STRUCTURES OF AUSTENITIC DUCTILE IRON CASTINGS
(5 x 5 x 10 In. Blocks Unless Marked Otherwise)

| Melt No. | TC | Si | Mn | Ni | Cr | Mg | Saturation Number SN = TC + 0.2 Si + 0.06 Ni | Graphite Structure | Remarks |
|---|---|---|---|---|---|---|---|---|---|
| 1 | 2.65 | 1.20 | (2.0) | (23.0) | 0.0 | (0.08) | 4.28 | S | |
| 2 | 2.72 | 2.01 | (2.0) | 22.2 | 0.0 | 0.078 | 4.46 | S | |
| 3 | 2.98 | 1.55 | (0.8) | 19.6 | 2.52 | 0.082 | 4.48 | S | |
| 4 | 2.85 | 2.22 | (0.8) | 20.1 | (2.45) | 0.090 | 4.51 | C | |
| 5 | 2.70 | 2.09 | 1.69 | 23.1 | 0.0 | 0.092 | 4.51 | C | |
| 6 | 2.79 | 1.74 | (2.0) | 23.1 | 0.0 | 0.110 | 4.54 | S | |
| 7 | 2.14 | 2.55 | (0.5) | 31.5 | 0.0 | 0.110 | 4.54 | C | |
| 8 | 2.72 | 2.16 | (2.0) | 23.2 | 0.0 | 0.095 | 4.55 | S | |
| 9 | 2.63 | 2.63 | (2.0) | 23.1 | 0.0 | 0.086 | 4.55 | S | |
| 10 | 2.85 | 2.52 | (0.8) | 20.0 | 2.64 | 0.095 | 4.56 | S | |
| 11 | 2.74 | 3.01 | (0.8) | 20.1 | (2.64) | 0.110 | 4.56 | C | |
| 12 | 2.91 | 2.14 | (0.8) | 20.3 | 2.42 | 0.100 | 4.56 | S | |
| 13 | 2.29 | 1.95 | (0.5) | 31.2 | 0.0 | 0.110 | 4.56 | C | |
| 14 | 2.41 | 3.84 | (2.0) | 23.1 | 0.0 | 0.110 | 4.57 | C | |
| 15 | 2.64 | 2.61 | (2.0) | 23.5 | 0.0 | 0.095 | 4.58 | C | |
| 16 | 2.98 | 2.03 | (0.8) | 19.6 | (2.50) | 0.074 | 4.58 | S | |
| 17 | 2.57 | 1.16 | 0.19 | 29.7 | 2.98 | 0.097 | 4.59 | S | |
| 18 | 2.86 | 1.80 | (2.0) | (23.0) | 0.0 | (0.08) | 4.61 | C | |
| 19 | 2.22 | 2.33 | (0.5) | (32.0) | 0.0 | (0.08) | 4.61 | C | |
| 20 | 2.15 | 2.72 | (0.5) | (32.0) | 0.0 | (0.08) | 4.62 | C | |
| 21 | 2.18 | 2.64 | (0.5) | (32.0) | 0.0 | (0.08) | 4.63 | C | |
| 22 | 2.56 | 1.69 | 0.19 | 28.8 | 3.07 | (0.08) | 4.63 | C | |
| 23 | 3.03 | 1.08 | (2.0) | (23.0) | 0.0 | (0.08) | 4.63 | C | |
| 24 | 2.58 | 3.18 | (2.0) | 23.6 | 0.0 | 0.110 | 4.64 | C | |
| 25 | 2.20 | 2.58 | (0.5) | (32.0) | 0.0 | (0.08) | 4.64 | C | |
| 26 | 1.56 | 6.26 | (0.5) | (32.0) | (4.50) | (0.08) | 4.74 | C | |
| 27 | 2.94 | 2.46 | (2.0) | (23.0) | 0.0 | (0.08) | 4.82 | C | |
| 28 | (2.70) | (1.70) | (0.5) | (32.0) | 0.0 | (0.08) | 4.96 | C | |
| 29 | 2.75 | 2.99 | (0.5) | 19.72 | 0.0 | 0.037 | 4.54 | C | |
| 30 | 2.14 | 3.33 | (0.5) | 29.67 | 0.0 | 0.045 | 4.59 | C | |
| 31 | 2.73 | 2.55 | (0.5) | 19.72 | 0.0 | 0.036 | 4.43 | C | |
| 32 | 2.86 | 1.52 | (0.5) | 19.70 | 0.0 | 0.043 | 4.35 | S | |
| 33 | 2.15 | 2.68 | (0.5) | 29.82 | 0.0 | 0.043 | 4.48 | C | |
| 34 | 2.17 | 2.17 | (0.5) | 29.80 | 0.0 | 0.044 | 4.39 | S | |
| 35 | 2.13 | 1.67 | (0.5) | 29.63 | 0.0 | 0.046 | 4.25 | S | |
| 36 | 2.86 | 2.07 | (0.5) | 19.68 | 0.0 | 0.036 | 4.46 | C | |
| 37 | 2.57 | 1.96 | (2.0) | 24.0 | 0.0 | 0.041 | 4.41 | C | |
| 38 | 2.47 | 2.15 | (2.0) | 21.9 | 0.0 | 0.042 | 4.22 | S | |
| 39 | 2.50 | 2.12 | (2.0) | 21.9 | 0.0 | 0.046 | 4.25 | S | |
| 40 | 2.70 | 2.41 | (2.0) | 23.0 | 0.0 | 0.050 | 4.57 | C | |
| 41 | 2.79 | 2.37 | 2.07 | 23.0 | 0.0 | 0.053 | 4.65 | C | |
| 42 | 2.56 | 2.12 | (2.0) | 21.9 | 0.0 | 0.064 | 4.31 | S | |
| 43 | 2.73 | 2.39 | (2.0) | 23.1 | 0.0 | 0.066 | 4.60 | S | |
| 44 | 2.38 | 2.03 | (0.5) | 31.9 | 0.0 | 0.110 | 4.70 | S | |
| 45 | 2.22 | 1.99 | (0.5) | 31.0 | 0.0 | 0.120 | 4.48 | S | |
| 46 | 1.95 | 1.05 | (0.5) | 39.1 | 0.0 | 0.130 | 4.51 | C | |
| 47 | 1.64 | 1.83 | (0.5) | (40.0) | 0.0 | (0.14) | 4.41 | S | |
| 48 | 1.59 | 2.92 | (0.5) | 39.7 | 0.0 | 0.140 | 4.56 | C | |
| 49 | 2.56 | 2.04 | (2.0) | (23.0) | 0.0 | (0.080) | 4.36 | I | No post inoculation |
| 50 | (2.50) | (1.50) | (0.5) | (32.0) | 0.0 | (0.08) | 4.72 | I | No post inoculation |
| 51 | (2.50) | (1.60) | (0.5) | (32.0) | 0.0 | (0.08) | 4.74 | I | No post inoculation |
| 52 | (2.95) | (2.50) | (2.0) | (23.0) | 0.0 | 0.100 | 4.82 | S | 4 in. step bar |
| 53 | 2.30 | 4.67 | (0.4) | 30.5 | 4.34 | 0.096 | 5.07 | C | 2 in. keel block |
| 54 | 2.88 | 1.86 | (2.0) | (23.0) | 0.0 | (0.08) | 4.64 | C | Slowly cooled 7 in. cube |
| 55 | 2.25 | 2.50 | (1.0) | (22.0) | 0.0 | 0.120 | 4.08 | S | Slowly cooled 7 in. cube |
| 56 | 2.40 | 1.77 | (2.0) | 22.5 | 0.0 | 0.044 | 4.11 | S | 8 in. cube |
| 57 | 2.44 | 1.57 | (2.0) | 23.0 | 0.0 | 0.10 | 4.14 | S | 8 in. cube |
| 58 | 2.31 | 2.56 | (2.0) | 22.5 | 0.0 | 0.045 | 4.18 | S | 8 in. cube |
| 59 | 2.48 | 2.55 | (2.0) | 23.3 | 0.0 | 0.120 | 4.40 | S | 8 in. cube |
| 60 | 2.66 | 1.77 | (2.0) | 23.5 | 0.0 | 0.110 | 4.43 | S | 8 in. cube |
| 61 | 2.70 | 2.67 | (2.0) | 21.7 | 0.0 | 0.200 | 4.54 | C | 8 in. cube |
| 62 | 2.89 | 1.89 | (2.0) | 21.9 | 0.0 | 0.200 | 4.59 | S | 8 in. cube |
| 63 | 2.64 | 2.79 | (2.0) | 23.4 | 0.0 | 0.120 | 4.61 | C | 8 in. cube |
| 64 | 2.27 | 2.64 | (0.5) | (32.0) | 0.0 | (0.08) | 4.72 | C | 0.17% phosphorus |
| 65 | 2.42 | 2.49 | 0.77 | 20.2 | 0.0 | 0.110 | 4.14 | C | 1 in. keel block containing 0.0086% Ce |
| 66 | 2.73 | (2.0) | (0.8) | (20.0) | (2.0) | (0.06) | 4.34 | C | 2 in. keel block containing 0.003% Ce |

Legend:
S — Spheroidal graphite
C — Chunk graphite present
I — Coarse Irregular graphite
Values in parentheses are calculated.

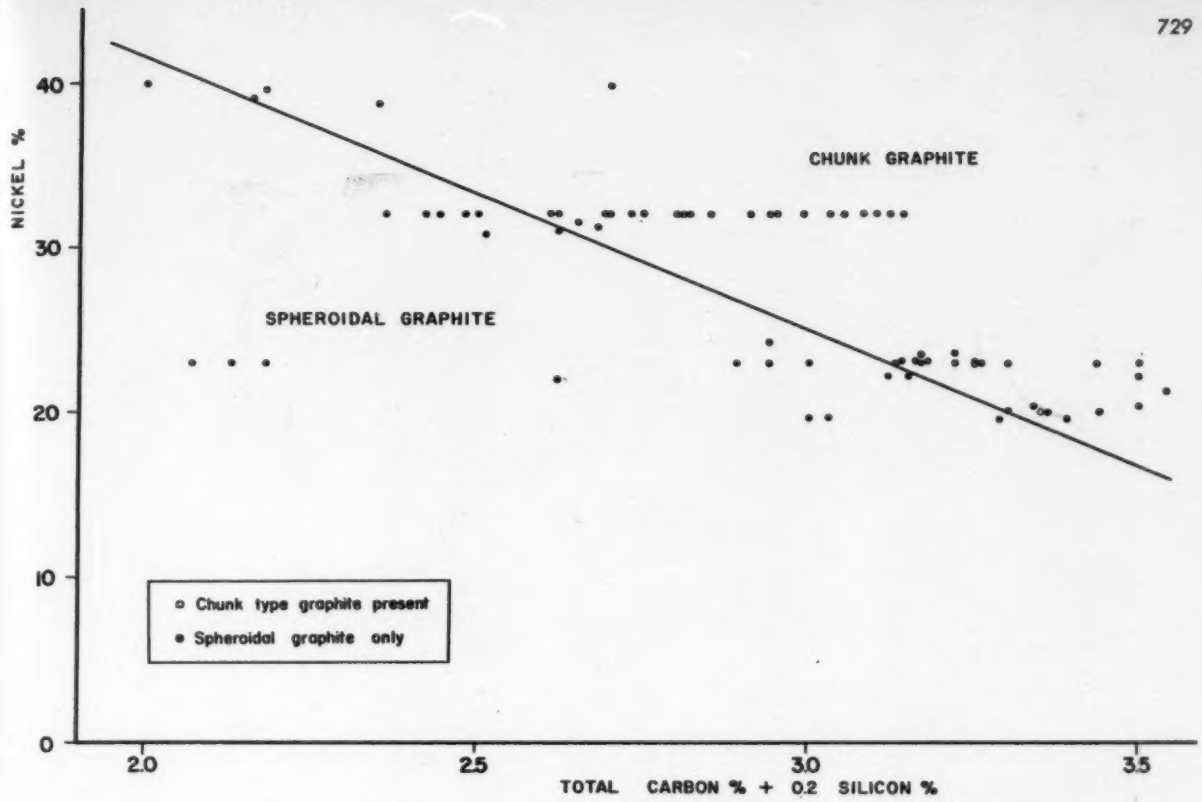
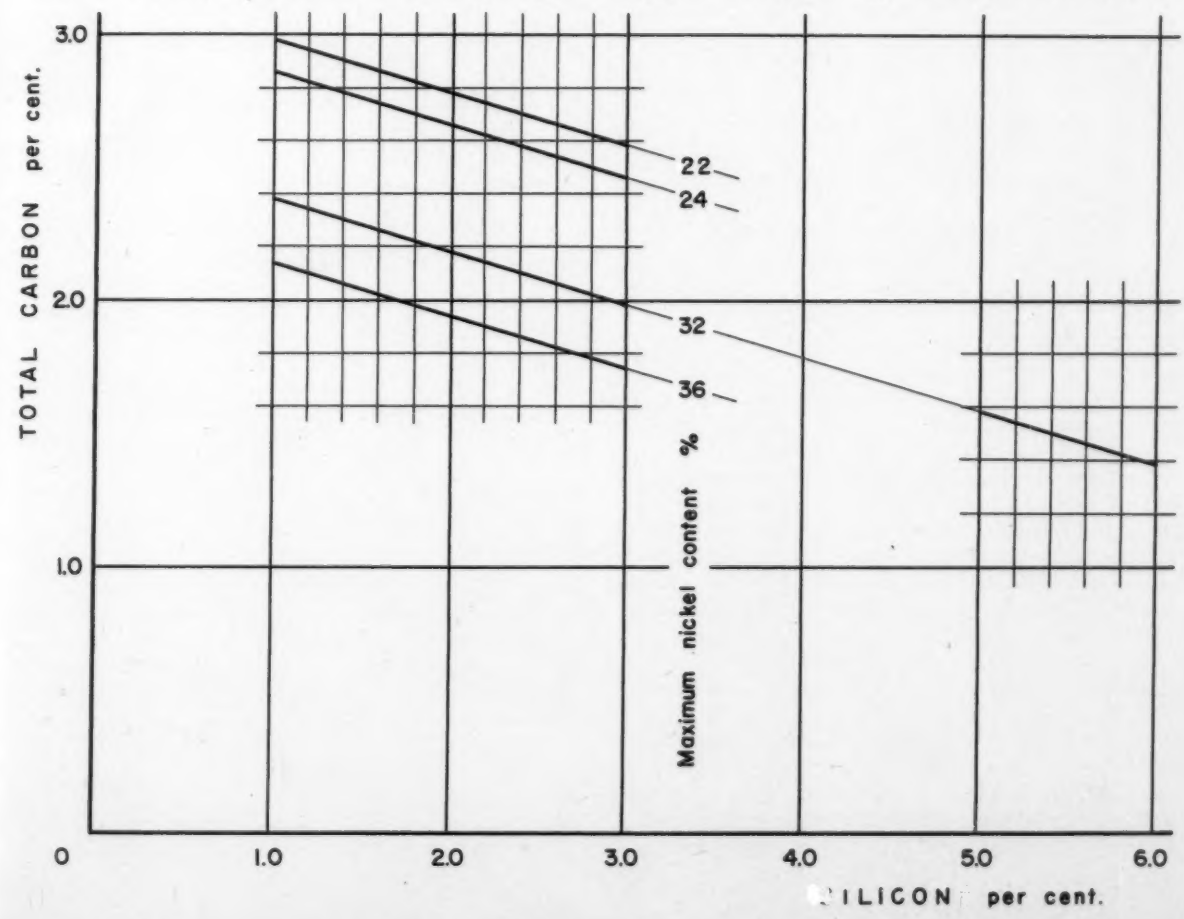



Fig. 7 — (above) — Occurrence of chunk graphite in slowly cooled austenitic ductile iron blocks (solidification time 40 min) as a function of the sum of TC per cent plus 0.2 Si per cent and the nickel content. Fig. 8 — (below) — Nickel, carbon and silicon contents effect on the occurrence of chunk graphite in the common types of austenitic ductile iron.



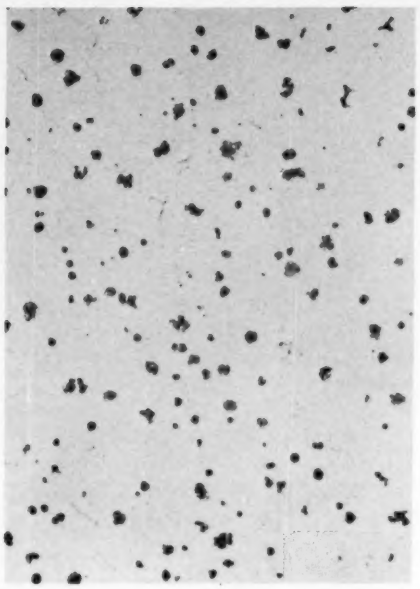
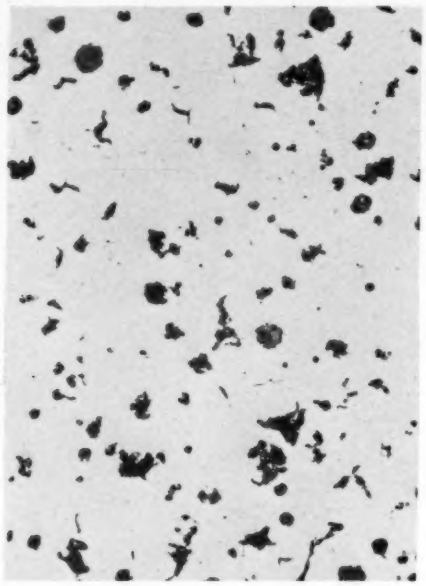


Fig. 10 — Irregular graphite in noninoculated slowly cooled austenitic ductile iron casting (melt 50). Unetched. 30 X.

Fig. 9 — Spheroidal graphite in 5 x 5 x 10 in. austenitic ductile iron casting (melt 47) with a saturation number of 4.41. Unetched. 30 X.



perhaps other such elements this detrimental effect of cerium does not occur.

Other elements also were investigated and were found to have no effect upon the occurrence of chunk graphite:

Chromium between the concentrations of 0 and 4.5%
Manganese between the concentrations of 0.19 and 2.0%
Phosphorus between the concentrations of 0.015 and 0.17%

(See the table for further details).

### *Freezing Rate Effect*

As mentioned previously, a saturation number of 4.4-4.5 is critical for castings freezing in 40 min. Small castings, which solidify more rapidly, have been made satisfactorily with higher saturation numbers for many years. Results from a limited number of castings solidifying either faster or slower than 40 min demonstrate the influence of the solidification rate. In an 8 in. cube, solidifying in approximately 80 min, a satisfactory structure was obtained in heat 60 with a saturation number of 4.43, while a value of 4.61 in heat 63 produced chunk graphite.

A 4 in. cube with a freezing time of about 15 min exhibited satisfactory graphite at a saturation number of 4.82, while a value of 5.07 in a casting with a similar freezing time produced chunk graphite (heats 52 and 53). These values show that as the freezing time decreases below 40 min the permissible saturation number increases rapidly. With freezing times longer than 40 min the permissible value decreases only slightly if at all. These values are for irons containing about 0.08 per cent magnesium.

## CONCLUSIONS

Austenitic ductile irons with nickel contents of 20 to 40 per cent will exhibit fully spheroidal graphite structures in heavy sectioned castings, if the carbon or silicon contents are not excessive. The composition must satisfy the formula:

$$TC + 0.2\,(\%\ Si) + 0.06\,(\%\ Ni) \leqq 4.4$$

Irons not satisfying this formula contain fine, flake-like chunk graphite in heavy sections. In light castings much higher carbon and silicon contents can be tolerated.

The chunk form of graphite observed in heavy sections is an abnormal effect of inoculation in irons of high carbon and silicon contents. Addition of a small amount of cerium to high nickel ductile iron produces chunk graphite even in light castings with low carbon and silicon contents.

## ACKNOWLEDGMENT

The authors are grateful to The International Nickel Co., Inc. for permission to publish the results of this investigation. Ductile Iron is a patented material produced under license from The International Nickel Co.

## REFERENCE

1. R. D. Schelleng, "Influence of Nickel up to 37 Per Cent and Silicon up to 3 Per Cent on the Eutectic Carbon Content of Cast Iron," AFS TRANSACTIONS, *vol. 68*, p. 301 (1960).

# IMPROVE STEEL CASTING FATIGUE LIFE BY CONTROLLED GAS ATMOSPHERE HEAT TREATMENT

by R. D. Block


## ABSTRACT

**Design and foundry engineers have made a concentrated effort to eliminate stress raisers in steel castings, such as sharp corners, nonmetallic inclusions and casting surface defects. However, stress raisers caused by scaling and decarburization are too frequently ignored by heat treaters. Here is where this report puts emphasis — on the heat treatment of steel castings under various types of controlled gas atmosphere. Through the use of these controlled gas atmospheres, the heat treaters, according to the author, can aid the foundryman in fatigue life improvements.**


## STEEL CASTING FATIGUE

Steel castings, like their counterparts, forgings and wrought products, must be capable of withstanding the numerous alternating tensile and compressive stresses encountered in aircraft and missile applications. Dynamic loading causes a reduction of the allowable yield stress to well below the allowable yield stress under static loading. This is what is meant by fatigue. The allowable dynamic yield stress is even further reduced by inherent or imposed stress raisers.

Some of the items that concentrate residual stresses and act as stress raisers are – (1) sharp fillets, (2) keyways, (3) nicks, (4) sharp corners, (5) cracks, (6) surface seams, (7) nonmetallic inclusions that break the surface, (8) oxidized or scaled surface and (9) decarburized surface.

An example of a splined shaft which failed from fatigue prematurely due to a stress raiser at the bottom of one of the splines, is illustrated in Fig. 1. The shell-like markings on the fractured face (starting at the arrow), reveal the progress of a fatigue crack, and are typical of fatigue fractures. Anyone of the listed stress raisers are capable of causing this same type of failure. Steel castings, when free of these stress raisers, have inherent uniformly distributed, residual compressive stresses at the surface and residual tensile stresses in the interior.

This is beneficial. Both fatigue testing and experience have proved that compressive residual stresses at the surface delay fatigue failures, while tensile residual stresses hasten fatigue failures. From this it can be seen that steel castings when properly designed, melted, cast and heat treated make excellent parts for aircraft and missile applications.

## HEAT TREATMENT EMPHASIS

Both design and foundry engineers have made a concentrated effort to eliminate such stress raisers as



Fig. 1 — Typical fatigue fracture.

R. D. BLOCK is with Barber-Colman Co., Wheelco Ind. Instruments Div., Rockford, Ill.

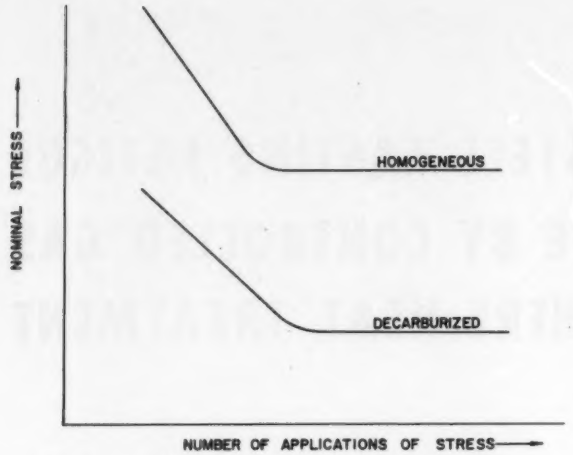



**Fig. 2 — Typical results of fatigue tests conducted with the same medium carbon steels quenched and tempered to the same core hardness level.**

sharp corners, nonmetallic inclusions and casting surface defects. However, heat treaters have too frequently ignored the stress raisers caused by scaling and decarburization. This, then, is where the major emphasis must now be placed.

The surfaces of steel castings when oxidized in heat treatment and especially when scaled, contain residual tensile stresses due to oxidation and surface irregularities. These irregularities are miniature notches and sharp corners as part of the scaled surface. The other concentrator of stresses on the surface of castings, resulting from improper heat treatment, is the change of the chemical composition of the steel surface by gaseous diffusion. This process of gaseous diffusion occurs during carburizing or decarburizing of steel castings.

A decarburized steel casting contains a detrimental residual tensile stress on the surface which in turn lowers its fatigue strength. The detrimental effects of decarburization of medium carbon alloy steels (0.30 - 0.60 per cent carbon) are shown in Fig. 2. These curves show typical results of samples of the same type of steel, quenched and tempered to the same core hardness level, with one having a homogeneous carbon content throughout and the other having lost some of the original carbon from its surface during heat treatment.

It follows, therefore, that the castings should not be permitted to become either oxidized or decarburized during their heat treatment. To avoid this oxidation and/or loss of carbon, the surfaces of the castings must be protected while they are subjected to the heat. Each and every alloy, besides having known temperature ranges to produce the necessary properties, have also known requirements for protective atmospheres.

## PROTECTIVE ATMOSPHERES

There are three basic types of protective atmospheres, namely — (1) gas, (2) liquid and (3) vacuum. Of these, gas is the most commonly used atmosphere for control of the carbon content of steels. The gas atmosphere envelops the castings with a special prepared gas that is produced externally to the heat treat furnace and piped into the gas tight (indirectly fired) furnace chamber which contains the castings.

The different protective gas atmospheres used during the heat treatment of steel castings are — (1) reacted or cracked fuel gas (endothermic), (2) burned fuel gas (exothermic), (3) nitrogen, (4) hydrogen and (5) dissociated ammonia.

### *Reacted or Cracked Fuel Gas Atmosphere*

The reacted or cracked fuel gas atmosphere, referred to in the heat treat industry as endothermic gas (Fig. 3) is produced by taking metered amounts of raw fuel gas (natural gas, propane, butane, etc.) and room air, mixing and then heating over a catalyst in a pressure tight, externally heated combustion chamber or retort. This heating of the fuel gas and air mixture generates a chemical reaction which cracks the fuel gas into a composite mixture of gaseous compounds and elements.

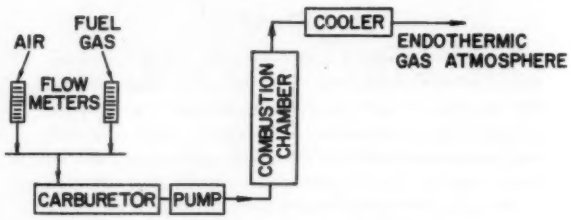



**Fig. 3 — Endothermic atmosphere generator.**

The resulting gas generated in this manner depends on the proportion of the air to the fuel gas, the temperature in the cracking combustion chamber, the time the airfuel gas mixture is exposed to the cracking temperatures and the cooling power of the cooler which sets the chemical reaction. The resulting endothermic gas from a mixture of 2.75 parts of air to one part of natural gas would consist approximately of:

| Analysis | % by Volume |
|---|---|
| Hydrogen | 37.8 |
| Carbon monoxide | 19.6 |
| Carbon dioxide | 0.5 |
| Water | 0.67 = dewpoint + 35 F (+ 2 C) |
| Methane | 0.8 |
| Nitrogen | 40.6 |

Variations in the percentages of these gaseous constitutents occur with a definite equilibrium relationship between each, as shown in Fig. 4.

### *Burned Fuel Gas Atmosphere*

The burned fuel gas atmosphere referred to as exothermic gas (Fig. 5) is generated by mixing metered raw fuel gas with larger quantities of room air than for the endothermic gas, so that more com-

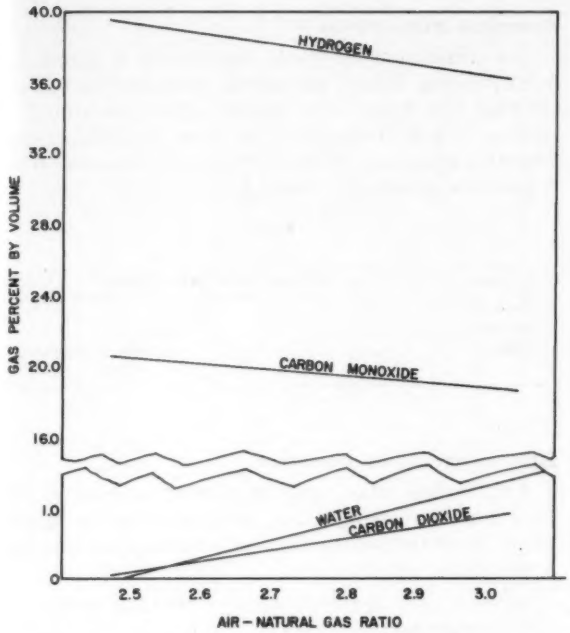



Fig. 4 — Equilibrium relationship of gaseous constituents in endothermic gas generated from natural gas and air.

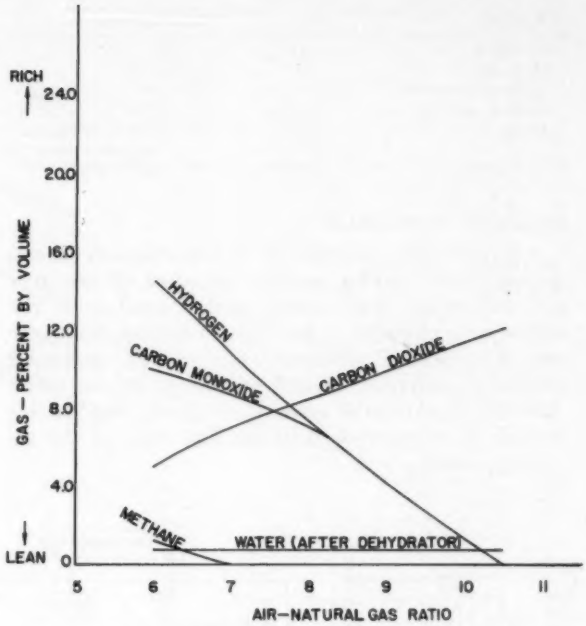



Fig. 6 — Equilibrium relationship of gaseous constituents in exothermic gas generated from natural gas and air.

bustion of the fuel gas can take place. The mixed air-fuel gas generates its own heat in the combustion chamber when it partially burns as it passes over a catalyst, forming carbon dioxide and water vapor.

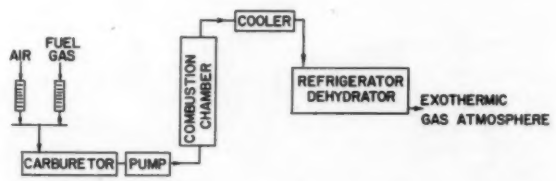



Fig. 5 — Exothermic atmosphere generator.

The exothermic gas, as its exits from the combustion chamber, is saturated with water vapor. This large quantity of water vapor is reduced to a specified quantity by passing the saturated gas through a water-cooled condenser (cooler) and refrigerator-dehydrator. A mixture of 10 parts of air to one part of natural gas removed after the cooler and dehydrator would provide an atmosphere of the composition:

| Analysis | % by Volume |
|---|---|
| Hydrogen | 1.0 |
| Carbon monoxide | 1.0 |
| Carbon dioxide | 11.5 |
| Water | 0.8 = dewpoint + 40 F (+ 4 C) |
| Methane | 0.0 |
| Nitrogen | 86.2 |

Figure 6 shows the equilibrium relationship between the constitutents as the ratio of air to natural gas is changed.

### *Nitrogen Atmosphere*

The nitrogen atmosphere (fuel gas-air type) is produced by completely burning the mixture of raw fuel gas and air (as in an exothermic generator), and removing the carbon dioxide and water vapor which is formed. Figure 7 is a schematic flow diagram of a nitrogen generator. The gas generated in the combination chamber, being high in water vapor and carbon dioxide, progressively passes through a water-cooled condenser, a carbon dioxide absorbing tower, a carbon monoxide to carbon dioxide converter, another water-cooled condenser, a second carbon dioxide absorbing tower, a refrigerator-dehydrator and an activated alumina drying tower. Through this process, the contaminants (carbon monoxide, carbon dioxide, and water) are reduced. A nitrogen gas produced from 10 parts of air and one part of natural gas in this type of a generator would have the approximate composition:

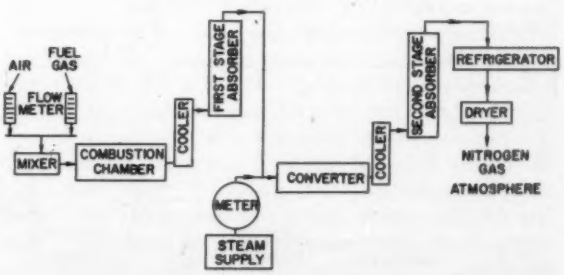



Fig. 7 — Nitrogen atmosphere generator (fuel gas-air type).

| Analysis | % by Volume |
|---|---|
| Nitrogen | 98.0 |
| Hydrogen | 1.8 |
| Carbon monoxide | 0.10 |
| Carbon dioxide | 0.10 |
| Water | 0.015 = dewpoint −40 F (−40 C) |

### *Hydrogen Atmosphere*

The hydrogen atmosphere is generated as shown in Fig. 8. By mixing metered amounts of raw fuel gas and steam, and heating over a catalyst in the combustion chamber, a gas high in carbon monoxide and hydrogen is produced. The carbon monoxide formed is converted to carbon dioxide by the introduction of additional amount of steam. The carbon dioxide is then removed in the first stage of the absorbing tower.

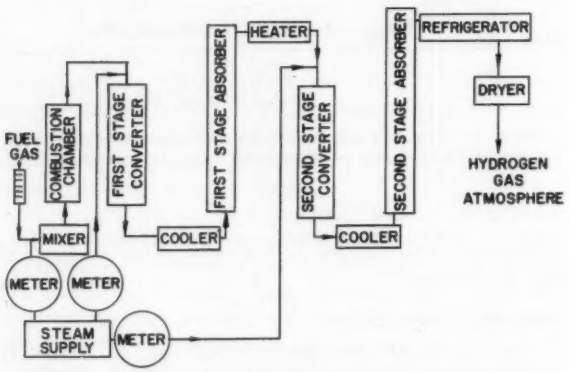



Fig. 8 — Hydrogen atmosphere generator.

After heating the partially purified gas, further purification is accomplished by converting the remaining minute quantity of carbon monoxide to carbon dioxide in the second stage converter, which is then removed in the second absorbing tower. The wet purified gas is then dried by refrigeration and desiccant dryers. Hydrogen gas generated from this type of generator would have the approximate composition:

| Analysis | % by Volume |
|---|---|
| Hydrogen | 99.76 |
| Carbon monoxide | 0.20 |
| Carbon dioxide | 0.02 |
| Water | 0.004 = dewpoint −60 F (−51 C) |

### *Ammonia Atmosphere*

The dissociated ammonia atmosphere is produced by vaporizing liquid anhydrous ammonia and then cracking this vapor in a heated retort containing a catalyst. Figure 9 diagrams the flow of a dissociated ammonia generator. The approximate composition of a gas thus generated would be:

| Analysis | % by Volume |
|---|---|
| Hydrogen | 75 |
| Nitrogen | 25 |
| Water | 0.004 = dewpoint −60 F (−51 C) |

### *Atmosphere Type Selection*

The casting alloy, the heat treat operation, the heat treat furnace and cost of producing the atmosphere determine which atmosphere should be used for a particular application.

Some typical applications and selected protective atmospheres are:

| Steel Casting Type | Heat Treatment | Gas Atmosphere |
|---|---|---|
| Low Carbon (less than 0.20% C) | Anneal, Normalize, Temper | Exothermic |
| Low Carbon (less than 0.20% C) | Harden | Endothermic |
| Medium Carbon (0.20% to 0.50% C) | Anneal, Normalize, Harden | Endothermic |
| Medium Carbon (0.20% to 0.50% C) | Temper | Exothermic |
| High Carbon (more than 0.50% C) | Anneal, Normalize, Harden, Temper | Endothermic |
| Low Alloy (total alloy less than 8%) | Anneal, Normalize, Harden | Endothermic |
| Low Alloy (total alloy less than 8%) | Low Temperature Temper | Exothermic |
| Low Alloy (total alloy less than 8%) | High Temperature Temper | Endothermic |
| Silicon-Iron | Anneal | Nitrogen |
| Heat Resistant (Iron-Chromium) | Anneal, Normalize, Harden, Temper | Dissociated Ammonia |
| Heat Resistant (Iron-Chromium-Nickel) | Anneal | Hydrogen |
| Heat Resistant (Iron-Nickel-Chromium) | Anneal | Hydrogen |
| Corrosion Resistant | Anneal, Harden, Temper | Hydrogen |
| Austenitic Manganese | Anneal | Endothermic |

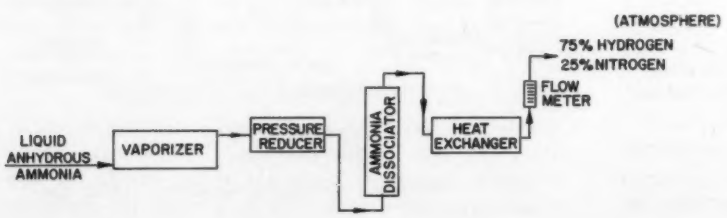



Fig. 9 — Dissociated ammonia generator.

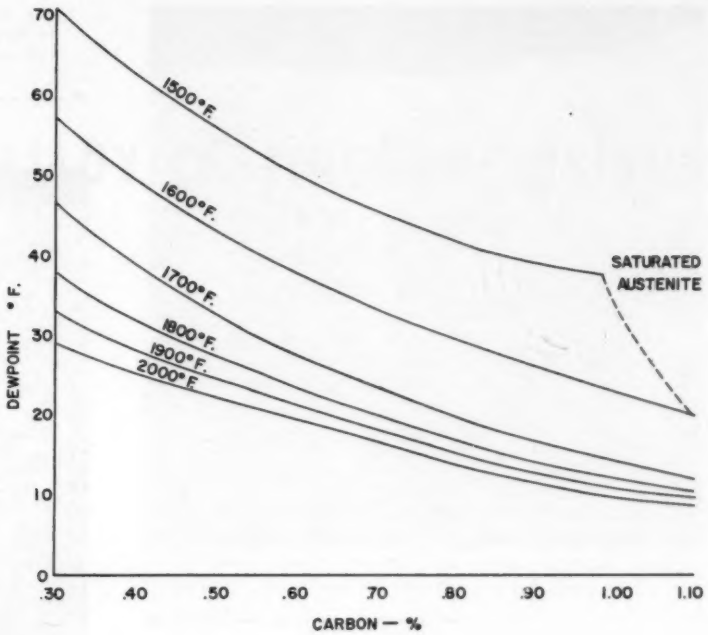



Fig. 10 — Typical available carbon vs. dewpoint curves.



Fig. 11 — 6150 alloy steel casting, as-cast. 2 per cent nital etch. 250 ×.

Each of these protective gas atmospheres have one thing in common; each contains a certain amount of moisture. Of particular significance is the fact that this amount of moisture is capable of being controlled. Also, since gas atmospheres under equilibrium conditions react according to exact physical chemistry laws, the measurement and control of one of the constitutents (in this case the moisture content) will control the whole atmosphere.

The thermodynamic law also reveals that for a particular alloy composition and specific temperature, an atmosphere can be either oxidizing or reducing, carburizing or decarburizing. These laws, then, allow the heat treater to measure and control the moisture content of the atmosphere and in turn obtain bright heat treated castings. This type of control also provides a tool to maintain or add carbon to the surfaces of the steel castings during heat treatment.

Furnace manufacturers, through years of laborious testing, have given heat treaters a family of curves of moisture content (measured as dewpoint) versus per cent of available carbon. These are based on endothermic atmospheres of approximately 20 per cent carbon monoxide, 40 per cent hydrogen and less than one per cent methane for various heat treat temperatures. The family of curves, shown in Fig. 10, represents an average of values taken from curves developed by four different furnace manufacturers.

These curves are used throughout the industry for carburizing, carbon restoration and neutral hardening operations. The carbon content of the casting, together with the available carbon in the endothermic atmosphere at the particular temperature, produces what is known as the "carbon potential" of the atmosphere.

## CARBON RESTORATION

Castings of medium and high carbon contents, in particular, are prone to decarburize in the molds as they solidify. To withstand those cyclic stresses during service, the carbon must be restored. For example, the microstructure of a low alloy steel casting, S.A.E. 6150 (carbon content – 0.50 per cent), in the as-cast condition (Fig. 11) reveals that this casting was decarburized. The sampled casting, along with the balance of the castings on the order, were then carbon restored, quenched and tempered.

The hardened sample was then normalized to aid the microscopic examination. The microstructure of

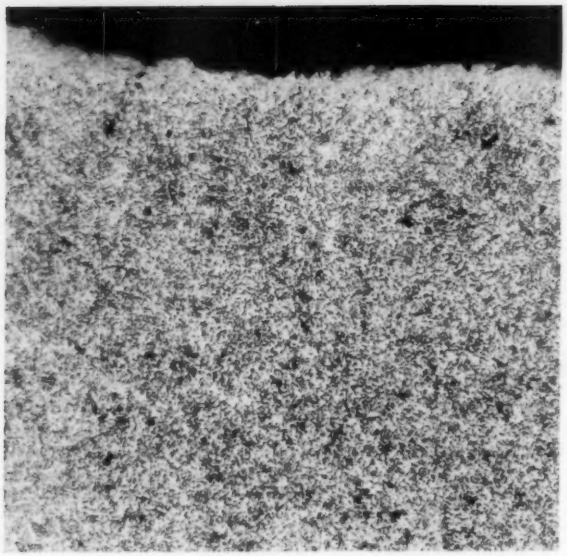


Fig. 13 — 3140 alloy steel casting, as-cast and hardened. 2 per cent nital etch. 250 ×.

this carbon restored, normalized S.A.E. 6150 casting is shown in Fig. 12. Another low alloy steel casting, S.A.E. 3140 (carbon content – 0.40 per cent), which had been hardened without carbon restoration, revealed a decarburized surface layer upon being examined microscopically. Figure 13 shows this decarburized casting. The castings, along with the sample, were given a carbon restoration heat treatment and quenched.

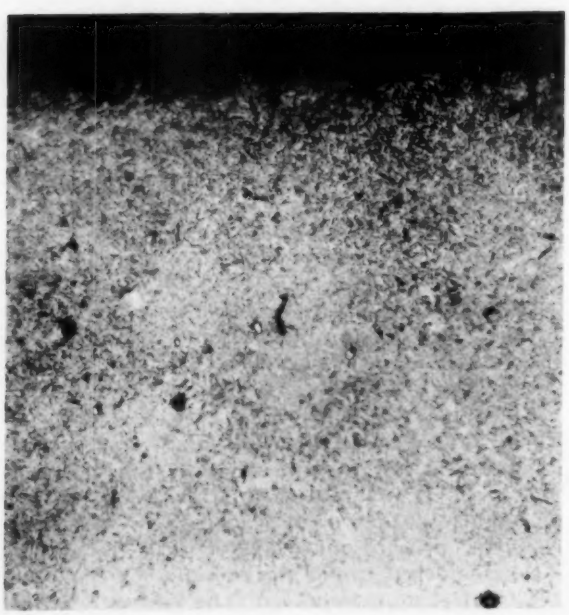
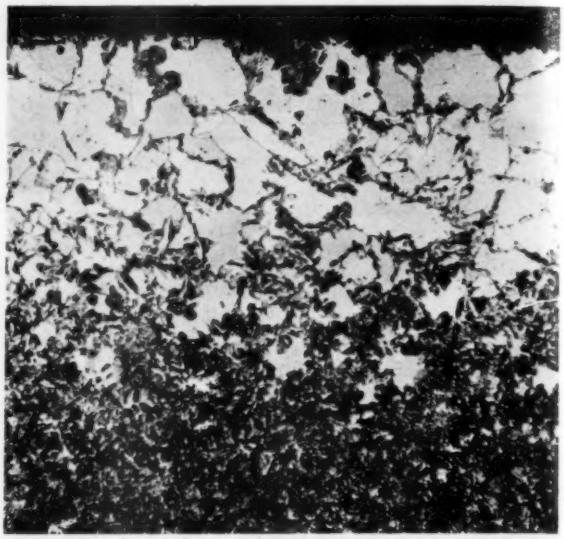


Fig. 12 — Same 6150 alloy steel casting as in Fig. 11, carbon restored, hardened and normalized. 2 per cent nital etch. 250 ×.



The photomicrograph of the sample casting after carbon restoring and hardening is shown in Fig. 14. Endothermic gas, containing approximately 20 per cent carbon monoxide, 40 per cent hydrogen and less than one per cent methane, was used as the protecting atmosphere during these carbon restoration treatments. The available carbon in these gas atmospheres, using the curves in Fig. 10 as a guide, were dewpoint (moisture content) controlled by instrument.

The controlled dewpoints at the 1700 F (926.7 C) heat treat temperatures were +33 F (+1 C) for the S.A.E. 6150 steel castings and +39 F (+4 C) for the S.A.E. 3140 steel castings. These carbon restored castings in the quenched and tempered condition produced residual compressive stresses on their surfaces which provided improved fatigue life. Design, foundry practice and heat treatment – these three, must contribute equally in the production of steel castings capable of withstanding the alternating stresses demanded by today's aircraft and missiles.

Foundry engineers and designers have progressed further in the recognition of stress raisers and their control than have heat treaters. However, commercial instrumentation for automatic atmosphere control and its contribution to the elimination of stress raisers is available. The steel foundry heat treaters, through the use of controlled gas atmospheres, can then aid in the needed fatigue life improvements.

Fig. 14 — Same 3140 alloy steel casting as in Fig. 13, carbon restored and hardened. 2 per cent nital etch. 250 ×.

# ALUMINUM ALLOY CASTINGS INCLUSIONS

by J. M. Fox


## ABSTRACT

**Inclusions in aluminum alloy castings are defined as foreign particles of any size mechanically entrapped within the solid metal. In the well-controlled aluminum foundry operation they do not constitute a serious problem, but when they are encountered they may affect machinability, surface appearance, mechanical properties, corrosion resistance, pressure tightness and wear resistance, of the castings. Inclusions may be found by machining, polishing, x-ray inspection or fracturing, after which they must be identified and corrective action taken. Inclusions can be oxide films, oxidation products, reaction products or may be introduced through mold coatings and mold materials. Adherence to good foundry practices should minimize or eliminate inclusions in aluminum castings.**


## INTRODUCTION

Inclusions in metals have attracted the attention of many investigators, and several definitions have been proposed to describe them. For a discussion of inclusions in aluminum alloy castings, the following definition is suggested: *Inclusions are foreign particles of any size mechanically entrapped within the solid metal.* The word *foreign* specifically describes the nature of inclusions in aluminum castings, as they are always physically or chemically different from any structural feature normally present in the metal. A generic definition of inclusions in metals is given in the AFS GLOSSARY OF FOUNDRY TERMS.[1]

In a well-controlled, aluminum foundry operation, inclusions in castings do not constitute a serious problem, and on a current industry-wide basis, inclusions cause relatively few difficulties. However, in instances where inclusions are encountered, they may be found to affect some of the following useful properties of aluminum castings:

1. *Machinability.* Inclusions may affect machinability by causing undue tool wear and a poor surface.
2. *Surface appearance.* The effect of inclusions on surface appearance is an obvious one. Even small inclusions may be noted on buffed surfaces, and they are greatly emphasized by anodizing.
3. *Mechanical properties.* Inclusions may affect the mechanical properties of castings if they are sufficiently numerous and of large enough size to reduce the effective cross-sectional area of sound metal.

Sachs, Dana and Ebert[2] reported that the magnitude of reduction in tensile properties was somewhat greater than the reduction of the cross-section by the inclusions would indicate, and concluded that inclusions provided a notch effect. Howell, Stickley and Lyst[3] reported an effect on fatigue strength which was greater for inclusions at the edge of test specimens than for those at the center. The also found that linear or "ribbon-shaped" inclusions had an effect on fatigue strength when they extended in a transverse direction, but had no effect when parallel to the direction of loading.

4. *Corrosion resistance.* If flux products or materials high in chlorides are entrapped in aluminum castings, they may cause a localized chemical attack in the presence of moisture. If an inclusion is metallic, it is conceivable that an electrolytic cell could be established between it and the surrounding casting resulting in galvanic corrosion.
5. *Pressure tightness.* It is not unusual for porosity to be associated with an inclusion, and if such a condition were to provide a continuous path of unsoundness through the cross-section of a casting, it would permit leakage of contained liquids or gases.
6. *Wear resistance.* Large inclusions in aluminum castings may be so hard and brittle that, if they were present on a surface in motion against another part, they might fragment into small abrasive particles and accelerate wear.

## IDENTIFICATION METHODS

Inclusions in unmachined castings are not likely to be detected by visual inspection, since they are usually located beneath the cast surface. They are most frequently found by machining, polishing, x-ray inspection or fracturing. Figures 1 to 4 illustrate the appearance of some typical inclusions when they are revealed by the various foregoing procedures.

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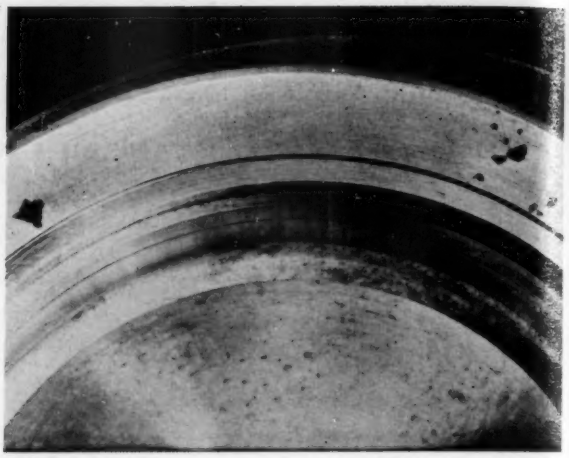


Fig. 1 — Dross inclusions on the machined surface of an aluminum casting.

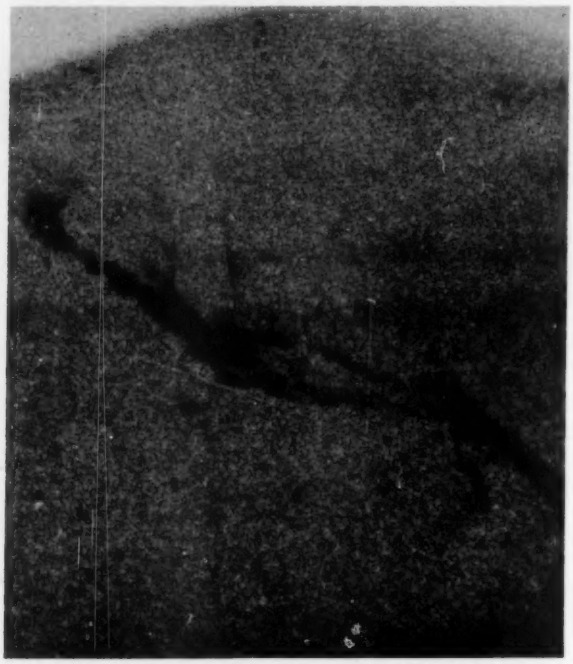


Fig. 2 — Oxide films in a radiograph of an aluminum casting.

After detection of inclusions, it is necessary to identify them, to determine their cause and to take corrective action in the foundry. Several methods of identification are available, and most of them require that an inclusion be first opened to view. A visual or x-ray examination may reveal sufficient characteristics of an inclusion to permit its identification. Dross, oxide films, sand and gross objects in general can usually be readily identified visually. Color, hardness, shape, magnetism and reaction to chemical spot tests are some characteristics which can be used in conjunction with a visual examination.

Microscopic examination can expand the limits of visual examination. Inclusions can be identified by optical methods, and both low and high power microscopes are useful. Figures 5 to 8 illustrate the



Fig. 3 — Comet tails on a buffed surface caused by spinel inclusions in an aluminum casting.

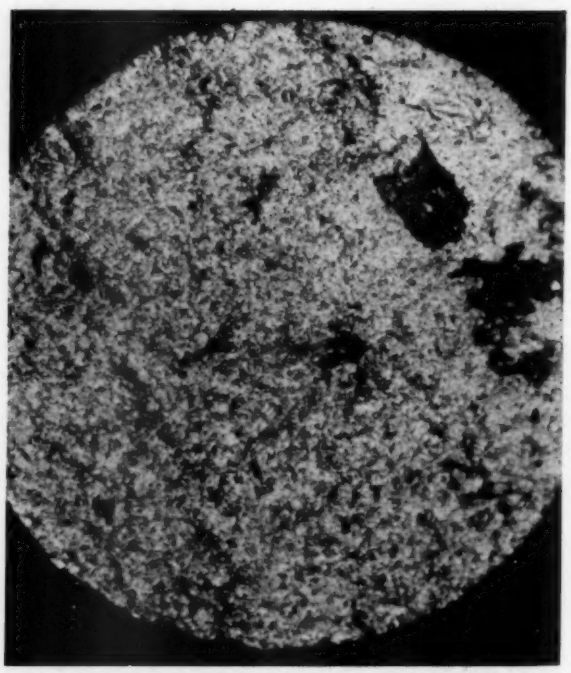


Fig. 4 — Dross inclusions in a fracture of a cast aluminum test bar. 7 ×.

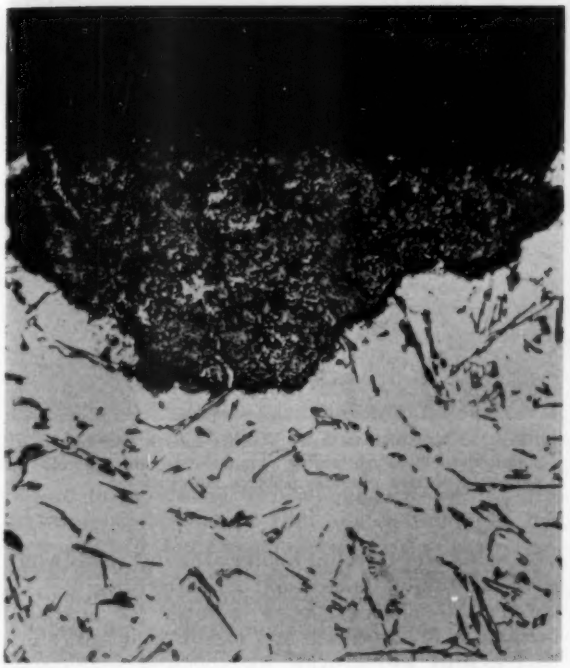


Fig. 5 — Nonmetallic inclusions in an alloy 356 casting. 100 ×.

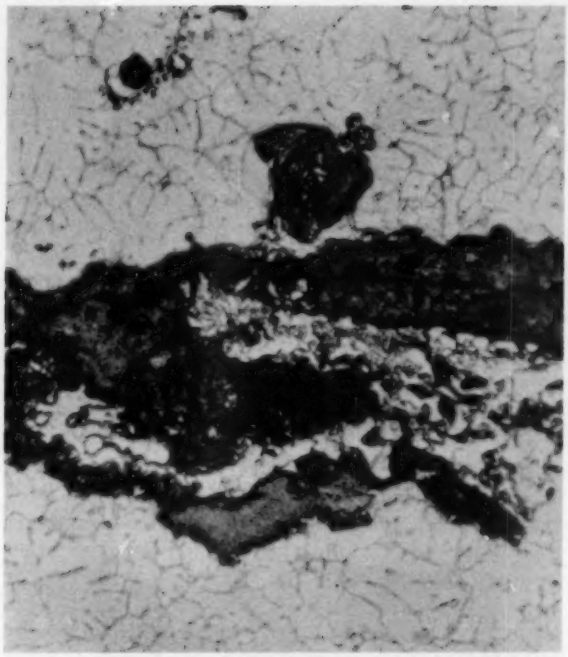


Fig. 6 — Corundum, spinel and magnesium oxide in an aluminum-magnesium alloy casting. 100 ×.

microscopic characteristics of some inclusions which have been found in aluminum castings. With higher power microscopes (500× magnification), distinctions between metallic and nonmetallic particles and observations of more than one material in an inclusion can be made. After metallographic polishing, metallic inclusions can be etched to reveal their microstructures and permit more exact identification.

Nonmetallic inclusions can be identified in many instances with a metallurgical microscope, but generally, the identification is more difficult than that for metallic inclusions. Quartz, graphite, dross and



Fig. 7 — Al-Fe-Mn-Si segregate in a sample of sludge from an alloy 380 melt. 100 ×.

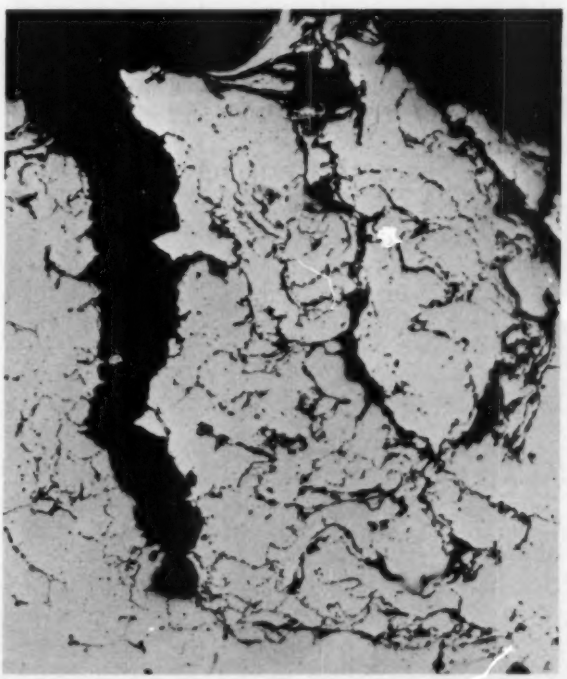


Fig. 8 — Oxide films in the microstructure of an aluminum-magnesium alloy casting. 100 ×.

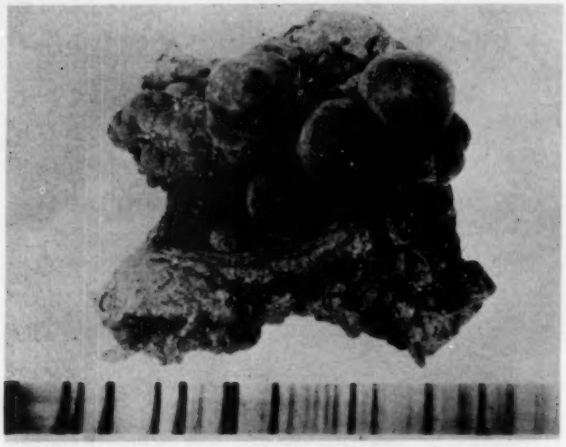


Fig. 9 — Corundum formed in open hearth melting and identified from its characteristic x-ray diffraction pattern.

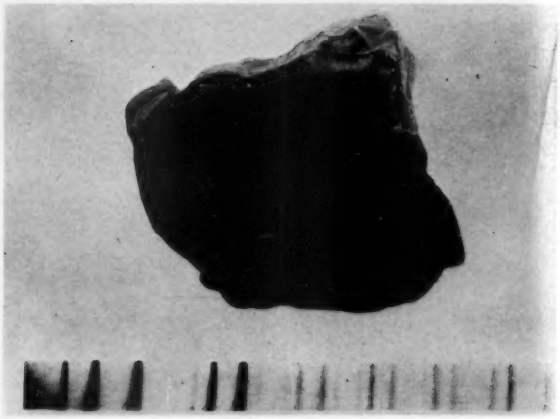


Fig. 10 — Spinel formed in open hearth melting an aluminum alloy containing magnesium. The x-ray diffraction pattern provides positive identification.

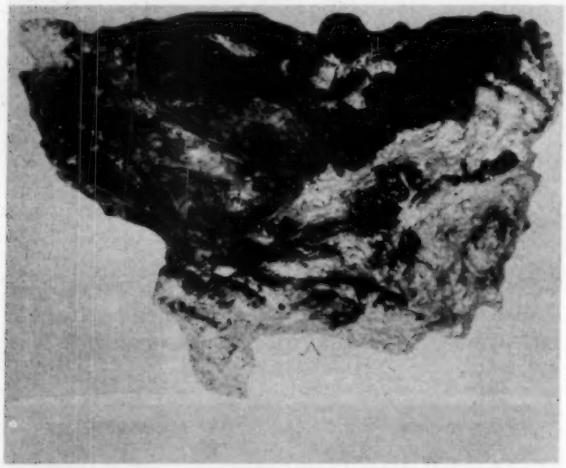


Fig. 11 — Dross accumulation from the upper wall of a refractory crucible.

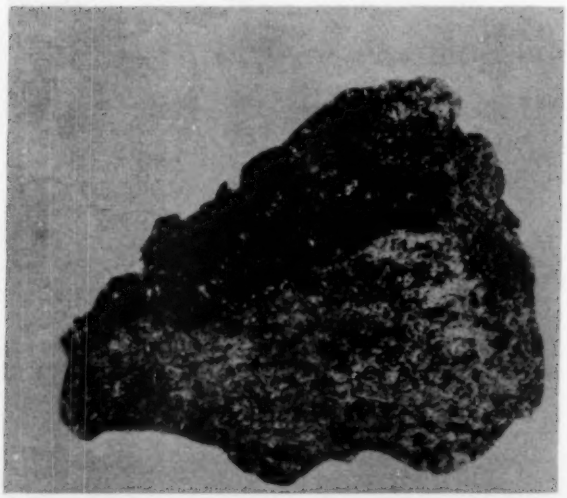


Fig. 12 — Sludge removed from the bottom of a die casting holding furnace.

iron oxide are typical of nonmetallic inclusions which have identifiable microscopic characteristics.

Chemical analyses by a number of analytical methods may be utilized as a means of identification. Such results will provide the chemical composition of an inclusion, but will not reveal the combinations in which the detected elements are present. Despite this limitation, the chemical method may be completely satisfactory in many cases. Often a test for chlorides will be sufficient when added to the results of a visual examination to identify an inclusion as a flux product. This and the spot tests mentioned earlier illustrate how the chemical method can be used in a simple way to permit reliable identifications.

Should the visual, optical or chemical methods leave any uncertainties, they may be resolved usually by an x-ray diffraction analysis. This method provides the most satisfactory and positive means of investigation since it identifies chemical compounds and the relative proportion of components, should more than one material be present in an inclusion. It is limited only by the requirement that the inclusion be crystalline and, to some extent, by the size of the specimen to be identified. X-ray diffraction analysis could be called the referee method when visual examination is not sufficiently positive.

## SOME COMMON INCLUSIONS

Figures 9 to 12 are photographs of four inclusion forming materials which can be generated in the foundry by careless melting practices. When practices are unsatisfactory, these materials may be found in melting furnaces or in holding pots, and are frequently ladled into casting molds.

**Dross or aluminum oxide.** This is perhaps the most common inclusion found in aluminum alloy castings. It forms naturally and readily as a film at interfaces between molten aluminum and air. Dross is identified by its unique visual, optical and radiographic appearance. It will most often appear as an individual film or a cluster of films. Dross inclusions will range in color from light gray to black, depend-

ing on factors such as film thickness and associated impurities.

Dross films can block feeding and frequently they are found to be associated with localized shrinkage porosity. Sometimes dross is formed by turbulent pouring and is found associated with entrapped air voids. Should dross be uncovered by machining, it will usually appear on the machined surface as a lacy area often with some localized porosity.

**Corundum.** This inclusion is chemically identical to dross, occurs in a massive and crystalline form. It is designated alpha alumina. Corundum can be formed by oxidation of melts at high temperatures by agglomeration of dross or by the decomposition of refractories. It can be found on the inside of various melting units. Stalactites hanging from the roofs of open hearths and deposits on the inside of crucibles have been found to contain corundum.

With the advent of high alumina refractories, better resistance to reaction with molten aluminum has been achieved, but mechanical damage to the refractory might still provide a source of alumina inclusions. These will appear as solid homogeneous masses.

**Spinel, or magnesium-aluminum oxide.** This is a nonmetallic compound formed under similar conditions to those for corundum. It is crystalline and less dense than corundum. Spinel generally forms in melting units by the oxidation of aluminum alloys containing magnesium.

**Magnesium oxide.** This is often found associated with spinel in deposits accumulated on the inside surfaces of melting units, wherein aluminum alloy heats containing magnesium were melted. Usually, this material is gray, although it can also be quite dark.

**Aluminum nitride.** This inclusion has also been found as a hard nonmetallic deposit on the inside surfaces of melting crucibles, and in the melting channels of induction melting furnaces. In the majority of cases, it will probably be associated with some or all of the oxidation products listed previously. It readily decomposes with the evolution of ammonia when the inclusion is powdered or wet.

**Chlorides.** Chlorides such as aluminum chloride, magnesium chloride, zinc chloride and potassium chloride can be included in aluminum castings, and their origins usually can be traced to solid or gaseous fluxes. These chlorides are usually white in color. Some are deliquescent so that, although they may not be large, their presence may be manifest as dark spots on casting fractures after a short exposure to moist air. As a group, their densities are all less than that of aluminum, and their melting points are within the melting range of aluminum alloys. Flux inclusions can cause obvious chemical attack in aluminum castings in the presence of moisture, and this can be a clue to their identity.

**Refractories.** Furnace refractories are a potential source of inclusions. Fireclay refractories usually consist of mixtures of mullite and alpha cristobalite or silicon dioxide and can react with molten aluminum to form corundum. Deterioration and subsequent spalling of refractories may cause their inclusion in the melt. Pieces of refractory may also be dislodged by the impact of charging or furnace tools.

**Crucibles.** Crucibles can become damaged and introduce inclusions into castings. Plumbago, silicon-carbide and iron oxide scale are typical pot materials which can be sources of inclusions. Small amounts of metallic iron dissolve readily in molten aluminum when iron pots and furnace tools are used. Excessive solution of iron forms sludge in furnaces and inclusions in castings.

**Silica or alpha quartz.** These can form inclusions from green or core sand parts. Occasionally, sand may react with the metal if any magnesium is present in the alloy. This can be observed microscopically by the presence of (Mg-Si) constituent segregation around the inclusion. The physical appearance of a sand inclusion is readily recognized.

**Wash coatings.** These coatings may break away from tools, crucibles or molds to form inclusions. Calcium carbonate, kaolin, graphite and hematite may occur as relatively soft inclusions and be traced back to wash coatings.

All of the preceding inclusions are either metal oxides or nonmetallic compounds. Metallic inclusions are not as likely to occur in aluminum castings, because molten aluminum readily dissolves the common metals. However, some metallic elements, particularly iron, manganese and chromium, form intermetallic compounds which have high melting points. If proper practices are not observed during melting and holding, it is possible for these high melting point intermetallics to solidify and segregate in the melt.

Because of their specific gravity, they will sink to the bottom of the melt to form what is described as sludge. When portions of this sludge are then ladled out with the melt, they can occur as inclusions or hard spots in castings. The formation of sludge is most likely to be encountered in die casting operations where it is common practice to hold molten metal for relatively long periods at low temperatures.

## PREVENTION METHODS

### *Oxide Films*

Films of aluminum oxide or dross inclusions are the most common inclusions found in aluminum castings, because an oxide film will form immediately on any freshly exposed surface of the molten metal. The principle to observe in avoiding these films, therefore, is to minimize them by disturbing the surface as little as possible during melting, pouring and mold filling from the gates. Films or particles which are entrapped within the liquid metal in a furnace may be removed by fluxing and subsequently carefully skimming them off the surface.

After a good job of cleaning a melt has been done, it is equally important to avoid formation of new films during transfer, pouring and entry of the metal into the mold. All metal movement subsequent to fluxing should take place with a minimum of agitation or turbulence. During transfer or pouring, the molten metal should not be permitted to drop vertically from any greater height than is absolutely neces-

sary. Pouring boxes, sprues, runners and gates enter into metal movement control.

AFS sponsored research has demonstrated effective means of pouring a casting through proper gating. This work is presented on motion picture film and is available from AFS. In some cases, stainless steel mesh, perforated sheet screens or steel wool can be utilized at gates to screen out inclusions.

### *Oxidation Products*

Corundum and spinel are both high temperature products of oxidation involving reactions with air or refractories. The lower the melting temperature, and the shorter the time that metal is held in the molten state, the less will be the tendency to form these two types of inclusions. Both these products will be observed as buildup on the furnace wall or sides of crucibles at the metal line. Any metal splashed on the refractory above the metal line will also form this buildup.

Stalactites which can form on the roofs of open hearths result from splashed metal. Actually, it is difficult to avoid some buildup during continuous melting operations. The practical objective is to remove the accumulated material from the refractories or crucibles as often as possible. This will aid in minimizing corundum and spinel inclusions. Scraping and cleaning of melting units should always be done at the end of a casting run and while the buildup is hot and soft. Cleaning during casting operations will introduce corundum into the melt and castings being poured.

If the furnace is cold, the refractories themselves can be damaged. Scrupulous attention to cleaning out of buildup is a must in low frequency induction furnaces. If the channels in these furnaces are not cleaned frequently, and in some cases as often as every heat, they will become blocked and prevent operation of the furnace. As mentioned earlier, the replacement of fireclay type of refractories by the newer high alumina refractories will provide better resistance to metal-refractory reaction and help to reduce corundum and spinel formation.

### *Reaction Products*

Inclusions originating as reaction products from fluxes are generally less dense than molten aluminum so they should be easily skimmed from the surface, if given adequate time to rise to the surface. A holding period of approximately 15 min in commercial size melts should be sufficient to accomplish this. The fluxing action also can carry heavier particles to the surface where they will adhere to the surface oxide and be removed in skimming.

Metallic inclusions can be avoided by prevention of sludge formation in melting or holding and by prevention of the accidental introduction of metallic objects to the metal during casting operations. Colwell and Tichy[4] have described the mechanics of sludge formation, and recommended mixing during melting and maintenance of adequate temperatures during holding as the means of prevention. Bonsack[5] has recently presented a comprehensive discussion of melting practices, including precautions to avoid inclusions.

Other possible inclusions are those introduced from coatings and mold materials. These constitute the least frequently observed class of all inclusions. They are prevented by careful practices. Maintenance of good wash coatings, prevention of damage to refractories, proper mold construction to prevent pieces of molds themselves from breaking away and good housekeeping to keep everything as clean as possible in the foundry are the measures of prevention.

## SUMMARY

Although inclusions can be present in castings in such a manner that the serviceability of the castings is impaired, they can also be present in the interior of noncritical areas, or in locations where they will be removed with no adverse consequences. It would be uneconomical, therefore, to arbitrarily assign castings in which inclusions are detected to the scrap pile. Such an assignment may be necessary, but it should be done intelligently.

The foregoing discussions of the types and compositions of inclusions will assist in their identification and elimination. The practical significance to the problem of inclusions is that every effort should be made to avoid them. Actually, the effort required is simply that of maintaining good foundry practices. Melting equipment, tools, ladles and molds should be kept clean and free of accumulated buildup. Metal to be charged should be clean and free of corrosion and foreign matter. Charging operations should be done so as to avoid molten metal splashing.

Control of iron, manganese and chromium analysis and metal temperature should be a part of the practices used to avoid formation of sludge. Fluxing and skimming, followed by a settling period and additional skimming, should be employed. Splashing and turbulence should be avoided during molten metal transfers. It would be preferable to maintain minimum holding temperatures and to avoid unnecessarily long holding times. Proper pouring technique should involve—final metal skimming, careful pouring into pouring boxes and proper design of sprue, runner and gating systems. Obviously, foreign material should be excluded from the mold or die cavity.

Adherence to all of these practices should minimize or eliminate inclusions in aluminum castings.

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# BETATRON RADIOGRAPHY OF HEAVY SECTION CASTINGS

by L. J. Venne


## ABSTRACT

**The Radiographic Department at the author's company has a betatron, a 220 kv x-ray machine, two 10 curie sources of Cobalt 60 and a 30 curie source of Iridium 192. This wide selection of radiographic sources allows for examination of practically any section thickness of casting that might be produced. This discussion covers the role of the betatron in the field of nuclear energy castings.**


## INTRODUCTION

The author's company first became interested in nuclear energy castings in 1954. In this year a prototype for a land based power station was produced. In 1956 the first nuclear energy shipboard casting was produced. This was for submarine service.

Radiographic work on these first castings had to be done with small Cobalt sources. With 100 per cent radiographic examination required on these comparatively heavy sections, time became an important commodity. The examination of wall junctions particularly was time consuming. The resulting radiographs were generally inferior.

As production requirements grew, it quickly became evident that the cobalt sources would not meet the requirements of anticipated orders and deliveries. From experience it was felt that the cobalt 60 source available had a maximum practical thickness range of slightly over 6 in. at the sensitivity that would meet the requirements of applicable specifications. Some of the sections requiring radiography approached 12 in. The need of a high energy fast source of radiography also became evident when a total of 8000 hr of gamma exposure for the cobalt sources was evident on the company's backlog.

After considering all of the facts on available equipment and considering present and future requirements it was decided to purchase a 24 MEV betatron. Ease of handling and speed of exposure along with good sensitivity were important factors in this decision. The betatron was ordered in Feb. 1957 and put into operation in Sept. 1957.

One other situation that pointed out the urgent need for a betatron was the fact that there was no high energy equipment privately owned located west of the Mississippi at that time. The only equipment that could be used on a noninterfere basis required that the castings be for the end use of the Navy. It was located at the Arctic and Submarine Research branch of the Navy Electronics Laboratory at San Diego, and the installation was not compatable with casting radiography. Up to the time of installation of the betatron in Portland the company was permitted to use this equipment.

## CONSTRUCTION PROBLEMS

After the decision of the management of the author's company to purchase the betatron several basic construction problems had to be resolved.

It was obvious from the outset that the betatron facility should be as close as possible to the cleaning room so as to eliminate costly and time consuming transportation. Cost of construction for housing this high energy radiation had to be considered. In addition, the structure had to be able to support equipment to handle the extremely heavy castings being ordered for the nuclear energy program.

After much discussion and planning it was decided to house the betatron in an open pit. An existing cleaning building was extended to accommodate a pit 60 ft long and 25 ft wide. The pit would be 15 ft below ground level with a wall extending 5 ft above ground level and 3 ft thick to completely surround the open portion of the pit which measures 25 x 25 ft. Actually, the only floor space lost to manufacturing processes would be this 25 ft square opening. The remaining portion of the 60 ft length has been decked over with a 3 ft thick slab of concrete. This apron is used with complete safety by welding stations. The actual shooting area of the betatron room is 25 x 40 ft. The shooting area is separated from the control area by a 3 ft thick concrete bulkhead.

Besides the cost savings and safety features of the pit construction, it is felt that the location gives in effect a production station where radiography is a step in processing the casting as are the remaining stations in the building. An additional advantage is the fact that the existing overhead crane is used to deliver castings to and from the pit. This same crane can service the production stations in the balance of the building.

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Probably the best criterion for evaluation of the betatron is the fact that over 1200 castings have been produced to nuclear energy requirements through 1960. Of these at least 250 were large castings of heavy cross-section that would have been difficult if not impossible to produce and ship using the small Cobalt 60 sources used at the inception of the program.

## EFFECTIVE BETATRON USE

The betatron is effectively used as a quality control tool in several different ways.

**To help develop molding practices on heavy castings.** The speed of the betatron unit is an ideal tool to quickly check the quality of heavy castings. It is regularly used by the planning group on sample castings and first run castings. The results are the basis for changes where necessary. Sample and first run castings are given preference in scheduling, and the results are reviewed as quickly as possible with the planning committee.

**To check for defect removal in heavy sections.** It is often desirable to check a defective area prior to welding to ascertain that a defect has been completely removed prior to weld repair. The betatron serves a fine purpose in this regard, and the location of the unit permits this to be done with a minimum of interference.

**To check the quality of weld repair prior to final processing.** It is sometimes desirable to determine the quality of a weld repair before a casting is completely ready for shipment. In some cases this step can eliminate the necessity for additional heat treatment. Here again the speed and location of our unit is an advantage.

**For final radiography in accordance with customers requirements.** I am sure you are all acquainted with types of defects found by radiography. The table shows an exposure time comparison, while Figs. 1 to 10 show a series of castings produced by the author's company for nuclear energy application.

ROUGH COMPARISON OF EXPOSURE TIME BETWEEN BETATRON AND COBALT 60

| Steel, in. | Betatron min | Betatron sec | 100 Curie Cobalt 60 hr | 100 Curie Cobalt 60 min | 100 Curie Cobalt 60 sec |
|---|---|---|---|---|---|
| 2 | | 12 | | | 30 |
| 3 | | 12 | | 2 | 20 |
| 4 | | 18 | | 9 | |
| 5 | | 30 | | 32 | |
| 6 | | 48 | 1 | 40 | |
| 7 | 1 | 18 | 5 | 10 | |
| 8 | 2 | | 15 | | |
| 9 | 3 | 18 | 43 | | |
| 10 | 5 | | 98 | | |
| 11 | 8 | 48 | 130* | | |
| 12 | 14 | 24 | 170* | | |
| 13 | 25 | 24 | 202* | | |
| 14 | 37 | 42 | 260* | | |
| 15 | 62 | 12 | 280* | | |

*Theoretical times only in this range.

The approximate weights and metal sections will give you some appreciation of the size of the casting. The times for betatron radiography will allow you

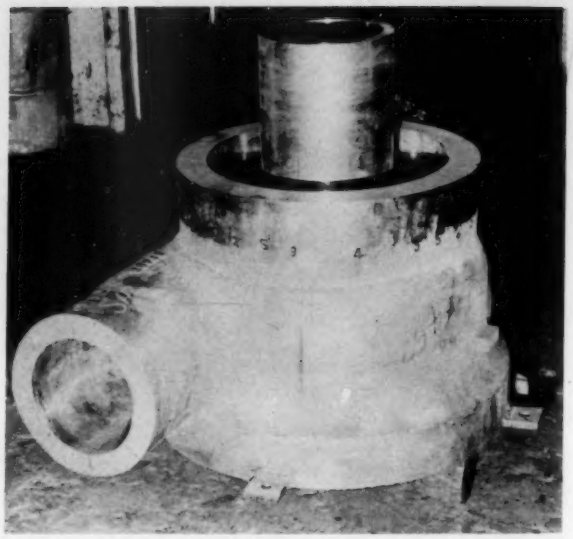


Fig. 1 — Weight 3000 lb. Nominal thickness 2⅞-in.; heaviest section 6 in. Betatron exposure 6 hr; iridium exposure 2 hr.



Fig. 2 — Weight 8000 lb. Nominal thickness 4¾-in.; heaviest section 8 in. Betatron exposure 8 hr.



Fig. 3 — Weight 8300 lb. Wall thickness 4 in.; heaviest section 8 in. Betatron exposure 16 hr.



Fig. 4 — Weight 8300 lb. Wall thickness 4 in.; heaviest section 8 in. Betatron exposure 16 hr.

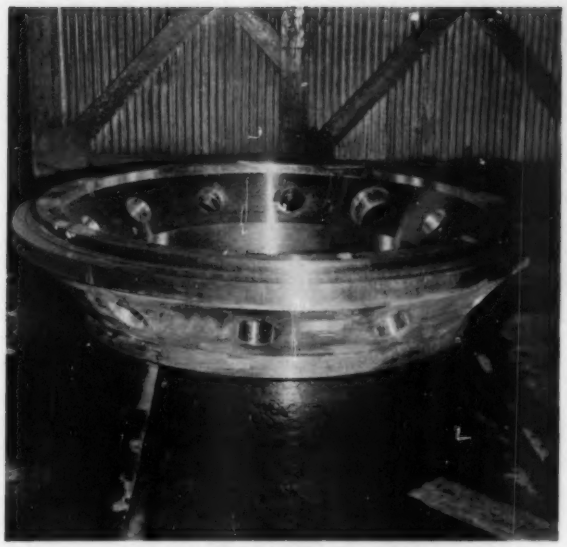


Fig. 5 — Weight 15,200 lb. Average thickness 5 in.; heaviest section 11 in. Betatron exposure 20 hr.

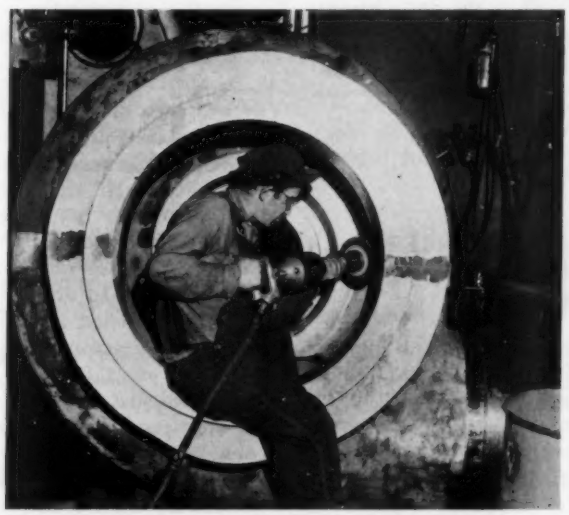


Fig. 6 — Weight 16,000 lb. Nominal thickness 5½-in.; heaviest section 14 in. Betatron exposure 24 hr.

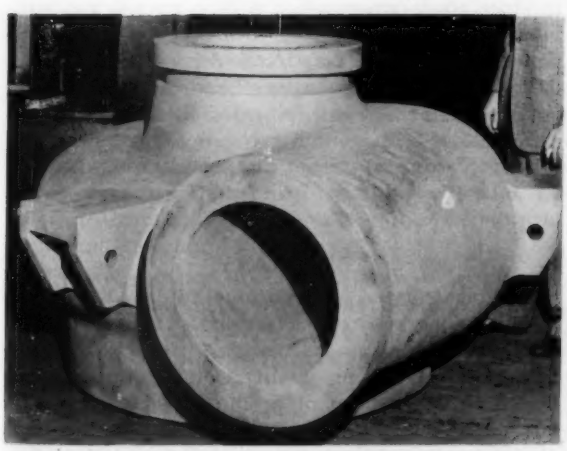


Fig. 7 — Weight 16,000 lb. Nominal thickness 5½-in.; heaviest section 14 in. Betatron exposure 24 hr.

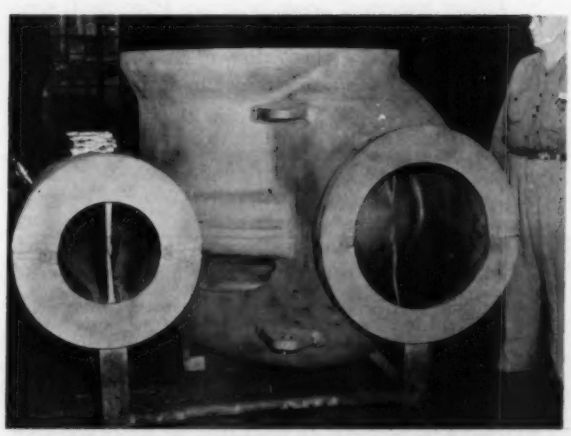


Fig. 8 — Casting made in 2 sections and welded. Weight 16,000 lb. Wall thickness 1¼-in.; heaviest section 5 in. Betatron exposure 6 hr; x-ray or iridium 20 hr.

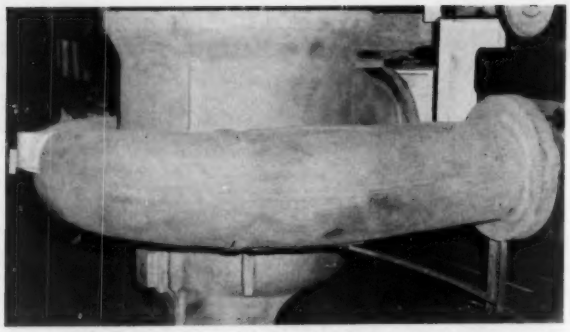


Fig. 9 — Weight 16,000 lb. Wall thickness 1¼-in.; heaviest section 5 in. Betatron exposure 6 hr; x-ray or iridium 20 hr. Casting made in 2 pieces and welded.

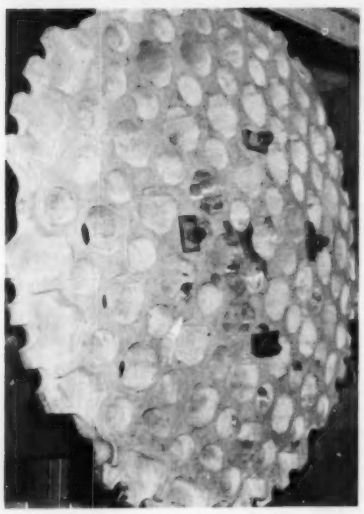


Fig. 10 — Weight 9000 lb. Heaviest section 12½-in.; diameter 100 in. Betatron exposure 20 hr.

to compare in what a comparable time with other sources of radiography might be. A number of points may be of interest on the use of the betatron for nuclear energy castings. While these concepts are usually the same for most castings, they apply particularly to the specific requirements of nuclear energy castings.

Section thickness limitations

A. Lower practical limit is 2 in. The short wave length of betatron does not register contrast under 2 in. as well as other means.
B. Intermediate sections from 4 to 10 in. show wide latitude.
C. In 10 in. and up to 20 in. sections the latitude falls off.

Film type and practice

A. Double film is standard practice; this is practical in betatron o/o as there is no need for a center screen.
B. The added cost of film it is felt is better than reshooting o/o film defects, and it helps take care of any processing or screen imperfections or inherent film defects.
C. Machine latitude is increased by allowing viewing singly for thin sections or doubly for thick sections, as permitted by some specs.
D. Normally use same speed film for both film of double film process.
E. Use fine grained medium speed film.

Sensitivity

A. Usually one per cent on all sections 4 in. to 16 in. 2 per cent for 2 to 4 in. sections.
B. Density usually 1.5 to 3.5. 1.5 may be double viewed.
C. Extremely small effective focal spot of betatron. 0.0005 sq in.

Screens

A. Usually 0.040 in. lead screen up to 10 thick front is used.
B. 0.010 - 0.020 in. lead screen back screen.
C. On sections above 10 in. use 0.080 in. front screen and 0.040 in. back screen.
D. Little scatter — use lead backing outside 10 in. up and if interfering section is near.

Cassettes

Use rigid cassettes, normally:

1. Intimate film screen contact.
2. Durable.
3. Speed of loading and unloading.
4. Structural support to hold film in place.
5. Increases life of screens.

The film source distance is rarely under 6 ft and is normally 9 ft, dependent on thickness and shape of piece being radiographed. The advantages of location of betatron are (1) preliminary shots prior to working a casting if necessary to determine quality can be obtained and (2) radiography during casting repair is speeded up. There are disadvantages of location of betatron. These are (1) cleaning room dust requires more frequent cleaning and the company is limited in casting size by pit opening.

The speed of operation of betatron allows defects or questionable areas to be quickly located and the radiography can be redone if necessary with little loss of time. Also, the fast results on sample castings can show where changes in molding, gating risering are needed. Less exacting calculations and skill to produce good radiograph o/o high energy and wide latitude are needed in betatron operation. The maneuverability of the betatron along with use of turntable and fixtures is valuable. The head can be raised and lowered, tilted and rotated. This type of radiography is safe with normal precautions, and it is comparatively easy to train people in its use. Another advantage is that there is little downtime o/o equipment failure.

## SUMMARY

The betatron has and continues to do a satisfactory job for radiography of nuclear energy as well as other castings. The operation is fast, sensitive and comparatively simple. Casting quality can be quickly determined in early stages of production if necessary. It has been a major tool in the author's company's ability to successfully compete in the field of nuclear energy castings for ships.

# DUCTILE IRON POST INOCULATION

by W. C. Jeffery and H. D. Bradshaw


## ABSTRACT

The production of ductile iron requires two separate treatments to obtain ductility and toughness in the as-cast condition. The treatments are the magnesium treatment and the addition of a graphitizing agent. The addition of the graphitizing agent is called post inoculation. The addition of a graphitizing element to ductile iron after the magnesium treatment is of great importance in controlling the mechanical properties of this type iron.

This investigation shows the effects of increasing the graphitizing agent addition after magnesium treatment and the effect of holding time after the inoculation. The graphitizing agents used in this investigation are 75 per cent ferrosilicon and calcium silicon containing 62.5 per cent silicon. For maximum ductility in the as-cast condition it is essential that the magnesium bearing ductile iron be inoculated with up to one per cent silicon as a post inoculant.


## INTRODUCTION

In producing ductile iron the importance of the post inoculation treatment with a suitable inoculant such as ferrosilicon or calcium silicon has often been minimized or overlooked. From previous experiments it has been noted that a pronounced effect on the microstructure had been produced by the post inoculation procedure.

This investigation was undertaken to determine the effect of adding increasing amounts of the inoculant to a magnesium containing iron. It was also an objective to determine the effect of holding them after the graphitizing inoculant was added to the molten magnesium treated iron. The term post inoculation will be used to indicate that a graphitizing agent was added to the iron after the magnesium treatment.

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## EXPERIMENTAL PROCEDURE AND EQUIPMENT

### *Melting*

The melting unit used in this investigation was a 350 lb indirect arc rocking electric furnace, lined with sillmanite. In this type furnace, heat to melt the charge is produced by radiation from the arc of approximately 120 degrees. The rocking motion is begun as soon as the charge has become molten. The rocking thoroughly mixes the metal and heat is absorbed from the furnace lining.

The charge consisted of 400 lb of low phosphorus pig iron of an analysis given in Table 1.

TABLE 1 — ANALYSIS OF PIG IRON

| | |
|---|---|
| C, % | 4.10 |
| Si, % | 2.00 |
| Mn, % | 0.27 |
| P, % | 0.03 |
| S, % | 0.013 |

The iron charge was heated to approximately 2850 F before tapping.

### *Magnesium Treatment*

The magnesium treatment was performed in a 100 lb ladle using the equipment shown in Fig. 1. One hundred lb of iron at 2800 F was tapped into a preheated ladle and treated with magnesium. Pure mag-

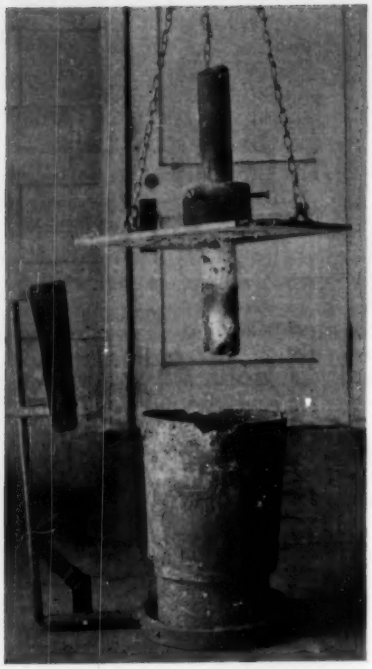


Fig. 1 — Inoculation equipment.

nesium was used to inoculate the iron. One-tenth of one per cent (45.4 gm) was placed into a graphite plunger and inserted into the iron.

The graphite plunger shown in Fig. 1 was constructed from a graphite electrode used in the indirect arc furnace. Details of the plunger are given in Fig. 2. The plunger had 1/8-in. holes drilled through to the magnesium chamber. The number of holes varied from two in the bottom of the plunger to two in bottom and six in the sides. The number of holes was varied with no change in results. The 1/8-in. hole was counter bored to 1/4-in. in diameter and 1/4-in. deep. The counter bore was plugged with a 1/4-in. magnesium rod.

The plunger with the magnesium charge was lowered into the molten iron and allowed to remain until all reaction had ceased. The time for the reaction to begin was approximately 75 sec and the reaction was completed in approximately 150 sec.

### *Post Inoculation*

The post inoculation was performed in a second 100 lb preheated ladle. The post inoculant was placed in the bottom of the ladle and the magnesium containing iron was poured directly onto the post inoculant. Ferrosilicon (75 per cent Si) and calcium silicon (63.5 per cent Si) were used as post inoculants. The ferrosilicon was 3/8-in. x 12 mesh in size. The calcium silicon was 8 mesh x down in size.

Various amounts of CaSi and FeSi were used to inoculate the magnesium treated irons. The amount and type of post inoculation are given in Table 2.

TABLE 2 — TYPE AND AMOUNT OF POST INOCULATION

| Heat | Type of Inoculant | Si Recovered |
|---|---|---|
| 4A | FeSi | 0.44 |
| 2B | FeSi | 0.48 |
| 3A | FeSi | 0.98 |
| 3B | CaSi | 0.34 |
| 3D | CaSi | 0.73 |
| 2A | None | |
| 5B | FeSi* | |

*FeSi added to furnace before tapping.

In heat 5B the silicon was added to the furnace to eliminate the inoculation effect, and to determine the relative value of the silicon when added as an inoculant and as added to the iron before the magnesium inoculation.

### *Pouring*

The ductile iron was cast in Y-block cores, as shown in Fig. 3. The first block was cast as soon as possible after the post inoculation. The time required for the first pour was approximately 30 sec. The remaining pours were of 3, 6 and 8 or 9 min. Heat 2B was poured at 9 min. Following the first pour a spectrographic sample was cast into a copper mold, illustrated in Fig. 3.

### *Test Specimens*

The test specimens were cut from a one in. x 6 in. Y-block test coupon, as shown in Fig. 4. The tensile

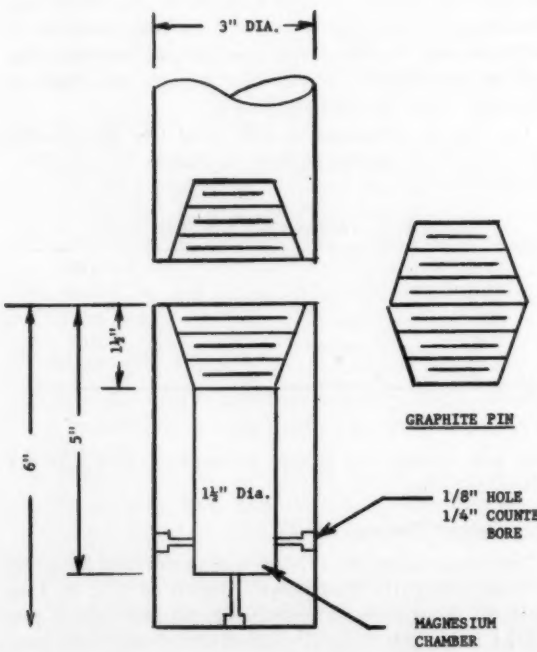



Fig. 2 — Cross-section of graphite plunger used for magnesium inoculation.

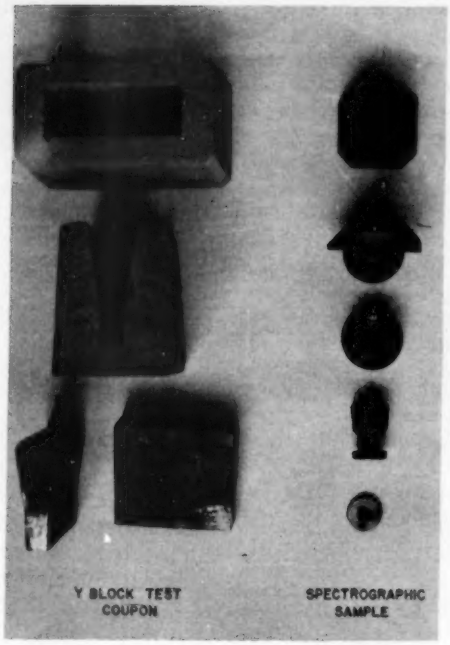



Fig. 3 — Test samples.

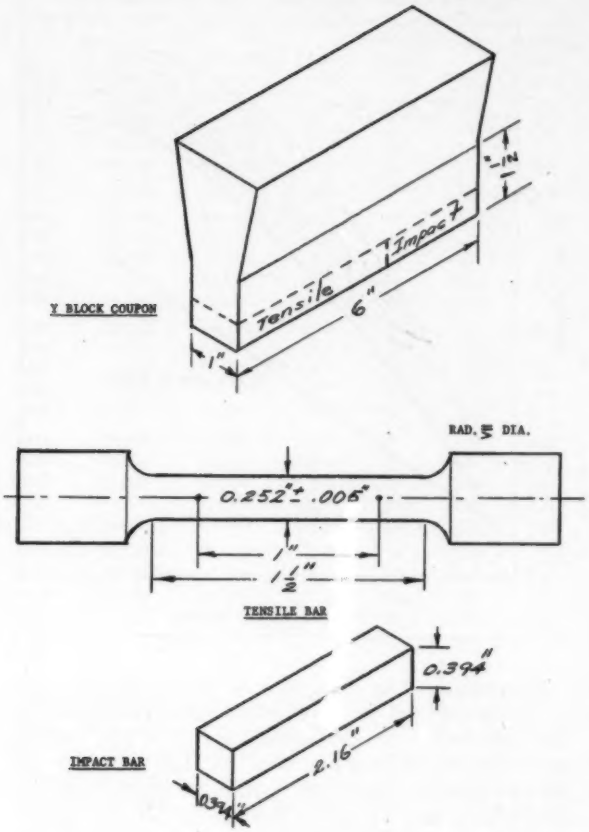



Fig. 4 — Test specimens.

bar was a standard tensile specimen 0.252 in. in diameter with a 1.0 in. gage length. The unnotched impact measured 10 x 10 x 55 mm.

### *Testing Physical*

Tensile strength, yield strength and elongation were determined on a standard tensile machine of 120,000 lb capacity. Brinell hardness was determined with a 10 mm ball loaded to 3000 kg. All other metallurgical tests were conducted in accordance with normally accepted procedures.

## DISCUSSION

The results of this investigation are reported in two sections. The first section shows the effects of the addition of silicon as ferrosilicon and calcium silicon to ductile iron after the magnesium treatment. The second section shows the effect of holding time after the magnesium and silicon addition on the properties of ductile iron.

### *Silicon Additions Effect*

The addition of silicon to ductile iron is important as all mechanical properties are, to a large extent, dependent upon the silicon content. However, the amount of silicon is not the sole determining factor in its effect upon the mechanical properties of ductile iron. The manner in which the silicon is introduced, the type of silicon alloy used and other factors are of great importance.

In this investigation silicon was added in the forms of ferrosilicon and calcium silicon. The silicon was added after the magnesium treatment, as described in the procedure section of this report. No addition of ferrosilicon was made to the furnace charge of heat 2A and no post inoculation was performed.

**Tensile Strength.** Silicon increases the tensile strength of ferrite approximately 11,000 psi for each per cent addition of silicon. Since many as-cast uninoculated ductile irons are essentially pearlitic, the addition of silicon as a post inoculant decreases the tensile strength. This decrease in tensile strength is produced by decreasing the pearlite and increasing the ferrite in the microstructure. In these experiments the addition of calcium silicon and ferrosilicon as post inoculants decreased the tensile strength of the as-cast ductile iron, as essentially ferritic structures were obtained with silicon additions approaching one per cent.

**Yield Strength.** Silicon additions to ferrite increase the yield strength in a manner similar to the increase in tensile strength. The strengthening of ferrite by silicon may be partially or completely compensated for by the decrease in yield strength due to the decomposition of pearlite. In this series of experiments the yield strength of ductile irons were generally decreased with increasing post inoculation.

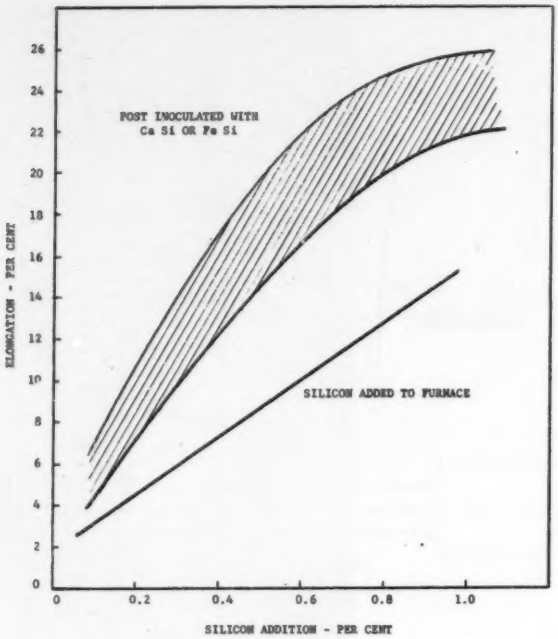



Fig. 5 — Silicon addition effect on as-cast elongation.

**Elongation.** The elongation of as-cast ductile iron is greatly affected by post inoculation. The effects of this post inoculation are shown in Fig. 5. The ductile iron with no inoculation, heat 2A, had only 1.5 per cent elongation. Heat 5B, a heat to which 0.80 per cent silicon was added to the furnace charge before the magnesium treatment, had elongations of 12.1 and 13.0 per cent. This increase in elongation would logically be expected as the higher silicon content irons contain more ferrite.

**Impact Strength.** The important effect of silicon additions after magnesium treatment on as-cast ductile increases with an increase in the amount of post inoculation. Both ferrosilicon and calcium silicon, when added as post inoculants, greatly increase the impact strength of as-cast ductile irons. It is interesting to note that when the silicon was added to the furnace, the impact strength increased only moderately. For maximum as-cast impact strength, it is important to have a substantial amount of the silicon added as a post inoculant. The pronounced effect of post inoculation on the impact properties of as-cast ductile iron is shown in Fig. 6.

**Brinell Hardness.** The hardness of ductile iron is markedly affected by post inoculation, as shown in Fig. 7. Hardness decreases with increasing silicon content, with the post inoculated irons showing a marked decrease in hardness. This change in hardness is directly related to the microstructure as post inoculation produced ferritic structures. As the amount of ferrite increases, the hardness decreases.

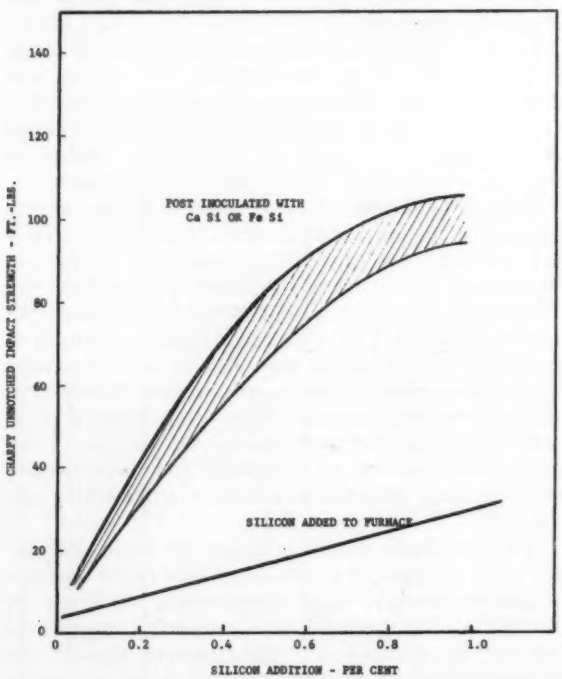



Fig. 6 — Silicon addition effect on as-cast impact strength.

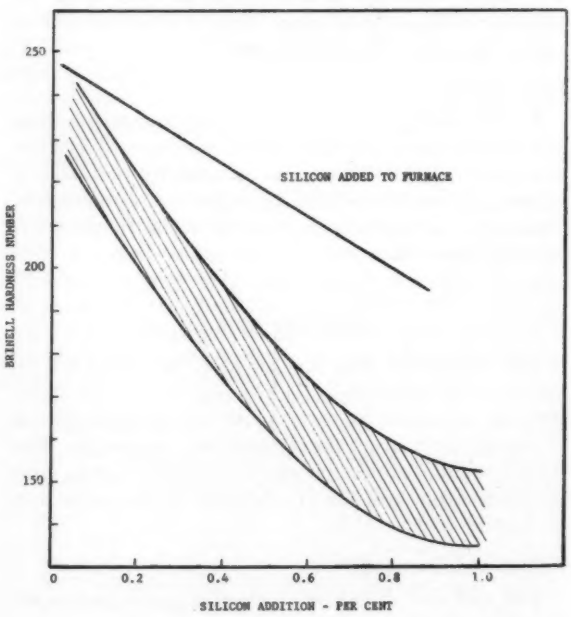



Fig. 7 — Silicon addition effect on Brinell hardness of as-cast ductile iron.

**Microstructure.** The microstructure is greatly affected by the post inoculation process. The number and size of the spheroids present are dependent on the silicon content and the manner in which it is added. Silicon added before the magnesium inoculation decreased the size of the spheroids and increased the number as shown in Figs. 8a and 8b. The graphite spheroids in Fig. 8b are approximately one-half the size of those in Fig. 8a.

When 0.73 per cent silicon was added as a post inoculant (Fig. 8f) the graphite spheroids were found to be about one-half the size of those shown in Fig. 8b, which had approximately the same silicon content. It is clear that silicon increases the number of spherulites and decreases their size. When added as a post inoculant, it is most effective in changing the graphite distribution in ductile iron.

### *Holding Time Effect*

Holding the magnesium treated iron after the post inoculation treatment does affect many of the properties of ductile cast iron.

**Tensile Strength.** The tensile strength of ductile iron is greatly affected by holding time after post inoculation. The maximum tensile strength was obtained in the ductile irons studied approximately 3 min after post inoculation with calcium silicon or ferrosilicon. After 3 min the strength generally decreased. Ductile iron with a lower silicon content seems to be more sensitive to holding time than a ductile iron with a higher percentage of silicon.

**Yield Strength.** As is the case with tensile strength, the maximum yield strength of post inoculated ductile iron is obtained approximately 3 min after the post inoculation. After 3 min the yield strength usually decreases.

**Elongation.** The elongation of ductile iron demonstrates a great sensitivity to holding time. It appears that the best elongation is obtained from 3 to 6 min after treatment. Irons inoculated with calcium silicon fade slower then irons inoculated with ferrosilicon. It does appear that the addition of calcium to a post inoculation is beneficial.

**Impact Strength.** The impact strength is greatly affected by holding time after post inoculation. The impact strength is usually best up to 3 min after inoculation, with the impact strength falling off after that time. Heats 3A, 2A and 2B had their maximum strength at zero holding time, whereas heats 3B, 4A and 3D had their maximum strength at 3 min. Heat 5B, which had no post inoculation, showed an increase in impact strength at 6 min. The strength, however, was low, and this slight increase may have been the results of other variables.

**Brinell Hardness.** Brinell hardness is little affected by holding time. This was true with all heats except heat 2B, which appeared to have an odd behavior in all physical tests. This heat was the one heat where a fading effect was quite noticeable.

**Microstructure.** The microstructure of post inoculated ductile irons is greatly affected by holding time. Figure 9 shows the effect of holding time on the microstructure of heats 3A and 2B. Heat 3A which had an 0.98 per cent silicon post inoculation showed little effect of holding time. Heat 2B, which had 0.48 per cent silicon post inoculation, showed a considerable effect of holding time on both graphite structure and matrix. The amount of ferrite increase with holding time with a corresponding decrease in pearlite. The graphite changed from a spheroidal type to a mixed structure containing some flake type of graphite. Heat 2B was affected more than any other heat by holding time.

TABLE 3 — MECHANICAL PROPERTIES

| Heat | Yield, psi | Tensile, psi | Elong., % | Unnotched Impact, ft-lb | Bhn | Pouring time, min |
|---|---|---|---|---|---|---|
| 4A1 | 46,300 | 71,600 | 16.0 | 71 | 187 | 0.5 |
| 4A2 | 50,750 | 81,600 | 15.2 | 77 | 192 | 3.0 |
| 4A3 | 49,300 | 79,000 | 15.4 | 72 | 187 | 6.0 |
| 4A4 | 48,100 | 72,000 | 7.5 | 42 | 187 | 8.0 |
| 2B1 | 45,450 | 75,500 | 14.0 | 60 | 187 | 0.5 |
| 2B2 | 45,250 | 68,600 | 16.5 | 67 | 159 | 3.0 |
| 2B3 | 46,500 | 69,600 | 13.2 | 42 | 156 | 6.0 |
| 2B4 | 45,550 | 73,900 | 3.7 | 21 | 207 | 9.0 |
| 3A1 | 45,200 | 69,250 | 23.9 | 107 | 159 | 0.5 |
| 3A2 | 44,950 | 70,000 | 24.7 | 99 | 159 | 3.0 |
| 3A3 | 42,200 | 69,100 | 25.5 | 74 | 156 | 6.0 |
| 3A4 | 41,750 | 68,800 | 13.0 | 75 | 156 | 8.0 |
| 3B1 | 45,900 | 77,600 | 13.5 | 56 | 174 | 0.5 |
| 3B2 | 47,300 | 79,150 | 14.5 | 58 | 167 | 3.0 |
| 3B3 | 46,900 | 77,850 | 14.0 | 56 | 170 | 6.0 |
| 3B4 | 47,500 | 74,900 | 11.2 | 21 | 170 | 8.0 |
| 3D1 | 44,200 | 70,350 | 21.5 | 83 | 159 | 0.5 |
| 3D2 | 47,000 | 69,000 | 24.3 | 85 | 152 | 3.0 |
| 3D3 | 45,100 | 68,500 | 22.7 | 84 | 156 | 6.0 |
| 3D4 | 46,600 | 68,200 | 22.3 | 65 | 152 | 8.0 |
| 2A1 | 49,750 | 78,450 | 2.0 | 6 | 235 | 0.5 |
| 2A2 | 49,400 | 82,250 | 1.5 | 2 | 248 | 3.0 |
| 5B1 | 53,800 | 83,150 | 12.1 | 22 | 197 | 0.5 |
| 5B2 | 56,450 | 82,750 | 13.0 | 26 | 201 | 6.0 |

## CONCLUSIONS

Based on the actual and indicated results of this investigation, post inoculation of ductile iron produces these effects:

1. Tensile strength and hardness decreases as the amount of silicon added as a post inoculant is increased.
2. As-cast elongation and impact strength increase with increased amounts of silicon and post inoculation.
3. The number of graphite spheroids increases as the post inoculation increases.
4. The size of the graphite spheroids decreases as the post inoculation increases.
5. Pearlite and cementite are reduced and the amount of ferrite increased by post inoculation.
6. When silicon is added as a post inoculant, it is more effective than when added to the ductile iron before the magnesium treatment.
7. The best properties of post inoculated ductile iron are obtained in the first 3 min following post inoculation. The inoculation effect begins to fade after 3 min.
8. The calcium bearing inoculant appears to be more effective than regular ferrosilicon.

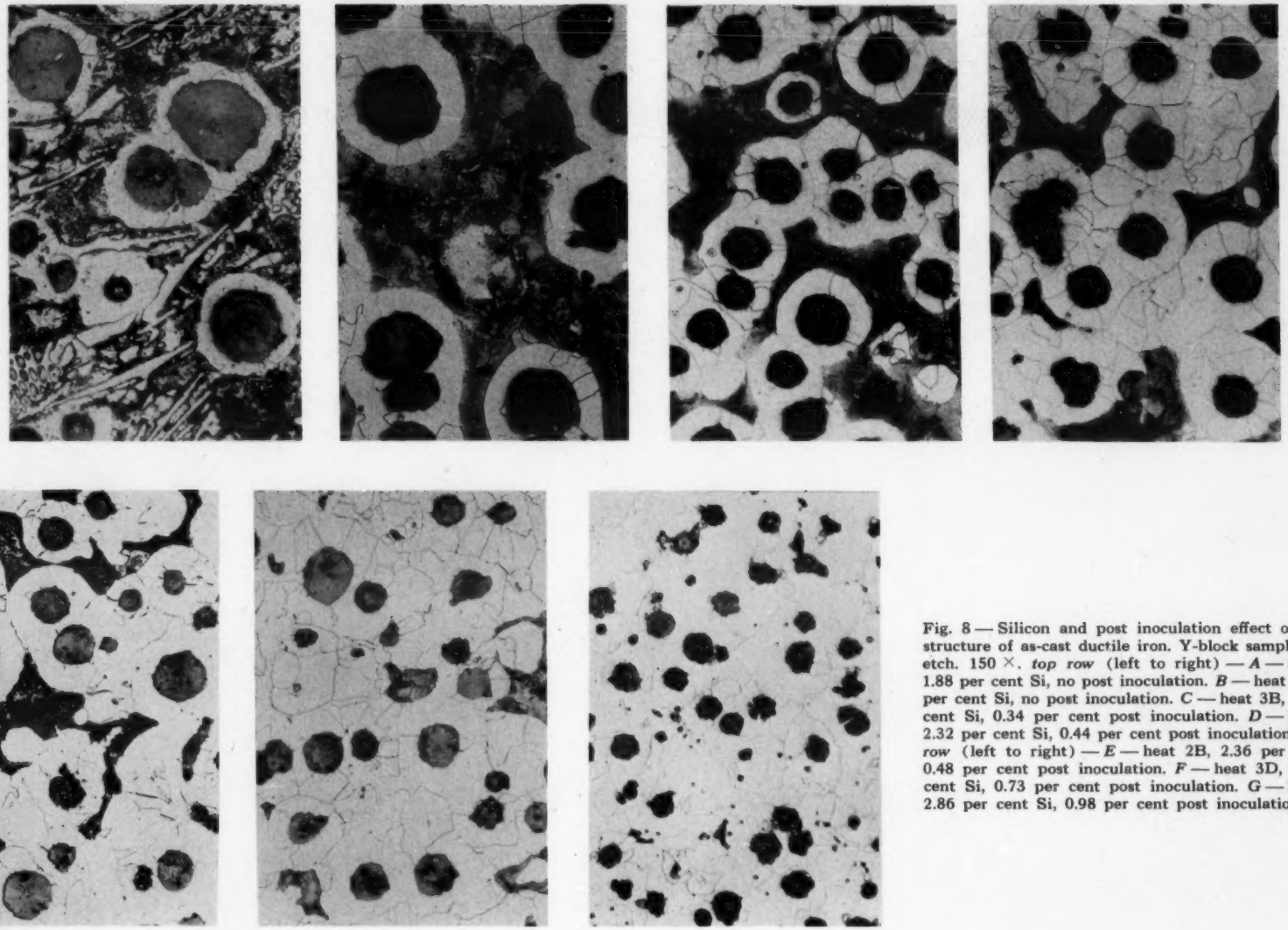


Fig. 8 — Silicon and post inoculation effect on microstructure of as-cast ductile iron. Y-block samples. Nital etch. 150 ×. *top row* (left to right) — *A* — heat 2A, 1.88 per cent Si, no post inoculation. *B* — heat 5B, 2.68 per cent Si, no post inoculation. *C* — heat 3B, 2.22 per cent Si, 0.34 per cent post inoculation. *D* — heat 4A, 2.32 per cent Si, 0.44 per cent post inoculation. *bottom row* (left to right) — *E* — heat 2B, 2.36 per cent Si, 0.48 per cent post inoculation. *F* — heat 3D, 2.61 per cent Si, 0.73 per cent post inoculation. *G* — heat 3A, 2.86 per cent Si, 0.98 per cent post inoculation.

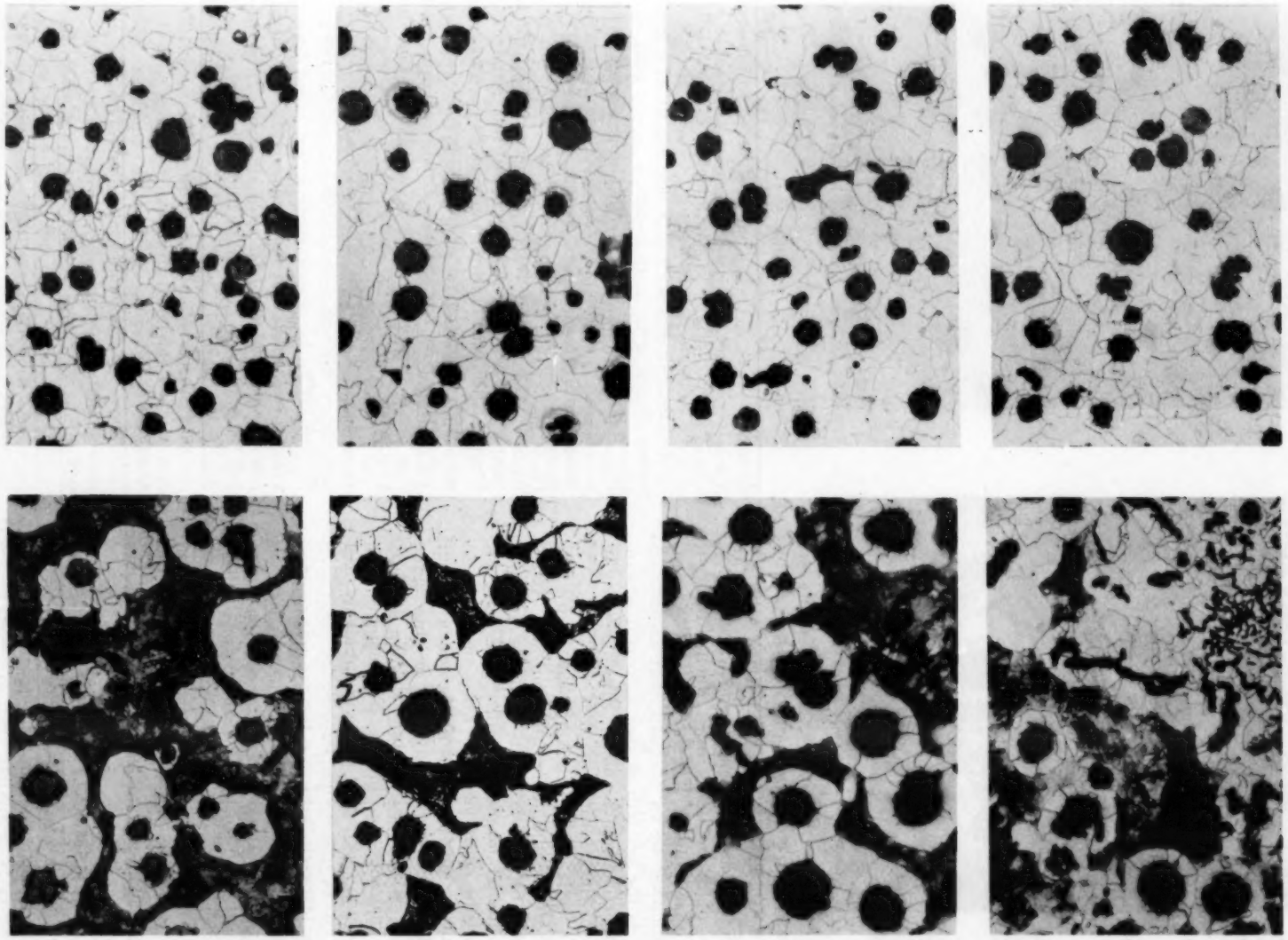
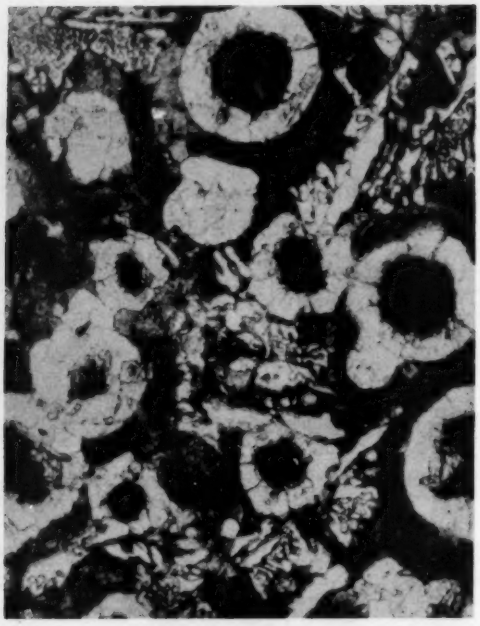
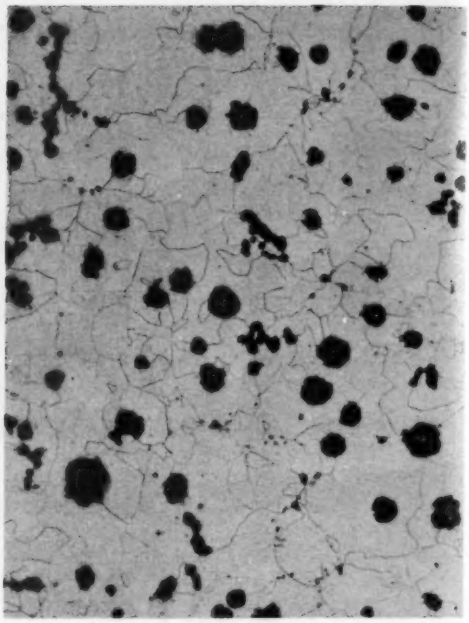


Fig. 9 — Holding time effect on microstructure of as-cast ductile iron. Nital etch. 150 ×. *top row* (left to right) — *A* — heat 3A1, ½-min holding time. *B* — heat 3A2, 3 min holding time. *C* — heat 3A3, 6 min holding time. *D* — heat 3A4, 8 min holding time. *bottom row* (left to right) — *E* — heat 2B1, ½-min holding time. *F* — heat 2B2, 3 min holding time. *G* — heat 2B3, 6 min holding time. *H* — heat 2B4, 9 min holding time.





## FOUNDRY APPLICATION OF EXPERIMENTAL RESULTS

In the preliminary experimentation with the post inoculation of magnesium treated ductile irons, it was shown that completely ferritic structures, as-cast, could be obtained by an adequate post inoculation treatment. As the per cent of post inoculant was increased, impact resistance and ductility increased in proportion to the addition.

Since the early experimentation was carried out using pure magnesium, some questions were raised as to whether or not the results would be the same if a conventional magnesium alloy were used for the nodularizing addition. Other questions were also raised about the small amount of chromium present in some of the base irons. To check on these questions, a heat was melted in the induction furnace, using raw materials low in residual elements. The melted base iron was treated with the commonly used 8 to 10 per cent magnesium ferrosilicon.

A 350 lb, 960 cycle induction furnace was charged with the materials:

Low phosphorus pig iron, lb ..... 180
Direct reduction briquettes, lb ..... 20

TABLE 5

| Chemical Composition, % | Base Iron Treated With 8-10% Mg FeSi | Mg Treated Iron Inoculated With 85% FeSi | Mg Treated Iron Inoculated With 75% FeSi |
|---|---|---|---|
| Carbon | 3.40 | 3.25 | 3.32 |
| Silicon | 1.75 | 2.89 | 3.25 |
| Sulfur | 0.016 | 0.009 | 0.006 |
| Phosphorus | 0.023 | 0.022 | 0.022 |
| Manganese | <0.10 | <0.10 | <0.10 |
| Magnesium | 0.047 | 0.039 | 0.038 |
| Chromium | <0.02 | <0.02 | <0.02 |
| Nickel | 0.0060 | 0.0064 | 0.0069 |
| Molybdenum | <0.01 | <0.01 | <0.01 |
| Copper | 0.085 | 0.096 | 0.092 |
| **Mechanical Properties** | | | |
| Tensile Strength, psi | 84,250 | 70,800 | 76,400 |
| Yield Strength, psi | 47,700 | 55,100 | 63,100 |
| Elongation, % | 7.8 | 20.5 | 18.0 |
| Brinell hardness number | 196 | 170 | 192 |
| Unnotched Impact Strength, 10 x 10 x 55 mm, ft-lb | 21-24 | 80-92 | 90-95 |

TABLE 4 — CHEMICAL COMPOSITION

| Sample | Composition, % | | | | | | | | | Post Inoculant, % |
|---|---|---|---|---|---|---|---|---|---|---|
| | C | Si | S | Mn | Ni | Cr | Cu | P | Mg | |
| 4A | 3.64 | 2.32 | 0.007 | 0.27 | 0.035 | — | 0.135 | 0.026 | 0.054 | 0.44 |
| 2B | 3.87 | 2.36 | 0.009 | 0.23 | 0.075 | 0.057 | 0.130 | 0.035 | 0.040 | 0.48 |
| 3A | 3.72 | 2.86 | 0.005 | 0.25 | 0.081 | 0.049 | 0.137 | 0.03 | 0.042 | 0.98 |
| 3B | 3.72 | 2.22 | 0.004 | 0.27 | 0.046 | 0.042 | 0.137 | 0.03 | 0.049 | 0.34 |
| 3D | 3.70 | 2.61 | 0.006 | 0.25 | 0.074 | 0.043 | 0.137 | 0.026 | 0.042 | 0.75 |
| 2A | 3.84 | 1.88 | 0.006 | 0.25 | 0.094 | 0.092 | 0.123 | 0.047 | 0.043 | 0.0 |
| 5B | 3.32 | 2.68 | 0.008 | 0.26 | 0.145 | 0.150 | 0.134 | 0.026 | 0.044 | 0.0 |

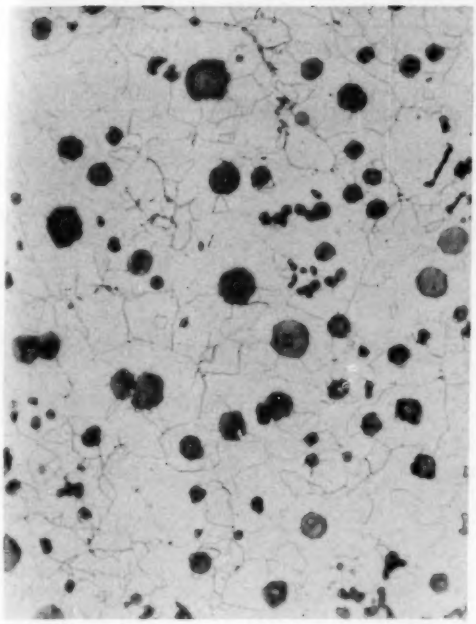


Fig. 10 — (left to right) — Structure of ductile iron treated with 10 per cent magnesium ferrosilicon. 2 per cent nital etch. 150 ×. Structure of magnesium treated metal inoculated with 85 per cent ferrosilicon. 2 per cent nital etch. 150 ×. Structure of magnesium treated metal inoculated with 75 per cent ferrosilicon. 2 per cent nital etch. 150 ×.

Ferrosilicon — 35% Si, lb ..... 4

After melting and raising the temperature to approximately 2750 F, the metal was dumped onto 6 lb of 8 to 10 per cent magnesium ferrosilicon which had been placed in the bottom of a 300 lb ladle. After the reaction had subsided, chill, Y-block and spectrographic specimens were poured.

The remainder of the magnesium treated metal was then split into two equal portions, one-half of which was treated with 85 per cent calcium bearing ferrosilicon and the other half was treated with calcium bearing 75 per cent ferrosilicon. Chill, Y-blocks and spectrographic samples were poured from the irons post inoculated with 75 and 85 per cent ferrosilicon.

## PLANT TEST RESULTS

The chill samples were broken, and the magnesium treated metal had a characteristic white fracture of ductile iron. The magnesium post inoculated samples bent when attempts were made to break them, and they could only be broken by bending them back and forth until fracture took place. The fracture of both post inoculated samples indicated a ductile fracture.

Tensile specimens were cut from the Y-blocks and tested according to standard procedures. Chemical and spectrographic determinations were conducted on all three portions of this heat. Photomicrographs of all three portions of this heat were made at 150 diameters with a nital etch.

**Tensile Strength.** The tensile strength of the as-cast sample was somewhat higher then either post inoculated samples, due to the pearlite present in the microstructure. The additions of 75 and 85 per cent ferrosilicon lowered the tensile strength of the as-cast ductile iron.

**Yield Strength.** The effect of post inoculation on the yield strength is quite important. Yield strengths determined by the 0.2 per cent offset method, of the post inoculated irons were higher than the as-cast material, even though the tensile strength of the as-cast material was higher than the post inoculated strengths.

**Elongation.** The elongation of as-cast ductile iron is greatly improved by post inoculation. In this particular experiment the elongation was increased from 7.8 to 20.5 per cent by post inoculation.

**Microstructure.** The microstructure of the uninoculated, magnesium treated iron (Fig. 10) consisted of graphite nodules in a matrix of ferrite, pearlite and cementite. The ductile irons inoculated with ferrosilicon have microstructures consisting of small graphite nodules in a ferritic matrix. The ferrosilicon inoculated ductile irons were completely ferritic and the microstructure contained smaller and more numerous graphite nodules.

## CONCLUSIONS

The results of this plant test verify the results obtained in the earlier experiments. When the magnesium was added in the form of ferrosilicon magnesium and the post inoculant was added as either 75 ferrosilicon or 85 per cent ferrosilicon, a completely ferritic ductile iron was produced. As indicated by the earlier results, elongation and impact strength values were increased by post inoculating the magnesium treated ductile iron with ferrosilicon.

Microstructural examinations indicate that the post inoculation of magnesium treated iron with ferrosilicon tends to produce graphite spheroids which are much smaller is size and more numerous in number.

## ACKNOWLEDGMENTS

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# PREMIUM QUALITY MAGNESIUM SAND CASTINGS

by K. E. Nelson


## ABSTRACT

**Gating, risering and chilling, metal handling and heat treatment practices used to produce premium quality magnesium sand castings are described. AZ91C-T6, AZ92A-T6, ZK51A-T5, ZK61A-T6, ZH62A-T5, EK31XA-T6, QE22A-T6 and HK31A-T6 alloys were studied. Casting designs amenable to effective gating, risering and chilling practices develop mechanical properties equal to or better than those guaranteed for separately cast test bars. Mechanical properties of AZ92A-T6, ZK61A-T6 and QE22A-T6 compare favorably on an equal volume basis with aluminum casting alloys A356 and C355.**


## INTRODUCTION

Demands from aircraft and missile manufacturers for high integrity, high strength castings have created considerable discussion among producers and users of ferrous and light metal castings.[1-7] Elizondo[1] pointed out the shortcomings of the foundrymen in their failure to meet the aircraft requirements for strength, reproducibility, inspection criteria and promised delivery dates for quality steel castings. Rassenfoss,[2] in his reply for the steel casting industry, suggested that the aircraft industry carefully define its goals in realistic and tangible terms, and seek out foundries having the necessary qualifications required for the difficult development and production work.

Sparling[3] presented a discussion on why aircraft engineers do not use more castings, and what can be done about it. Flemings, et al,[4] described foundry techniques used in producing several prototype AZ91C magnesium castings for missile applications and mechanical properties obtained on bars sectioned from these castings. Gronvold[5,6] reported on the feasibility of obtaining premium quality aluminum and magnesium castings.

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A development project was undertaken by a company to determine the optimum molding and metal handling procedures for the production of fine grain, relatively sound castings. A special test casting in the form of a simple I-beam designed by an aircraft company[7] was used in this program. These castings were then heat treated before sectioning for tensile property determinations.

This paper presents a comparison of castability and tensile properties for several magnesium casting alloys. All alloys were poured into molds essentially prepared under the same molding conditions. The term castability, as described in an earlier paper,[8] refers to the ability of an alloy to fill a sand mold satisfactorily and to form a casting relatively free of deleterious defects.

The techniques of alloy preparation, casting production and heat treatment, as well as testing methods, are given in the Appendix.

## DISCUSSION OF RESULTS

### *Castability*

As in similar evaluations previously published, the measure of castability of the alloys is limited to the observations made during the pouring of the castings, and to the quality of these castings as shown by various inspection techniques. Pouring times required to fill the molds were consistent for all alloys.

Table 1 presents the analysis and casting quality for each alloy composition. In many cases, the x-ray quality was obtained using the casting sections, shown as levels 1, 2 and 3 in Fig. 1, as well as the whole casting. The gating, risering and chilling procedures developed for ZH62A were used to produce all castings. Three basic approaches were investigated:

1. Maximum feed (risering) with the maximum chilling feasible under such feeding conditions.
2. Average feed with maximum chilling.
3. Minimum feed with maximum chilling.



61-69

TABLE 1 — METAL PREPARATION, ANALYSIS, HEAT TREATMENT AND CASTING QUALITY

| Alloy Composition | Chemical Analysis, % Ag | Al | Mn | RE* | Th | Zn | Sol. Zr | Metal Preparation | Heat Treatment | Casting Quality |
|---|---|---|---|---|---|---|---|---|---|---|
| AZ91C, Casting A | | 8.64 | 0.16 | | | 0.72 | | Alloy at 1400 F, chlorinate and superheat for 30 min at 1650-1700 F. Remove crucible from setting, cool 1300 F and pour. | 2 hr 500 F to 790 F (max.). 6 hr at 790 F (max.). 2 hr at 665 F, 20 hr at 790 F (max.). Fast fan cool. 3 hr at 420 F. | Generally sound with some porosity at junctions of thin webs and flanges. |
| AZ91C, Casting B | | 8.63 | 0.24 | | | 0.73 | | Same as above. | Same as above. | Same as above. |
| AZ92A, Casting A | | 9.58 | 0.24 | | | 2.18 | | Same as above. | 2 hr 500 F to 770 F (max.). 6 hr at 770 F (max.). 2 hr at 665 F, 33 hr at 770 F (max.). Quench in 191 F water. 10 hr at 350 F. | Generally sound. |
| AZ92A, Casting B | | 9.48 | 0.25 | | | 2.25 | | Same as above. | Same as above. | Same as above. |
| ZK51A, Casting A | | | | | | 5.27 | 0.78 | Alloy at 1400 F, flux refine and settle at 1400-1450 F. Pour at 1300 F. | 12 hr at 350 F. | Some cope pitting. Heavy sections in critical areas generally sound but contained small areas of porosity. Thin line of porosity in pockets between chill and heavier wall. Nodular-type of segregation mostly in drag section of 2x4x4 in. noncritical area. The cope section in the 2x4x4 in. area also showed porosity. |
| ZK51A, Casting B | | | | | | 5.44 | 0.73 | Same as above. | Same as above. | |
| ZK51A, Casting C | | | | | | 5.39 | 0.70 | Same as above. | Same as above. | |
| ZK61A, Casting A | | | | | | 5.98 | 0.73 | Alloy at 1400 F, flux refine and settle at 1400-1450 F. Pour at 1300 F. | 2 hr at 935 F (max.). Quench in 175-200 F water, 48 hr at 265 F. | Heavy sections in critical and noncritical areas contained greater porosity than found in ZK51A. Porosity in pockets and segregation in casting as in ZK51A. Some cope pitting. |
| ZK61A, Casting B | | | | | | 5.89 | 0.79 | Same as above. | Same as above. | |
| ZK61A, Casting C | | | | | | 6.13 | 0.81 | Same as above. | Same as above. | |
| ZK61A, Casting D | | | | | | 6.12 | 0.88 | Same as above. | Same as above. | |
| ZK61A, Casting E | | | | | | 6.20 | 0.75 | Same as above. | Same as above. | |
| ZK61A, -1, Cstg. F | | | | | | 6.38 | 0.83 | Same as above. | Same as above. | |
| ZK61A, -4, Cstg. G | | | | | | 6.38 | 0.83 | Same as above. | Same as above. | |
| ZK61A, -5, Cstg. H | | | | | | 6.38 | 0.83 | Same as above. | Same as above. | |
| ZH62A, Casting A | | | | | 1.67 | 5.47 | 0.80 | Alloy Zn and Zr at 1400-1450 F, alloy Th at 1500 F. Flux refine and settle at 1400 F. Pour at 1300 F. | 2 hr at 625 F. Fast fan cool. 16 hr at 350 F. | Small amount of cope pitting. Tendency for occasional areas lined-up type of eutectic segregation in nonchilled areas. These areas showed the nodular-type of segregation, heaviest in the drag section of the casting, and an occasional small area of porosity. Except for occasional porous areas in pockets, rest of casting free of porosity. |
| ZH62A, Casting B | | | | | 1.82 | 5.74 | 0.89 | Same as above. | Same as above. | |
| ZH62A, Casting C | | | | | 1.90 | 5.76 | 0.88 | Same as above. | Same as above. | |
| EK31XA, Casting A | | | | 3.10 | | | 0.53 | Used Mg-RE-Zr foundry scrap. Alloy RE and Zr at 1400-1450 F, flux refine and settle. Pour at 1300 F. | 13 hr at 980 F (max.). Quench in 164 F water. 16 hr at 400 F. | Small amount cope pitting. Heavy sections relatively sound but contained areas of porosity. Some porosity in pocket areas. Nodular-type segregation heavies in drag section nonchilled area. |
| EK31XA, Casting B | | | | 3.09 | | | 0.55 | Same as above. | Same as above. | |
| QE22A, Casting A | 2.43 | | | 1.71 | | | 0.53 | Used Mg, QE22A foundry scrap or a mixture. Alloy Ag, RE and Zr at 1400-1450 F, flux refine and settle. Pour at 1300 F. | 8 hr at 985 F (max.) or 4 hr at 995 F (max.). Quench in 160-180 F water. 8 hr at 400 F. | Similar to EK31XA. Possibly slightly less porosity at lower RE level. |
| QE22A, Casting B | 2.46 | | | 1.91 | | | 0.61 | Same as above. | Same as above. | Same as above. |

*(continued on next page)*

TABLE 1 — *(continued)*

| Alloy Composition | Chemical Analysis, % Ag | Al | Mn | RE* | Th | Zn | Sol. Zr | Metal Preparation | Heat Treatment | Casting Quality |
|---|---|---|---|---|---|---|---|---|---|---|
| QE22A, Casting C | 2.59 | | | 2.09 | | | 0.54 | Same as above. | Same as above. | Same as above. |
| QE22A, Casting D | 2.37 | | | 2.18 | | | 0.56 | Same as above. | Same as above. | Same as above. |
| QE22A, Casting E | 2.45 | | | 2.20 | | | 0.61 | Same as above. | Same as above. | Same as above. |
| HK31A, Casting A | | | | | 3.38 | | 0.66 | Alloy Zr at 1400-1450 F, alloy Th at 1500 F. Flux refine and settle at 1400 F. Pour at 1325 F. | 2 hr. at 1060 F (max.). Fast fan cool. 16 hr at 400 F. | Some cope pitting. Some porosity in the heavy nonchilled areas. Also, in this area, nodular-type segregation heaviest in drag. |
| HK31A, Casting B | | | | | 3.48 | | 0.68 | Same as above. | Same as above. | |

*Present essentially as 85% neodymium and 15% praseodymium.

Only small differences were found in tensile properties of bars cut from ZH62A castings and from ZK61A castings produced by these processes. However, improvements in quality of the castings, using variation 2, were found by the normal inspection techniques. This molding procedure, shown in Figs. 2, 3 and 4, incorporates the use of 78 lb of chills in a casting whose poured weight was 70 lb. The shipped weight of the casting was 10 lb.

"Cope-pitting,"[8] was present in Zr-containing I-beam castings. All of the alloys containing Zr exhibited this defect to a comparable degree. It was neither serious nor extensive, indicating that turbulence of the metal stream had been reasonably well controlled before it entered the mold cavity.

The presence of segregation in magnesium alloy castings has been discussed.[16,17,18,19,20] The "nodular" segregation, characteristic of magnesium alloy castings containing zirconium, is seen frequently in the I-beam castings. In radiographs, they are seen as small, dense inclusions of high contrast. By light metallography they show up as idiomorphic crystals. They consist mainly of Zn-Zr and Zn-Zr-Fe compounds in alloys containing zinc, and Th-Zr-Fe compounds in alloys containing thorium. Such inclusions are frequently associated with oxide skins. It can be expected generally that such inclusions can be controlled by proper metal handling procedures. There is also evidence that segregates may contain reacted or partially reacted $SiO_2$ (from the molding sand) in addition to zirconium.[20]

### *Zirconium-base Segregates*

Segregates of a Zn-Zr compound may also form and grow in a casting, depending upon metal handling procedures and cooling rate in the mold. This type of segregation is not necessarily associated with oxides, however. The heaviest concentration of small, dense segregates was always found in the drag section of the unchilled, noncritical part of the casting. The greatest concentration of segregate was on the bottom 1/8-in. of the casting in these areas. Rapid solidification of the

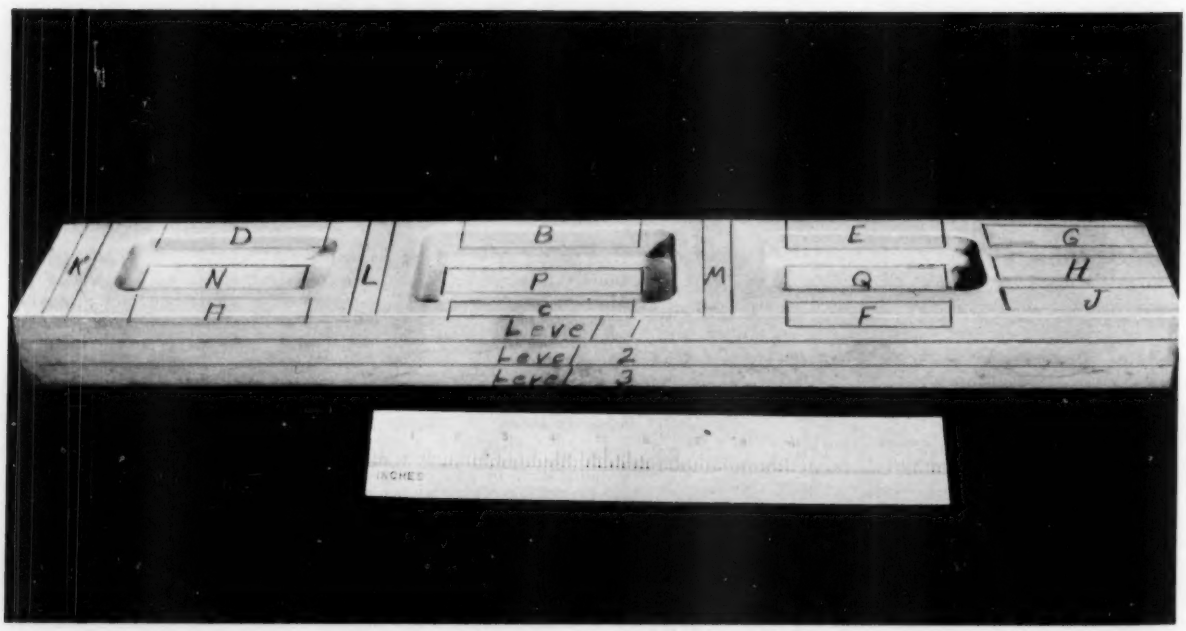



Fig. 1 — Special I-beam casting showing location of test bars.

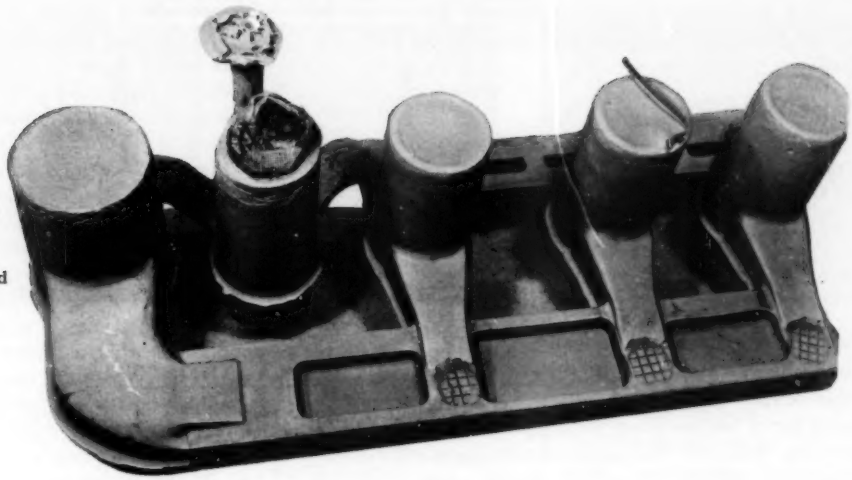


Fig. 2 — Gating and risering used on special test casting.

metal in the mold is necessary to minimize or eliminate such segregation. Additional work is required to understand and control all of the mechanisms of formation of zirconium-base segregates.

Only ZH62A showed any indication of a lined-up segregation, although it has been found previously in other casting designs in each of the alloys containing Zr included in this paper with the exception of ZK51A. This second type of segregation found in magnesium alloys is an eutectic-type usually observed as a dense lined-up formation in a radiograph. This formation occurs when the lower melting eutectic material flows into small cracks or hot tears, sometimes completely healing these defects.

The presence of this type of segregation should serve to warn the foundryman that the gating and risering has been improperly applied, is marginal to say the least and that cracked castings may result if conditions are such that the eutectic does not completely fill the voids.

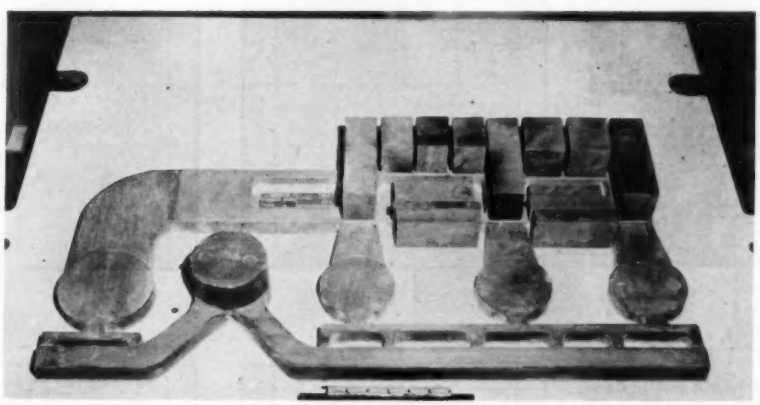


Fig. 3 — Chilling used on special test casting. Drag view.



Fig. 4 — Chilling used on special test casting. Cope view.

When discussing microshrinkage, the alloys can be divided into four groups in increasing order of microshrinkage tendency:

1. AZ91C, AZ92A.
2. ZH62A, HK31A.
3. EK31XA, QE22A.
4. ZK51A, ZK61A.

The difference between groups 1 and 2 is not great. In QE22A, there is an indication of lower microshrinkage at the lower rare earth metal level, although the difference is only slight.

### *Tensile Properties*

One of the requirements for premium quality castings[6] is that minimum properties in the cast part be equal to, or higher than, those specified for separately cast test bars of the same composition. Minimum properties are defined as the lowest value obtained on any bar. It is recognized that some casting designs will be highly stressed in all areas, whereas others will be of a critical nature only in specific areas. The I-beam casting contains both critical and noncritical areas within the part. Bars numbered E, Q, F, G, H and J (Fig. 1) are within the noncritical area in which no chills are used. The other bars are from the designated critical area where maximum chilling is applied.

Distribution plots of tensile properties were made to provide a picture of the spread in properties of separately cast test bars and bars from critical and noncritical areas in the I-beam castings. In some cases, results of separately cast test bars poured with I-beam castings not reported in this paper were included. Some of the plots show a definite skewness.

Figures 5, 6 and 7 are frequency distribution histograms illustrating tensile properties of separately cast test bars and bars from castings of AZ92A-T6, ZK61A-T6 and QE22A-T6 alloys. These plots indicate a property distribution having no definite form. Such distribution is typical of data from the other alloys. In no case is there adequate evidence to state with confidence that a statistically normal distribution exists. The "least-of-n" analysis[21] appeared to be a reasonable means of predicting dependable property levels. It is a nonparametric approach (i.e., it involves no assumed form of the frequency distribution), and makes it possible to arrive at values having the same significance from groups of observations of widely varying size.

For the present analysis it was decided to show, in addition to the range in average properties for each casting within an alloy composition and the overall average property level for each alloy, a property level corresponding to the "least-of-ten." The significance of the value given is the same, regardless of number of tests, as if there were ten tests in every case and the lowest of these ten were selected. As

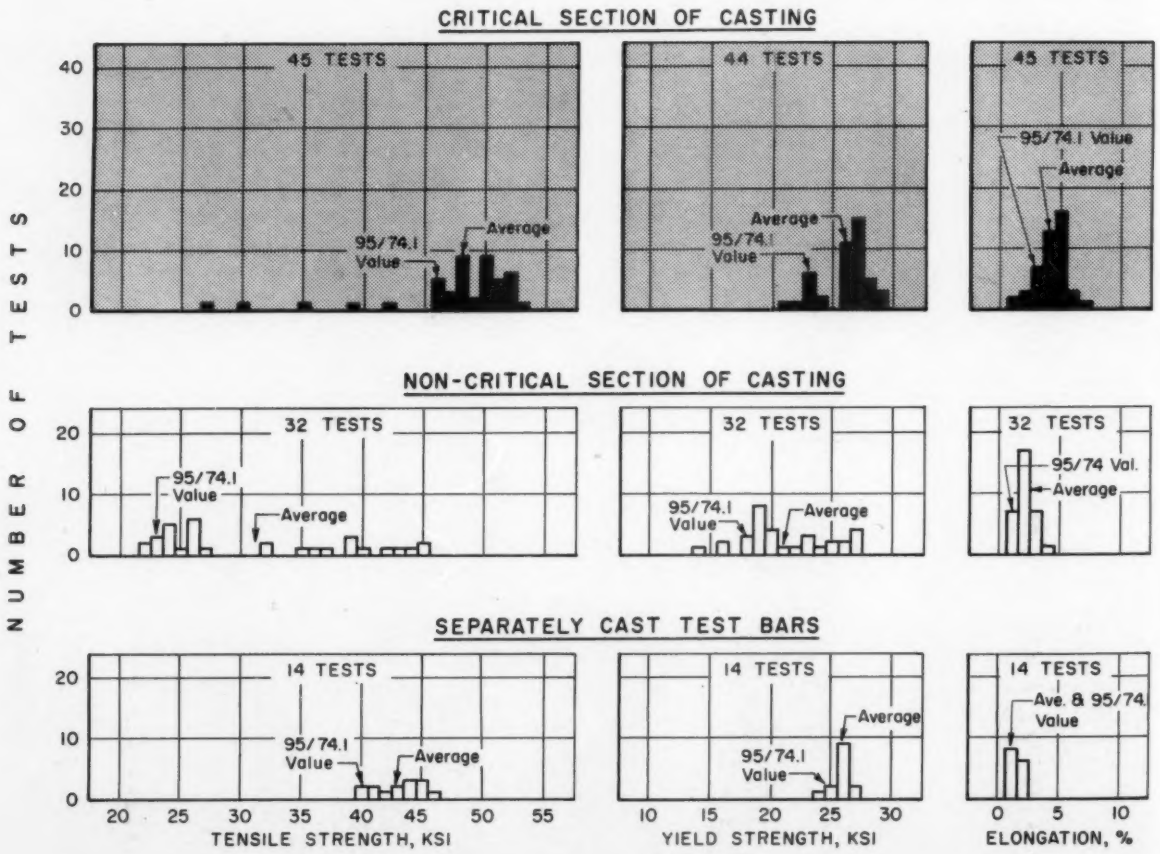



Fig. 5 — Distribution of AZ92A-T6 tensile properties.

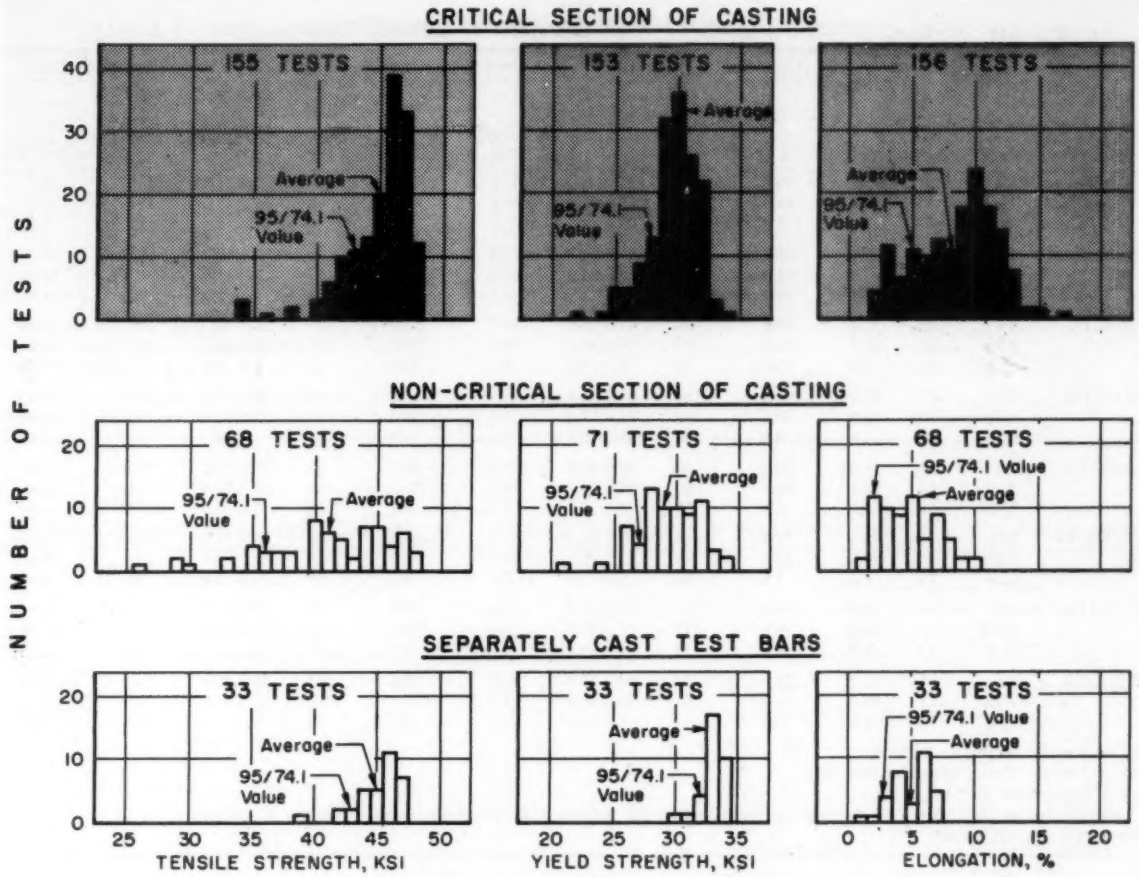



Fig. 6 — Distribution of ZK61A-T6 tensile properties.

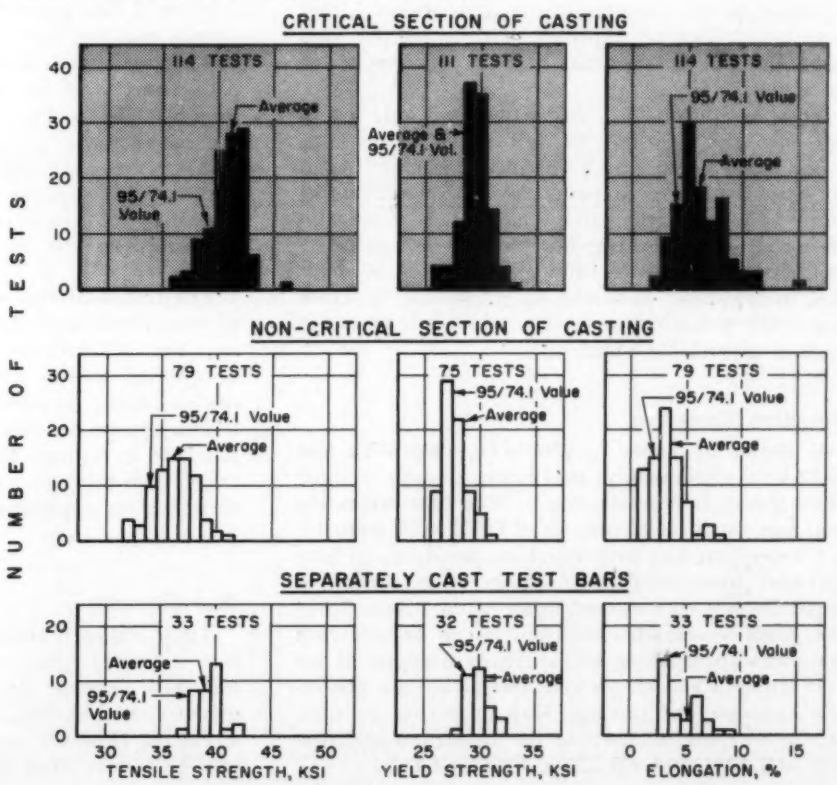



Fig. 7 — Distribution of QE22A-T6 tensile properties.

TABLE 2 — COMPARISON OF TENSILE PROPERTIES*

| Alloy and Temper | Mechanical Property | Separately Cast Test Bars: No. of Tests | Avg. | 95/74.1 No.‡ | Bars from Critical Section: No. of Tests | Range, Individual Casting Avg. | Overall Avg. | 95/74.1 No.‡ | Bars from Noncritical Section: No. of Tests | Range, Individual Casting Avg. | Overall Avg. | 95/74.1 No.‡ |
|---|---|---|---|---|---|---|---|---|---|---|---|---|
| AZ91C-T6 | TS | 16 | 37.8 | 36.9 | 46 | | 37.7 | 37.0 | 32 | 29.2-31.6 | 30.5 | 26.0 |
| | TYS | 16 | 19.8 | 18.0 | 43 | 18.0-18.3 | 18.1 | 16.0 | 31 | 16.7-17.0 | 16.8 | 15.0 |
| | % E | 16 | 3.7 | 3.0 | 46 | 5.2- 5.3 | 5.2 | 4.6 | 32 | 2.9- 3.5 | 3.2 | 2.0 |
| AZ92A-T6 | TS | 14 | 43.1 | 40.0 | 45 | 47.2-48.2 | 47.6 | 46.0 | 32 | 29.4-32.2 | 30.9 | 23.0 |
| | TYS | 14 | 25.9 | 24.6 | 44 | 25.7-26.5 | 26.1 | 23.0 | 32 | 20.6-21.8 | 21.2 | 18.0 |
| | % E | 14 | 1.4 | 1.0 | 45 | 3.7- 4.5 | 4.1 | 3.0 | 32 | 1.8- 2.3 | 2.1 | 1.0 |
| ZK51A-T5 | TS | 26 | 39.0 | 38.0 | 64 | 35.6-38.6 | 37.3 | 35.0 | 47 | 33.9-34.5 | 34.2 | 30.8 |
| | TYS | 26 | 23.7 | 22.0 | 63 | 19.9-20.6 | 20.3 | 20.0 | 47 | 18.6-20.4 | 19.1 | 18.0 |
| | % E | 26 | 7.0 | 6.0 | 64 | 6.4-10.8 | 8.8 | 6.0 | 47 | 4.8- 5.4 | 5.1 | 3.0 |
| ZK61A-T6 | TS | 33 | 45.1 | 43.2 | 155 | 42.9-45.8 | 44.6 | 43.0 | 68 | 30.3-43.9 | 41.1 | 36.0 |
| | TYS | 33 | 33.0 | 32.0 | 153 | 28.6-30.6 | 29.6 | 28.0 | 71 | 27.1-30.6 | 29.4 | 27.0 |
| | % E | 33 | 4.9 | 3.0 | 156 | 6.2-10.9 | 8.4 | 5.0 | 68 | 1.3- 5.9 | 4.8 | 2.0 |
| ZH62A-T5 | TS | 51 | 39.0 | 38.0 | 60 | 37.9-40.0 | 38.8 | 38.0 | 42 | 37.0-39.2 | 37.8 | 36.0 |
| | TYS | 50 | 25.1 | 24.0 | 59 | 23.0-23.2 | 23.1 | 22.0 | 41 | 20.5-22.4 | 21.6 | 20.7 |
| | % E | 51 | 6.7 | 5.0 | 60 | 6.4- 8.1 | 7.1 | 5.0 | 42 | 5.9- 6.5 | 6.2 | 5.0 |
| EK31XA-T6 | TS | 13 | 35.5 | 34.4 | 46 | 37.1-38.4 | 37.7 | 36.0 | 32 | 33.7-34.9 | 34.2 | 33.0 |
| | TYS | 13 | 22.3 | 22.0 | 45 | 21.5-22.1 | 21.9 | 21.0 | 32 | 20.1-20.9 | 20.8 | 20.0 |
| | % E | 13 | 4.2 | 3.4 | 46 | 5.9- 7.0 | 6.4 | 5.0 | 32 | 4.2- 5.4 | 4.8 | 4.0 |
| QE22A-T6 | TS | 33 | 39.3 | 38.0 | 114 | 39.9-41.2 | 40.6 | 39.0 | 79 | 35.2-36.7 | 36.1 | 34.0 |
| | TYS | 32 | 29.9 | 29.0 | 111 | 28.1-30.3 | 29.4 | 29.0 | 75 | 27.2-28.8 | 27.7 | 27.0 |
| | % E | 33 | 4.6 | 3.0 | 114 | 4.6- 8.7 | 6.0 | 4.0 | 79 | 2.0- 4.3 | 3.1 | 2.0 |
| HK31A-T6 | TS | 15 | 33.4 | 31.8 | 44 | 34.2-35.6 | 34.6 | 33.0 | 32 | 31.7-32.0 | 31.9 | 30.0 |
| | TYS | 13 | 16.8 | 16.0 | 41 | 16.6-18.0 | 17.3 | 16.0 | 31 | 16.3-17.6 | 17.0 | 15.8 |
| | % E | 15 | 8.1 | 6.6 | 44 | 7.5- 8.0 | 7.7 | 6.0 | 32 | 5.9- 6.0 | 5.9 | 5.0 |

TS = Tensile Strength, 1000 psi.
TYS = Tensile Yield Strength, 1000 psi.
E, % = Elongation in 2 in.
* Separately cast test bars, bars from noncritical section of castings and bars from critical section of castings.
‡ See text for definition.

explained in the reference,[21] the value shown is that value which can be expected, with a 95 per cent probability, to be exceeded in 74.1 per cent of all future tests.

It is, for convenience, designated in Table 2 and Figs. 5, 6 and 7 as the "95/74.1 value." This value is not recommended as a specification level, but it has been used here to permit a comparison among alloys on a consistent basis. The average value also has a consistent meaning, but provides no useful information regarding the influence of scatter. Without such information, it would be impossible to know how much reduction in level would be necessary to reach a dependable value.

### *Maximum Properties*

As shown in Table 2, AZ92A-T6, ZK61A-T6 and QE22A-T6 alloys exhibit maximum strength of those alloys tested. It is interesting to note that increasing total rare earth metal content of QE22A-T6 from 1.7 to 2.2 per cent had little effect on properties of bars sectioned from castings. A rather wide scatter in properties was encountered in all of the alloys. However, under optimum conditions it can be seen that properties approaching 50 ksi tensile strength, 35 ksi yield strength and 15 per cent elongation are obtainable in magnesium castings. Such properties are comparable to those reported in the aluminum specification MIL-C-21180A for alloys A356 and C355.

Table 3 lists guaranteed tensile properties of separately cast test bars and bars machined from castings, as given in Federal Specification QQ-M-56b, AMS Draft 44AZ for QE22A-T6 and in publications for EK31XA-T6. Table 4 is a summary table, indicating whether the properties of separately cast bars and bars from the I-beam castings met current magnesium specifications listed in Table 3, and the possibility of producing castings having properties equal to or greater than those specified for separately cast bars.

The results of this work are based upon a casting of relatively simple design. Additional work will be required to define the limitations in design that can be applied to the production of premium quality castings. Close cooperation between the designer and foundryman is necessary to make use of the maximum potential in castings. High integrity and high strength magnesium castings can be produced, thus permitting their use in applications previously considered impractical.

### *Metallography*

Table 5 reports average grain size measurements on bars sectioned from castings and on separately cast test bars. Tensile properties of the bars are also given. Generally finer grain results in higher yield strengths. However, in the Zr-containing alloys there may be factors other than grain size influencing yield

TABLE 3 — MECHANICAL PROPERTY REQUIREMENTS FOR MAGNESIUM CASTING ALLOYS

| Alloy and Temper | Separately Cast Specimens | | | Specimens Cut from Castings | | | | |
|---|---|---|---|---|---|---|---|---|
| | | | | TS | | TYS | | |
| | TS, min.* | TYS, min.* | E, %, min.* | Avg.** | Min.* | Avg.** | Min.* | E, %, avg.** |
| AZ91C-T6 | 34.0 | 16.0 | 3 | 25.5 | 17.0 | 14.5 | 12.0 | 0.75 |
| AZ92A-T6 | 34.0 | 18.0 | Not required | 25.5 | 17.0 | 16.0 | 13.5 | — |
| ZK51A-T5 | 34.0 | 20.0 | 5 | 29.0 | 24.0 | 17.0 | 14.0 | 1.25 |
| ZK61A-T6 | 39.0 | 26.0 | 5 | 34.0 | 30.0 | 24.0 | 21.0 | 1.25 |
| ZH62A-T5 | 35.0 | 22.0 | 4 | 31.5 | 26.5 | 17.5 | 15.5 | 1.0 |
| EK31XA-T6 | 31.0 | 19.0 | 2.5 | 26.0 | 22.5 | 17.5 | 16.5 | 0.62 |
| QE22A-T6 | 35.0 | 25.0 | 2 | 32.0 | 28.0 | 23.0 | 20.0 | 2 |
| HK31A-T6 | 27.0 | 13.0 | 4 | 23.0 | 19.0 | 11.7 | 10.5 | 1.0 |

TS — Tensile strength, 1000 psi.
TYS — Tensile yield strength, 1000 psi.
E, % — Elongation in 2 in.

* Guaranteed properties based upon one test.
** Guaranteed properties based upon four to ten tests.

TABLE 4 — SUMMARY — COMPARISON OF VARIOUS ALLOYS FOR PREMIUM QUALITY REQUIREMENTS

| Alloy and Temper | Met Guaranteed Values of Separately Cast Bars | Met Guaranteed Values of Bars from Castings | Values of Bars from Castings Were Equal to or Greater Than Guaranteed Minimum Values for Separately Cast Bars | |
|---|---|---|---|---|
| | | | Noncritical | Critical |
| AZ91C-T6 | Yes | Yes | No — low % E | No — one bar low YS. |
| AZ92A-T6 | Yes | Yes | No — low YS and TS. | No — two bars low TS. |
| ZK51A-T5 | No — 1 bar low % E | Yes | No — low % E, YS and TS. | No — low % E, YS and TS. |
| ZK61A-T6 | No — low % E | No — two bars in non-critical section low TS. | No — low % E and TS, one bar low YS. | No — low % E, YS and TS, close on YS. |
| ZH62A-T5 | Yes | Yes | No — low YS. | No — one bar low % E and TS. |
| EK31XA-T6 | Yes | Yes | Yes | Yes |
| QE22A-T6 | Yes | Yes | No — low % E and TS. | Yes |
| HK31A-T6 | Yes | Yes | Yes | Yes |

TABLE 5 — TENSILE PROPERTIES VS. GRAIN SIZE, SPECIAL TEST CASTING AND SEPARATELY CAST TEST BARS

| Alloy Composition | Cstg. No. | Casting | | | | | Test Bar | | | | |
|---|---|---|---|---|---|---|---|---|---|---|---|
| | | Bar Location* | Avg. Grain Diam., in. x 10-3 | Tensile Properties, 1000 psi | | | Poured with Cstg. No. | Avg. Grain Diam., in. x 10-3 | Tensile Properties, 1000 psi | | |
| | | | | TS | TYS | % E | | | TS | TYS | % E |
| AZ91C-T6 | B | 1H | 5.0 | 31.9 | 20.0 | 3 | B | 5.0 | 38.0 | 20.5 | 3.7 |
| | B | 2A | 5.0 | 38.9 | 18.6 | 6 | A | 5.5 | 38.0 | 21.3 | 3.5 |
| | A | 2A | 5.0 | 38.6 | — | 5 | A | 5.5 | 37.4 | 18.5 | 3.5 |
| | B | 2C | 6.0 | 39.1 | 21.4 | 4 | | | | | |
| | B | 3B | 6.0 | 31.3 | 20.1 | 3 | | | | | |
| | A | 1B | 6.0 | 40.4 | 16.5 | 5 | | | | | |
| | B | 1C | 6.0 | 39.4 | 16.2 | 6 | | | | | |
| | A | 3G | 10.0 | 26.1 | 15.6 | 2 | | | | | |
| | A | 2J | 10.0 | 25.5 | 15.1 | 1 | | | | | |
| | B | 2J | 10.0 | 27.2 | 12.0 | 2 | | | | | |
| | A | 1G | 12.0 | 25.5 | 16.8 | 2 | | | | | |
| AZ92A-T6 | B | 2A | 2.5 | 50.0 | 26.6 | 2 | A | 4.0 | 41.2 | 24.5 | 1.1 |
| | B | 2D | 3.0 | 52.2 | 29.2 | 6 | A | 4.5 | 41.5 | 26.8 | 1.1 |
| | B | 1C | 3.0 | 52.2 | 27.2 | 5 | B | 5.0 | 44.8 | 26.4 | 2.0 |
| | B | 3A | 3.0 | 51.1 | 26.4 | 5 | B | 5.0 | 39.0 | 23.3 | 1.0 |
| | A | 3B | 4.0 | 34.8 | 27.2 | 2 | | | | | |
| | A | 1C | 5.0 | 50.0 | 28.7 | 5 | | | | | |
| | A | 2A | 5.0 | 47.4 | 26.6 | 4 | | | | | |
| | A | 2H | 10.0 | 26.1 | 20.1 | 2 | | | | | |
| | B | 1J | 10.0 | 23.9 | 18.5 | 1 | | | | | |

*(continued on next page)*

TABLE 5 — *(continued)*

| Alloy Composition | Cstg. No. | Casting: Bar Location* | Casting: Avg. Grain Diam., in. x 10-3 | Casting: Tensile Properties, 1000 psi TS | Casting: TYS | Casting: % E | Test Bar: Poured with Cstg. No. | Test Bar: Avg. Grain Diam., in. x 10-3 | Test Bar: Tensile Properties, 1000 psi TS | Test Bar: TYS | Test Bar: % E |
|---|---|---|---|---|---|---|---|---|---|---|---|
| | A | 2G | 12.0 | 25.5 | 19.1 | 2 | | | | | |
| | B | 2J | 12.0 | 24.5 | 16.1 | 2 | | | | | |
| | A | 1H | 12.0 | 24.0 | 16.0 | 2 | | | | | |
| | A | 2J | 15.0 | 24.5 | 16.1 | 2 | | | | | |
| | B | 3G | 15.0 | 23.4 | 13.9 | 1 | | | | | |
| ZK51A-T5 | A | 2Q | 1.0 | 37.0 | 24.0 | 5 | A | 1.0 | 39.0 | 24.8 | 5.0 |
| | A | 1L | 1.0 | 37.9 | 23.0 | 7 | | | | | |
| | A | 2A | 1.5 | 34.9 | 18.4 | 6 | | | | | |
| | A | 3D | 2.0 | 43.4 | 22.4 | 10 | | | | | |
| | B | 1A | 2.0 | 39.8 | 22.2 | 11 | | | | | |
| | C | 1L | 2.0 | 40.6 | 21.7 | 13 | | | | | |
| | B | 2A | 2.0 | 39.1 | 20.5 | 10 | | | | | |
| | C | 2A | 3.5 | 39.4 | 19.6 | 11 | | | | | |
| | A | 1H | 3.5 | 34.4 | 19.5 | 5 | | | | | |
| | A | 3G | 3.5 | 35.4 | 19.3 | 6 | | | | | |
| | B | 2F | 4.0 | 29.9 | 17.1 | 3 | | | | | |
| | C | 3J | 7.0 | 33.6 | 14.2 | 5 | | | | | |
| ZK61A-T6 | D | 3D | 0.7 | 45.1 | 32.3 | 5 | C | 2.0 | 47.2 | 34.0 | 6.2 |
| | C | 2P | 1.5 | 44.3 | 33.3 | 7 | D | 2.0 | 46.5 | 33.9 | 5.5 |
| | D | 1M | 2.0 | 46.9 | 32.3 | 12 | A | 2.0 | 46.4 | 33.3 | 5.5 |
| | D | 2D | 2.0 | 48.0 | 32.1 | 12 | C | 3.0 | 46.9 | 33.0 | 6.2 |
| | B | 2N | 2.0 | 48.0 | 32.1 | 11 | B | 4.0 | 46.5 | 33.4 | 5.1 |
| | C | 1B | 2.0 | 42.3 | 31.6 | 4 | | | | | |
| | D | 2A | 2.0 | 38.0 | 27.6 | 4 | | | | | |
| | C | 3B | 2.0 | 35.4 | 25.2 | 2 | | | | | |
| | C | 1F | 2.5 | 47.4 | 34.4 | 7 | | | | | |
| | B | 1A | 2.5 | 45.8 | 31.3 | 10 | | | | | |
| | B | 3A | 2.5 | 43.2 | 28.7 | 6 | | | | | |
| | D | 3L | 2.5 | 42.6 | 25.4 | 9 | | | | | |
| | A | 2Q | 3.0 | 48.9 | 32.1 | 10 | | | | | |
| | H | 1B | 3.0 | 48.4 | 31.9 | 10 | | | | | |
| | A | 1A | 3.0 | 47.3 | 31.2 | 10 | | | | | |
| | C | 2A | 3.0 | 46.9 | 31.1 | 14 | | | | | |
| | D | 1G | 3.0 | 37.0 | 27.8 | 3 | | | | | |
| | B | 3L | 3.0 | 42.8 | 24.9 | 11 | | | | | |
| | B | 2A | 3.5 | 41.7 | 27.6 | 6 | | | | | |
| | D | 2G | 3.5 | 37.5 | 25.5 | 3 | | | | | |
| | B | 2F | 3.5 | 28.6 | 23.5 | 2 | | | | | |
| | F | 1K | 4.0 | 44.2 | 31.9 | 4 | | | | | |
| | A | 3E | 4.0 | 46.9 | 31.6 | 7 | | | | | |
| | A | 2A | 4.0 | 37.0 | 27.7 | 3 | | | | | |
| | A | 2L | 4.0 | 44.6 | 26.5 | 12 | | | | | |
| | C | 2F | 4.0 | 34.9 | 26.2 | 2 | | | | | |
| | F | 3K | 4.0 | 45.2 | 24.5 | 9 | | | | | |
| | A | 3A | 4.5 | 33.9 | 22.3 | 3 | | | | | |
| | F | 2A | 6.0 | 47.8 | 30.3 | 10 | | | | | |
| | H | 1C | 6.0 | 47.2 | 25.4 | 13 | | | | | |
| ZH62A-T5 | C | 1B | 1.5 | 41.5 | 25.9 | 10 | A | 2.5 | 37.5 | 24.5 | 5.3 |
| | C | 2B | 1.5 | 39.6 | 24.7 | 6 | A | 3.0 | 37.3 | 25.1 | 4.0 |
| | A | 1D | 2.0 | 39.6 | 23.9 | 7 | C | 3.0 | 39.0 | 24.9 | 5.6 |
| | C | 2A | 2.0 | 39.8 | 22.1 | 7 | | | | | |
| | A | 2L | 2.5 | 37.5 | 25.5 | 5 | | | | | |
| | B | 1B | 3.0 | 41.0 | 25.1 | 8 | | | | | |
| | A | 2A | 3.0 | 37.5 | 22.7 | 6 | | | | | |
| | B | 2L | 4.0 | 41.4 | 23.9 | 9 | | | | | |
| | B | 2A | 4.0 | 38.9 | 22.7 | 7 | | | | | |
| | A | 1H | 4.0 | 36.5 | 21.9 | 5 | | | | | |
| | B | 3D | 5.0 | 39.4 | 23.2 | 6 | | | | | |
| | B | 1J | 5.0 | 39.1 | 22.2 | 6 | | | | | |
| | A | 2G | 5.0 | 38.5 | 21.0 | 8 | | | | | |
| | B | 2J | 6.0 | 38.3 | 21.9 | 6 | | | | | |
| | B | 2F | 7.0 | 38.3 | 22.4 | 5 | | | | | |
| | B | 3H | 7.0 | 40.2 | 21.5 | 8 | | | | | |
| | C | 1G | 7.0 | 37.0 | 19.8 | 6 | | | | | |
| | C | 2G | 8.0 | 36.5 | 19.5 | 5 | | | | | |

*(continued on next page)*

TABLE 5 — *(continued)*

| Alloy Composition | Cstg. No. | Casting: Bar Location* | Casting: Avg. Grain Diam., in. x 10-3 | Casting: Tensile Properties, 1000 psi: TS | TYS | % E | Test Bar: Poured with Cstg. No. | Test Bar: Avg. Grain Diam., in. x 10-3 | Test Bar: Tensile Properties, 1000 psi: TS | TYS | % E |
|---|---|---|---|---|---|---|---|---|---|---|---|
| EK31XA-T6 | B | 2C | 1.0 | 37.2 | 24.5 | 7 | A | 1.0 | 36.2 | 22.9 | 4.4 |
| | A | 1A | 1.5 | 39.6 | 22.6 | 8 | B | 1.0 | 34.2 | 22.4 | 3.3 |
| | B | 2A | 2.0 | 35.0 | 22.6 | 5 | | | | | |
| | A | 2A | 2.0 | 40.5 | 21.1 | 10 | | | | | |
| | A | 1J | 3.5 | 35.1 | 19.6 | 6 | | | | | |
| | B | 2J | 3.5 | 33.5 | 20.1 | 5 | | | | | |
| QE22A-T6 | E | 2A | 0.7 | 41.7 | 30.5 | 8 | A | 1.5 | 39.9 | 30.4 | 5.6 |
| | A | 3C | 1.0 | 45.8 | 31.2 | 15 | B | 1.5 | 40.0 | 29.4 | 8.6 |
| | C | 3B | 1.0 | 41.2 | 30.9 | 6 | D | 2.0 | 38.2 | 31.5 | 3.0 |
| | C | 2A | 1.0 | 41.3 | 30.2 | 5 | E | 2.0 | 38.5 | 31.4 | 3.2 |
| | B | 1B | 1.0 | 42.2 | 29.7 | 8 | E | 2.0 | 38.4 | 29.7 | 3.2 |
| | E | 2P | 1.0 | 41.7 | 28.6 | 7 | C | 2.5 | 38.5 | 29.6 | 2.1 |
| | D | 3H | 1.0 | 32.2 | 27.0 | 2 | D | 2.5 | 37.3 | 29.6 | 2.9 |
| | D | 3G | 1.0 | 32.1 | 26.5 | 3 | | | | | |
| | D | 2D | 1.5 | 42.7 | 31.5 | 8 | | | | | |
| | D | 1K | 1.5 | 41.9 | 30.3 | 7 | | | | | |
| | A | 2A | 1.5 | 41.9 | 29.4 | 11 | | | | | |
| | B | 1K | 1.5 | 35.9 | 25.7 | 3 | | | | | |
| | E | 2K | 2.0 | 39.7 | 32.6 | 5 | | | | | |
| | D | 2A | 2.0 | 41.7 | 30.7 | 5 | | | | | |
| | B | 2A | 2.0 | 40.5 | 27.0 | 6 | | | | | |
| | D | 3C | 3.5 | 40.2 | 30.4 | 5 | | | | | |
| | E | 1J | 3.5 | 36.2 | 27.7 | 2 | | | | | |
| | A | 1J | 3.5 | 36.4 | 26.6 | 4 | | | | | |
| | C | 1H | 3.5 | 35.2 | 26.0 | 3 | | | | | |
| | E | 3H | 4.0 | 32.8 | 27.3 | 1 | | | | | |
| | D | 1G | 4.0 | 35.4 | 26.6 | 3 | | | | | |
| | D | 2J | 4.0 | 33.3 | 26.3 | 1 | | | | | |
| HK31A-T6 | A | 1K | 3.0 | 35.6 | 19.7 | 8 | B | 2.5 | 29.8 | 17.5 | 4.7 |
| | A | 2A | 3.0 | 35.9 | 18.0 | 9 | A | 2.5 | 33.5 | 17.3 | 8.3 |
| | B | 1C | 3.5 | 36.5 | 17.9 | 9 | | | | | |
| | A | 2M | 3.5 | 35.3 | 16.6 | 6 | | | | | |
| | A | 3D | 4.0 | 35.9 | 20.2 | 7 | | | | | |
| | A | 3K | 4.0 | 34.2 | 17.5 | 7 | | | | | |
| | A | 1G | 4.0 | 31.2 | 17.4 | 6 | | | | | |
| | A | 2G | 4.0 | 31.7 | 17.0 | 6 | | | | | |
| | A | 3L | 4.0 | 34.8 | 16.3 | 8 | | | | | |
| | B | 2A | 4.5 | 32.5 | 15.5 | 7 | | | | | |
| | A | 3F | 4.5 | 32.1 | 14.8 | 6 | | | | | |
| | B | 3C | 5.0 | 34.1 | 17.7 | 7 | | | | | |
| | B | 2G | 5.0 | 30.7 | 17.5 | 6 | | | | | |
| | B | 3H | 5.0 | 28.2 | 16.2 | 4 | | | | | |
| | B | 2J | 5.0 | 31.5 | 15.4 | 6 | | | | | |
| | B | 3G | 7.0 | 29.3 | 14.5 | 5 | | | | | |

*Number refers to level in casting, letter refers to location in casting. See Fig. 1.

TS — Tensile strength, 1000 psi.

TYS — Tensile yield strength, 1000 psi.

E, % — Elongation in 2 in.

strength. One may hypothesize that wide variations in yield strength not attributable to variations in grain size are caused by variations in amount of solute available for precipitation during aging. Supporting evidence for this hypothesis has been found in Mg-Al-Zn and ZK61A alloys.

Generally the segregation encountered in the Zr-containing alloys was not responsible for low properties, although there were a few cases where the quantity of segregation was large and did result in a lowering of elongation and tensile strength. Quite often it was possible to attribute low elongations and tensile strengths to microporosity or nonmetallic inclusions.

## CONCLUSION

Premium quality magnesium castings can be produced at guaranteed minimum mechanical property levels. While this work was not sufficiently extensive to establish absolute minimum values for bars cut

from castings, it appears that properties of bars in such castings would be equal to, or better than, those guaranteed for separately cast test bars.

In promoting the concept of premium castings, it is necessary that the designer and foundryman understand each others problems, and realize that the development of properties will be an individual project for each casting design. The foundryman must pay close attention to all details of production in order to obtain sound castings of the right composition which can be heat treated to the maximum strength.

Alloys AZ92A-T6, ZK61A-T6 and QE22A-T6 show the most promise with respect to maximum strength castings.

## APPENDIX

### *Procedure*

The melting and alloying of magnesium alloys is covered in a published bulletin.[9] A note of caution should be injected at this time relative to the hazards of contamination of magnesium alloys with harmful impurities. In the Mg-Al-Zn alloys, contaminations with zirconium could result in a melt which would not respond to normal grain refining techniques and also result in a reduction in aluminum content. Likewise, serious losses of Zr may be encountered in Mg-Zr alloys if contaminated with aluminum and silicon. In EK31XA and QE22A alloys, contamination with cerium and lanthanum, such as found in mischmetal used in alloys EZ33A and ZE41A, could result in a lowering of properties.

Details of metal preparation are given in Table 1. Except for alloys EK31XA and QE22A, the melts were prepared using virgin materials. When foundry run-around scrap or a mixture of scrap and virgin material was used as in alloys EK31XA and QE22A, additions of didymium and zirconium were made to make up losses of these elements incurred during remelting. Didymium was added as a magnesium-didymium hardener, analyzing between 10 and 15 per cent didymium. Zirconium was added as a magnesium-zirconium hardener, containing between 30 to 50 per cent zirconium. Thorium was added as sintered thorium pellets.

The added elements were analyzed by wet chemical methods,[10] while the usual impurities were determined spectroscopically.[10]

Sand practice used in producing molds for magnesium alloys is basically the same as that used for other metals. Consideration must be given to certain inherent characteristics of magnesium when preparing such molding sand. The rapid oxidation rate of magnesium in the molten state, its low density and its low heat content are three of these characteristics. The molds for the castings in this program were produced using sand mixed with one of the relatively new waterless binders.

The absence of water allows for a significantly reduced inhibitor content with a resulting low level of generation of mold gases. As two of the requirements of the special I-beam casting were surface smoothness and close dimensional control, the use of the finer grained sand used with the waterless binders was deemed necessary. Also, the lower gas level in a mold using such binders was attractive, when considering pouring the more highly reactive alloys containing Zr.

Molding and gating and risering practices for magnesium are outlined in other publications.[11,12,13]

The rigging used in this evaluation was developed for alloy ZH62A. This was then used with the other alloys, thus allowing a good comparison of castability features. Figure 2 shows one view of the as-poured casting. Extensive use of chills was made, making certain not to use them in the designated noncritical area. Chills were cast iron. Fifty-two lb of chills were used in the drag and 26 lb in the cope section of the mold. The location of the chills is shown in Figs. 3 and 4. The poured weight of the casting was 70 lb, while the shipped weight was 10 lb.

Visual, penetrant inspection and x-ray examinations were given each casting. Heat treatments of the castings and separately cast bars poured with the castings were performed in production furnaces using the schedules outlined in Table 1. Standard magnesium heat treating practices have been described.[14] The quenching in hot water of alloys AZ92A, ZK61A, EK31XA and QE22A did not result in any dimensional or cracking problems in the castings.

After heat treatment, the castings were sectioned, as shown in Fig. 1. These sections were machined to conform to standard A.S.T.M. specified shapes for testing. Separately cast and machined bars were then tested to obtain strength determinations, again using recommended A.S.T.M. procedures. Grain size determinations were made using the comparison method.[15]

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# LADLE DEGASSING FOR FOUNDRY USE

by H. S. Philbrick, Jr.


## ABSTRACT

**A portable ladle degassing unit is described which can handle 4000 to 5000 lb of metal with no heat being added during the process. With optimum position of the unit in a shop, degassing operations should be accomplished in less than 10 min, according to the author. Temperature losses should be proportionately reduced with this time, and can further be reduced by ladle preheating.**


## INTRODUCTION

Vacuum degassing has been accepted by manufacturers of large steel forgings on the basis of reduced hydrogen flaking and shorter thermal cycles. Steel foundries, however, are not as concerned with these factors and must look for other reasons in order to justify the vacuum degassing process.

The literature gives us considerable information on the use of vacuum degassing for steel castings. Several references[4,5,6,7] report improvements in transverse properties, grain structure, density, porosity, surface defects, cleanliness and fluidity. These benefits are impressive but they have not persuaded foundries to undertake degassing on any major scale. A need is recognized for first hand information under actual shop conditions. Why this information has not been obtained may be due in part to the absence of proper tools.

Up to now, steel foundries have been offered a limited choice in the selection of vacuum degassing equipment. On the one hand, they have been shown installations as described by Oreheski and Hornak[1] and others[2,3] representing large capital investments, hardly the types to be used for developmental purposes. On the other hand, there are small units as described by Parsons,[7] limited in size by the capacity of the pump or, units of the kind described by Dietz,[5] limited in size by demands for bigger heating apparatus and better refractories.

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What then for the degassing of 5000 lb, 10,000 lb or 20,000 lb heats?

## SMALL HEAT PROBLEM

With small heats the crux of the problem lies with temperature losses. There must be a rapid handling and degassing cycle. Enormous pumping capacities are required, and distances between melting furnace, degassing unit and molds must be a minimum. Operation of chamber cover and movement of ladle in and out of unit must be rapid. Maximum preheating of ladle must be employed. The unit described herein has been developed to meet some of these conditions.

Figure 1 shows unit with chamber, evacuating system and control panel mounted on a common base. The control panel contains indicating instruments and manual control valves. Steam jets and condensers are seen in back of chamber and panel, and an observers platform at center. The operator stands at the panel except at end of a cycle, when he moves to platform to operate nitrogen and vacuum break valves and guides removal of ladle.

Steam jet pumping equipment is a must. If the pumping capacity of the unit shown were to be duplicated, the equivalent of 14 pumps each with a 40 horsepower motor would be needed. In addition, large filter capacities would be needed. The dead end pressure of the unit shown is less than 50 microns.

Figure 2 shows an end view of unit with chamber open ready to receive the ladle. Unit should be placed parallel with crane bridge so that crane operator looks into chamber and can rack in ladle with one motion. Figure 3 shows ladle entering tank. Figure 4 shows tapping the 5000 lb of steel from a 70,000 lb furnace. This illustrates the ability to treat partial heats.

## LADLE CONFIGURATION

As shown in Fig. 4, ladle has a conical hood and cover over top opening. The purpose of this configuration is threefold – (1) to contain splashing metal, (2) to reduce heat loss from metal and (3) to protect chamber from radiant heat of molten metal.

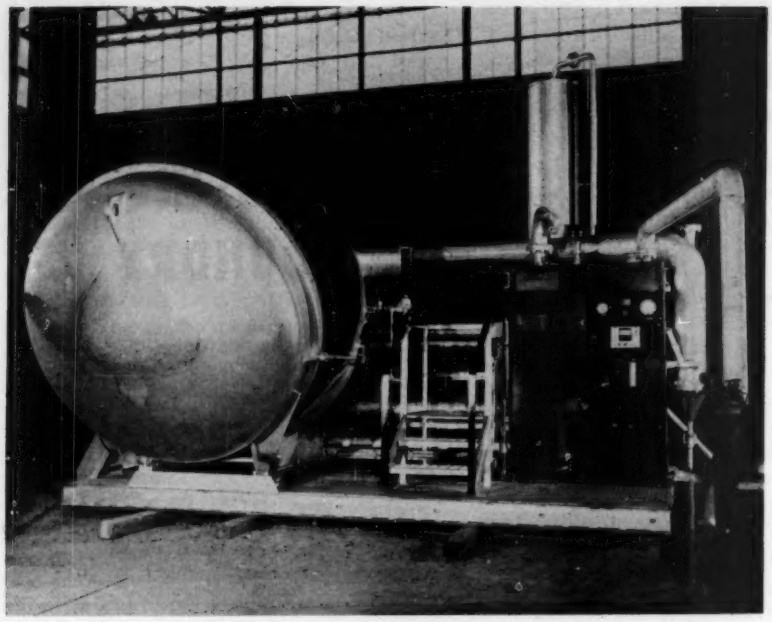


Fig. 1 — Front view of ladle vacuum degassing unit.

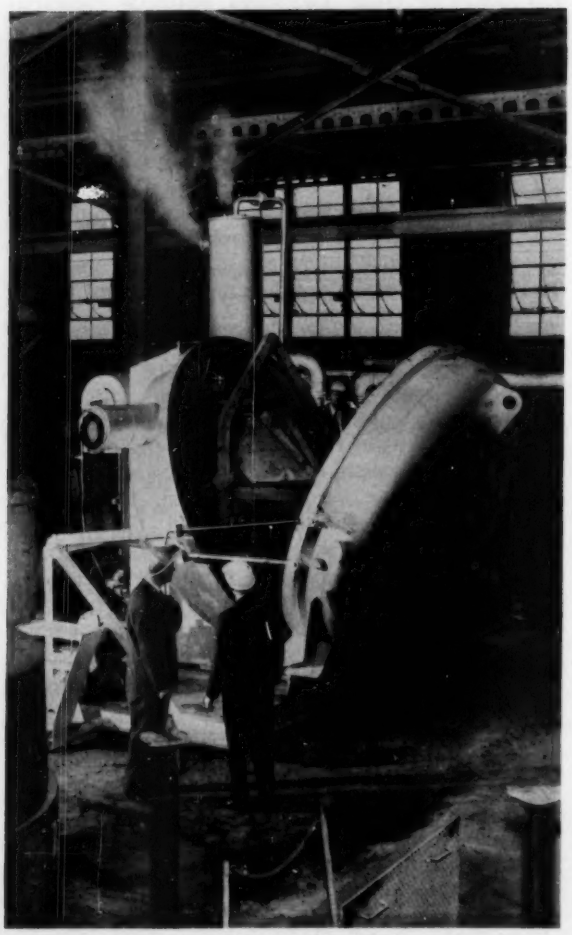


Fig. 2 — End view of unit with chamber open.

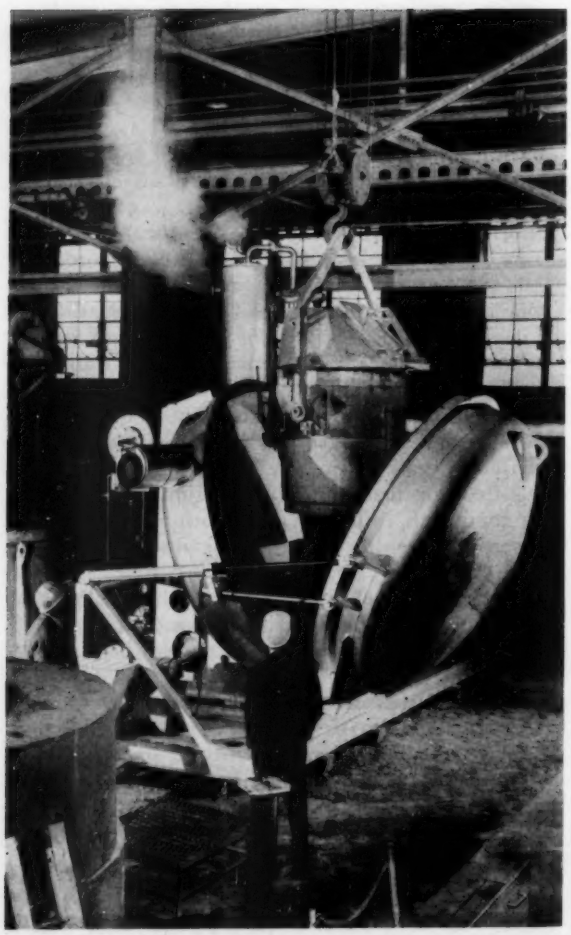


Fig. 3 — Ladle entering tank.

In operation, steel is tapped, and the ladle cover is closed and sealed with patching refractory. The ladle is lowered into chamber. Pumpdown starts, reaching 500-700 microns at the end of a 4 to 5 min cycle. The isolating valve between chamber and steam jets is closed and the chamber is pressurized to ½ atmosphere with nitrogen and then vented to atmosphere through a vacuum break valve. The chamber is opened, ladle lifted and moved to pouring floor. Total time from tap to teem has been running between 14 and 20 min.

Table 1 shows the results of gas removal, cycle time and tapping temperatures for seven heats carried out at a plant. Gas content was taken from pin tube samples and analyzed by the vacuum fusion method as described in a reference.[3] Heat averages show reduction in gas content as – hydrogen – 54 per cent, oxygen – 50 per cent and nitrogen – negligible.

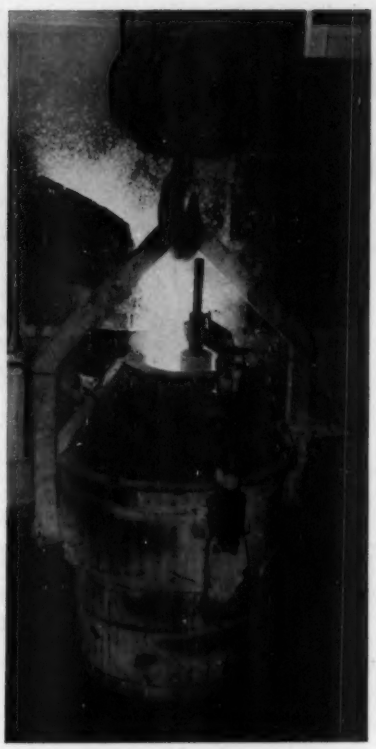


Fig. 4 — Tapping 5000 lb of metal into small ladle from a 70,000 lb heat.

TABLE 1 — DEGASSING RESULTS WITH 5000 LB UNIT

| Date | Heat No. Grade | Before Tap* $H_2$ | $N_2$ | $O_2$ | Degassed* $H_2$ | $N_2$ | $O_2$ | Start Tap To Tap, F | Start Tap To Teem, min |
|---|---|---|---|---|---|---|---|---|---|
| 2/17/61 | 213751 | 3.3 | 37 | 65 | 1.1 | 25 | 31 | 3120 | 23 |
| | FX | | | | 1.1 | 26 | 38 | | |
| 2/21/61 | 213755 | 3.4 | 20 | 125 | 1.4 | 28 | 50 | 3155 | 16 |
| | 4340 | | | | 1.2 | 39 | 44 | | |
| 2/22/61 | 213758 | 4.4 | 27 | 85 | 1.6 | 25 | 44 | 3120 | 19 |
| | FX | | | | 1.6 | 23 | 30 | | |
| 2/27/61 | 113844 | 3.5 | 23 | 126 | 2.0 | 26 | 71 | 3110 | 20 |
| | 8640 | 3.5 | 25 | 89 | 1.7 | 31 | 47 | | |
| | | | | | 1.8/1.6 | 27 | 37 | | |
| 3/1/61 | 213765 | 4.0 | 22 | 54 | 1.2 | 28 | 66 | 3135 | 15 |
| | FX | 3.7 | 17 | 23 | 1.4 | 27 | 53 | | |
| | | 3.8 | 11 | 19 | 1.1 | 27 | 28 | | |
| | | 4.0 | 19 | 49 | | | | | |
| 3/7/61 | 213773 | 2.9 | 28 | 94 | 1.5 | 21 | 25 | 3120 | 14 |
| | FX | 2.8 | – | – | 1.4 | 25 | 34 | | |
| 3/16/61 | 113867 | 3.7 | 30 | 59 | 1.9 | 29 | 59 | 3120 | 16 |
| | FX | | | | 1.1 | 29 | 35 | | |
| | Heat Averages | 3.6 | 27 | 84 | 1.3 | 27 | 42 | | |

*Gas Content in ppm.

Metal for the above tests was taken from 60,000 to 70,000 lb heats. After the small operation was finished, the balance of the full heat was tapped into a production ladle and degassed.[3] Comparison of the two operations show that the results from the 5000 lb unit closely parallel the 70,000 lb production unit, except that in the latter a 30 per cent reduction in nitrogen is obtained.

The above operations were made on high carbon steels tapped in a silicon killed condition.

## UNKILLED LOW CARBON STEELS

In view of the interest shown in the degassing of unkilled, low carbon steels, one test was made taking steel during the oxidizing period of a carbon steel heat. This gave an unkilled steel with a 0.07 per cent carbon content. Violent action of the bath was observed during the degassing of this steel. However, no metal splashed out of the ladle. Chamber pressures followed the same pattern as for fully killed steel reaching 550 microns after 5½-min under vacuum. Results of this test are shown in Table 2. Hydrogen was reduced 62 per cent, oxygen 57 per cent and nitrogen 27 per cent.

TABLE 2 — RESULTS FROM EXPERIMENTAL HEAT WITH LOW CARBON UNKILLED STEEL

| Date | Heat No. Grade | Before Tap* $H_2$ | $N_2$ | $O_2$ | Degassed* $H_2$ | $N_2$ | $O_2$ |
|---|---|---|---|---|---|---|---|
| 4/5/16 | 113893 | 3.8 | 67 | 455 | 1.6 | 43 | 231 |
| | Exp. | 3.7 | 51 | 482 | 1.3 | 48 | 169 |
| | Heat Averages | 3.7 | 59 | 468 | 1.4 | 45 | 200 |

*Gas Content in ppm

In the first month of operation a total of 11 heats have been degassed with little interruption in the normal flow of work through one melt shop. No mechanical difficulties were encountered. Little attention was given to the unit in the way of cleaning vessel, cover o-ring or other parts. No increase in leakup rate was observed. This rate was 5 microns/min a one mm Hg absolute measure in the author's company's shop.

A final point is illustrated by Figs. 5 and 6. Here the unit is shown moving through the shop on a crane hook, and then on low-boy truck on way to next user. Unit can be connected to services and put into operation in one day.

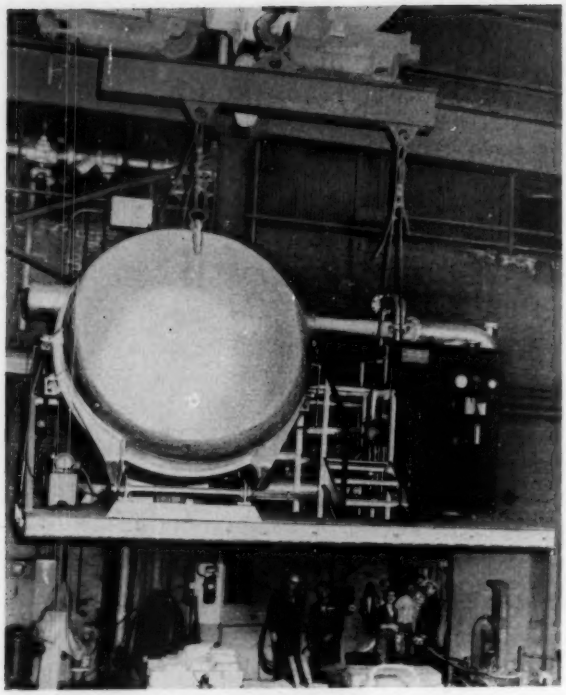


Fig. 5 — Unit suspended from crane hook and lifting beam.


## SUMMARY

A ladle vacuum degassing unit has been developed to handle 4000 to 5000 lb of metal, with no heat being added during the process. With optimum position of the unit in a shop, the degassing operation should be accomplished in less than 10 min. Temperature losses will be proportionately reduced by this improved time, and can further be reduced by greater ladle preheating. It is recognized that temperature losses become less as larger heats are handled. The package concept is feasible up to 20,000 to 30,000 lb heats.

The results of the 5000 lb unit predict the results for larger units. The unit illustrated has mobility, and may be moved from foundry to foundry for testing and evaluating purposes.



## ACKNOWLEDGMENT

The work done by the staff of A. Finkl and Sons in the operation of this unit, and in the presentation of results is gratefully acknowledged.


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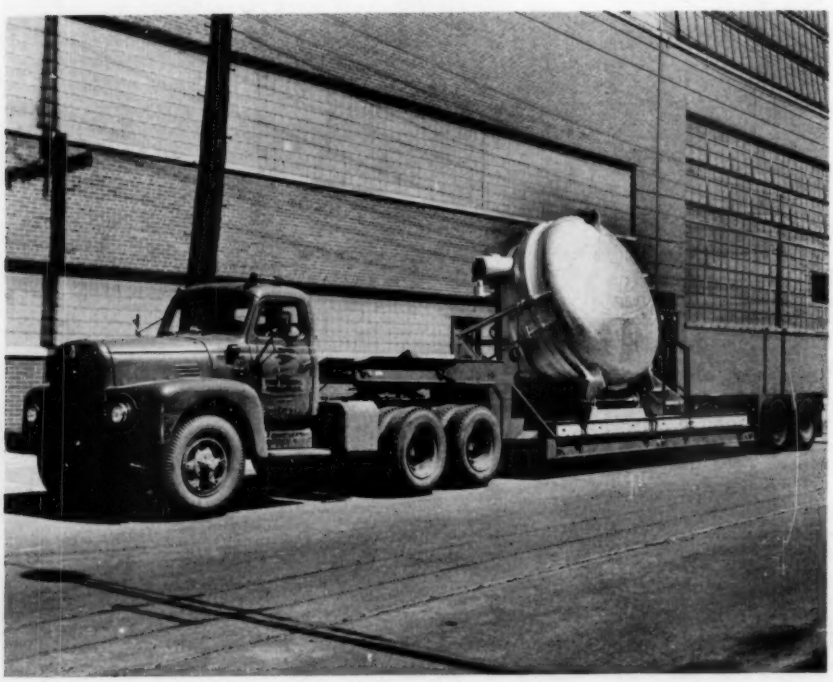



Fig. 6 — Unit moving on low-boy truck to next user.

# URANIUM ALLOY VACUUM INDUCTION MELTING

## *Some Theoretical and Practical Considerations*

by Philip S. Schaffer


### ABSTRACT

**Procedures of melting, refining and casting of depleted uranium alloys are developed utilizing vacuum induction melting techniques, a hand-rammed commercially pure magnesia crucible and a phosphate-bonded zircon sand mold. It is believed that this combination yields a melting and casting procedure that results in a reduction of residual gas content; second, minimizes crucible-melt reactions; and third, avoids possible metal-mold reactions.**

**Optimum results with this procedure are obtained by pouring the molten uranium under dynamic vacuum conditions to obtain a stream degassing effect, and then introducing a positive pressure of inert gas immediately upon filling the mold in order to prevent nucleation and possible evolution of gas during solidification. The overall processing procedure results in a reduction of residual gas concentration in the uranium alloy and produces castings with ductile qualities in the as-cast condition.**


### INTRODUCTION

In this investigation a processing schedule was developed for melting, refining and casting of depleted uranium alloys possessing strength combined with high ductility and density. Melting and refining is conducted in a magnesium oxide crucible mounted within a vacuum induction melting furnace.

A graphite crucible and graphite molds had been employed during previous investigations, but the graphite tended to dissolve in the molten charge especially when extensive refining times were required. Carbon contamination lowers mechanical properties, and thus limits potential usefulness of the uranium alloy product.

Accordingly, an objective was set to eliminate the graphite crucible and molds, to substitute a refractory oxide crucible and molds and then to evaluate the usefulness of this new procedure by conducting appropriate tests on alloys resulting from the newer technique. The actual investigation was divided into three parts – first, to select a melting procedure based on refractories that seemed theoretically advantageous and yet practical, second, to incorporate subject refractories in the melting and casting procedure, and third, to evaluate results of this new procedure by conducting gas analyses and mechanical property evaluation of the cast product.

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### EXPERIMENTAL PROCEDURE SELECTION

#### *Thermodynamic Considerations*

Uranium has been vacuum melted in graphite crucibles and cast into graphite molds, however, success is limited for carbon dissolves in the uranium with extended exposure time. A refractory oxide crucible and mold can be selected in lieu of graphite by considering the tendency for uranium to reduce the refractory. The oxide refractory should have the ability to withstand any tendency for this reaction to occur:

$$M_xO_y \text{ (crucible)} + yU \text{ (molten)} = xM \text{ (solution)} + yUO_2 \qquad (1)$$

Thermodynamic analysis of the reaction shows that absolute values of $F$, the free energy of formation of a compound or $\Delta F$, the change in free energy, cannot be calculated but may be defined relative to some standard or reference state. Generally, this may be considered the pure solid, liquid or vapor at one atmosphere pressure, dependent upon the temperature. A measure of the reacting tendency of a substance is therefore $F - F^\circ$, where $F^\circ$ is the standard free energy.

If the standard free energy change of the reaction $\Delta F^\circ$ is negative, then the reaction shows a tendency to proceed to the right as written. However, if $\Delta F^\circ$ is positive, then the reaction tends to proceed to the left. The system exists in a state of equilibrium, that is, there is no tendency to react, when $\Delta F^\circ = 0$.

The following oxides were considered for crucible

materials – alumina ($Al_2O_3$), beryllia (BeO), magnesia (MgO), silica ($SiO_2$), thoria ($ThO_2$) and zirconia ($ZrO_2$). The standard free energies of formation were calculated[1] and listed in Table 1 for the following six potential melt-crucible reactions (equations 2-7) at 1500 K (2241 F), 1600 K (2421 F), 1700 K (2601 F) and 1800 K (2781 F).

$$3\,U\ (\text{molten}) + 2\,Al_2O_3\ (\text{crucible}) = 4\,Al\ (\text{solution}) + 3\,UO_2 \qquad (2)$$

$$U\ (\text{molten}) + 2\,BeO\ (\text{crucible}) = 2\,Be\ (\text{solution}) + UO_2 \qquad (3)$$

$$U\ (\text{molten}) + 2\,MgO\ (\text{crucible}) = 2\,Mg\ (\text{gas}) + UO_2 \qquad (4)$$

$$U\ (\text{molten}) + SiO_2\ (\text{crucible}) = {}^{*}Si\ (\text{solution}) + UO_2 \qquad (5)$$

$$U\ (\text{molten}) + ThO_2\ (\text{crucible}) = Th\ (\text{solution}) + UO_2 \qquad (6)$$

$$U\ (\text{molten}) + ZrO_2\ (\text{crucible}) = Zr\ (\text{solution}) + UO_2 \qquad (7)$$

**TABLE 1 — STANDARD FREE ENERGY CHANGES FOR POTENTIAL CRUCIBLE-MELT REACTIONS GIVEN IN EQUATIONS 2-7**

| Crucible | $\Delta F°$ 2241 F | $\Delta F°$ 2421 F | $\Delta F°$ 2601 F | $\Delta F°$ 2781 F |
|---|---|---|---|---|
| $Al_2O_3$ | −22,300 cal. | −29,800 cal. | −33,200 cal. | −33,600 cal. |
| BeO | +18,900 | +17,000 | +16,300 | +16,600 |
| MgO | + 4,900 | − 2,440 | − 8,400 | −13,340 |
| $SiO_2$ | −51,400 | −52,800 | −52,950 | −52,650 |
| $ThO_2$ | +26,050 | +24,150 | +23,600 | +24,000 |
| $ZrO_2$ | + 3,850 | − 5,450 | − 5,700 | − 5,000 |

It is seen from Table 1 that the silica (m.p. 3142 F) reaction has the highest negative free energy indicating the greatest thermodynamic tendency for the melt-crucible reaction to proceed as written. Alumina (m.p. 3659 F) also yields a high negative $\Delta F°$ while beryllia (m.p. 4622 F) and thoria are thermodynamically satisfactory. Beryllia possesses good thermal shock resistance due to its moderate and uniform thermal expansion, but it is not as yet permissible for use at Watertown Arsenal because of the stringent health physics precautions necessitated by its toxic nature.

Thoria has the highest melting point of all the oxides (5972 ± 212 F) and a low vapor pressure at elevated temperatures; it is somewhat radioactive, expensive and has poor shock resistance. Zirconia (m.p. 4851 F) has less thermodynamic tendency than magnesia (m.p. 5072 F) to react at temperatures of 2601 F and higher. However, if the melt-crucible reaction should occur with zirconia, zirconium would enter into solution. Should the reaction occur with magnesia, and it is thermodynamically favorable at existing temperatures, the magnesium with its high vapor pressure, would volatilize out of solution.

If beryllia and thoria are eliminated because of precautions required in their use, selection is then limited to $ZrO_2$ and MgO. The thermodynamic evidence, combined with practical considerations such as existence of proved procedures for making rammed MgO furnace linings and zircon sand ceramic-bonded molds, pointed to eventual selection of MgO as a crucible material and a phosphate bonded zircon sand ceramic mold.

### *Volatilization*

Elements in the molten state, having a higher equilibrium vapor pressure than their vapor pressure above the melt, will volatilize out of solution as a function of temperature, vapor pressure, chamber pressure, time and composition. The vapor pressure of an element in solution is equal to the vapor pressure in the pure state multiplied by its activity in solution:

$$p_i = a_i\, p° \qquad (9)$$

where

$p_i$ = vapor pressure of element in solution.
$a_i$ = activity of element in solution.
$p°$ = vapor pressure of element in pure state.

If the activity is not known at small solute concentrations, the solution may be considered ideal; that is, the activity is equal to its mole fraction in accordance with Raoult's Law:

$$p_i = N_i\, p° \qquad (10)$$

where

$N_i$ = mole fraction of element in solution.

The succeeding relationship[2] can be used to estimate the rate of evaporation of an element from a molten alloy, with partial pressure being substituted for vapor pressure:

$$W_m = p_i \sqrt{\frac{M}{2\pi RT}} \qquad (11)$$

where

$W_m$ = maximum rate of evaporation, gm/cm²/sec.
$p_i$ = vapor pressure, dynes/cm².
$M$ = molecular weight, grams.
$R$ = universal gas constant, $8.3 \times 10^7$ ergs/deg C/mole.
$T$ = temperature, degrees Kelvin.

In refining of uranium, the losses due to vaporization are less than those predicted by this rate equation, because volatilization is obstructed by the accumulation of heavy metal vapor above the surface of the melt.[5] Also, the formation of dross mainly of uranium dioxide $UO_2$, at melting temperatures acts

*Although SiO (gas) is a possible product, it is assumed that reaction (5) is the predominant one since $\Delta F°$ of its reaction is more negative than the following reaction:

$$U\ (\text{molten}) + 2\,SiO_2\ (\text{crucible}) = 2\,SiO\ (\text{gas}) + UO_2 \qquad (8)$$

$$\Delta F°_{2241\,F} = -9{,}300\ \text{cal.}$$

as a vapor barrier. Finally, the rate determining step in volatilization may be the transport of the volatile element in solution to the melt surface. All impurities in uranium form compounds that are less dense than the uranium, thus they rise to the surface of the molten metal where they may be removed to some degree by vaporization.

Figure 1, a plot of uranium vapor pressure as a function of temperature, shows that at normal vacuum melting pressures of $10^{-2}$mm uranium vapor pressure will be approximately $10^{-4}$mm. Thus, appreciable uranium volatilization will not occur; however, alloying elements with vapor pressures in the $10^{-2}$mm range at the temperature involved here (2601 F) will tend to vaporize during refining. This is considered beneficial during refining if a high purity material is desired.

### *Dissociation*

Changing the pressure can influence the dissociation of a compound into its elemental forms, provided that at least one form is gaseous. From values of standard free energy of formation,[6] it can be calculated that lower pressure and/or high temperature will facilitate dissociation of metal oxides. Also, as the free energy of formation of the oxide becomes more negative, the oxide becomes more stable. The dissociation of uranium dioxide into its elements at 2601 F is given by the equation:

$$UO_2 \text{ (solid)} = U \text{ (liquid)} + O_2 \text{ (gas)} \qquad (12)$$

It is assumed that the pure oxide reacts and unit activity is used.

$$\Delta F^\circ_{2601\,F} = -RT \ln P_{O_2} = +95{,}000 \text{ cal.} \qquad (13)$$

$$P_{O_2} = 6.31 \times 10^{-13} \text{ atm} \qquad (13a)$$

It is seen that a partial pressure of oxygen less than $6.31 \times 10^{-13}$ atm is necessary in order for oxygen to be removed by dissociation, which is unlikely to be attained in any melting furnace currently in existence. Some oxygen removal is possibly achieved by a slight carbon monoxide boiling reaction during the early stages of refining, at which time some hydrogen and nitrogen may diffuse into these bubbles until the partial pressure of these gases in the bubbles are in equilibrium with the hydrogen and nitrogen in the melt.

Hydrogen and nitrogen removal is mainly accomplished by a combination process of dissociation of uranium hydride and nitride and diffusion of the hydrogen and nitrogen to the melt surface, assisted by the driving force of the dynamic vacuum environment.

The foregoing analysis leads to the selection of the magnesia furnace lining, the zircon sand ceramic mold and the expectation that uranium will not volatilize during extended refining; but that other materials with vapor pressures of $10^{-2}$mm and higher will tend to be eliminated from the melt. In addition, the analysis indicated that vacuum would not reduce the oxygen content of the melt, but that oxides and other separate distinct phases could quickly rise to the melt surface in view of the high density of the uranium melt.

## EXPERIMENTAL PROCEDURE

### *Vacuum Equipment*

The vacuum induction melting unit utilized in these investigations, as seen in Fig. 2, has a 1000 lb capacity for uranium. The unit consists of a horizontal chamber, 6 ft in diameter by 7 ft in length with standard accessories. These include alloy additions chamber; remote control unit for hydraulic tilting of furnace and vibratory feeder assembly; induction furnace; rotary mold table interchangeable with a stationary platform; valve assembly with immersion thermocouple, bridge-breaker and sampler; sight glasses with wipers and an inlet for introduction of an inert atmosphere. Cumulative pumping capacity is nearly 12,000 cfm under optimum conditions from mechanical roughing and holding pumps, an oil diffusion and three booster-type oil-vapor pumps; usually enabling a chamber pressure of 5-50 microns to be maintained over a molten bath of 650 lb of uranium. The power source is a motor generator set producing 175 kw operating at 400 volts and 960 cycles.

### *Melting and Refining*

A magnesia crucible (commercially pure –96.0 per cent MgO) is hand-rammed around a thin-walled steel shell to the desired dimensions. After slow drying, a wash heat of steel is melted under atmospheric pressure in the vacuum chamber in order to sinter the furnace lining. Upon reaching the molten state, the chamber is evacuated, and after the desired sintering time the melt is poured under vacuum. Steel is used in the wash heat as it can be superheated to a temperature greater than the maximum at which the uranium alloy will be refined, and thus the magnesium oxide crucible is fully sintered reducing the possibility of crucible reaction when melting uranium. The initial uranium heat dissolves a maximum contamination of 45 ppm of iron, while succeeding heats are free of dissolved iron.

Uranium "dingots" are used as melting stock and are produced by the direct thermit reduction of uranium tetrafluoride ($UF_4$, green salt) with magnesium in a bomb calorimeter. The method results in a metal of relatively high purity, that need not be remelted and cast into ingots. The shape of the dingot has been selected to promote the most favorable yield from the bomb reaction. The dingot and slag are separated and the dingot is then ready for vacuum processing after scalping to remove slag and refractory that may have adhered to the dingot surface.

A typical chemical and spectrographic analysis of the charge ingredients, depleted uranium dingot and molybdenum, appears in Table 2.

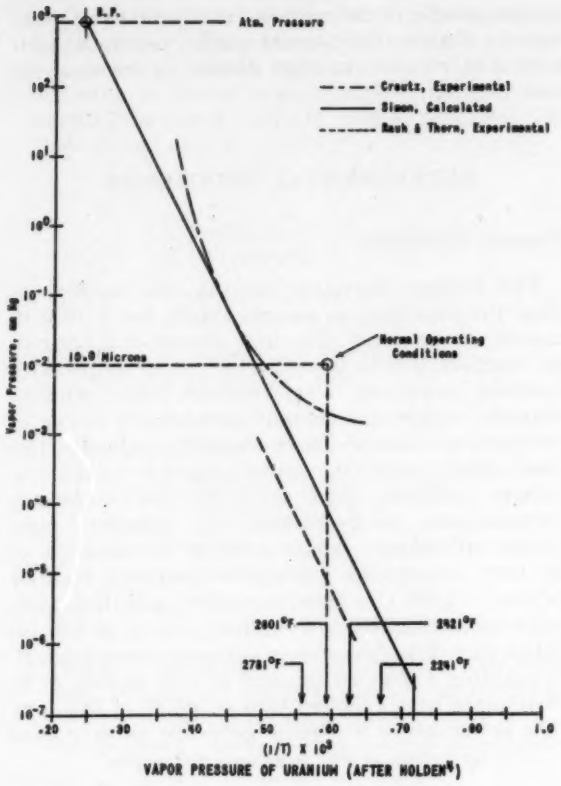



Fig. 1 — Vapor pressure of uranium.

TABLE 2 — ANALYSIS OF CHARGE INGREDIENTS

| Chemical | (ppm) | Uranium Dingot Analysis Spectrographic | (ppm) | | (ppm) |
|---|---|---|---|---|---|
| Carbon | 36 | Iron | 90 | Molybdenum | 5 |
| Oxygen | 42.1 | Aluminum | 20 | Vanadium | 20 |
| Hydrogen | 2.9 | Nickel | 31 | Manganese | 10 |
| Nitrogen | 16.3 | Silicon | 15 | Magnesium | 10 |
| Molybdenum Vacuum Fusion Gas Analysis (ppm) | | | | | |
| Oxygen | 53 | Hydrogen | 1 | Nitrogen | 7 |

Slow initial heating of the charge metal during an actual uranium processing cycle serves both to remove surface contaminants on the metals and further outgas the crucible. During meltdown, pressure gradually rises in the chamber with increasing temperature indicating further outgassing. This rise in pressure is caused by the inability of the limited pumping capacity to handle the increased gaseous load. If the temperature reaches 2800 F, the melt-crucible reaction occurs, and can be seen at the melt-crucible interface in the form of a violent boiling action which undoubtedly is a source of oxygen. Thus melt temperature is controlled within narrow limits. A minimum refining temperature of 2700 F is necessary because a layer of uranium dioxide dross and undissolved molybdenum forms at lower temperatures. The presence of such a film on the melt surface prevents volatilization of gaseous elements that normally can be pumped out of system.

During the refining process, much of the hydrogen, nitrogen and possibly oxygen is removed by desorption, shown by the reaction:

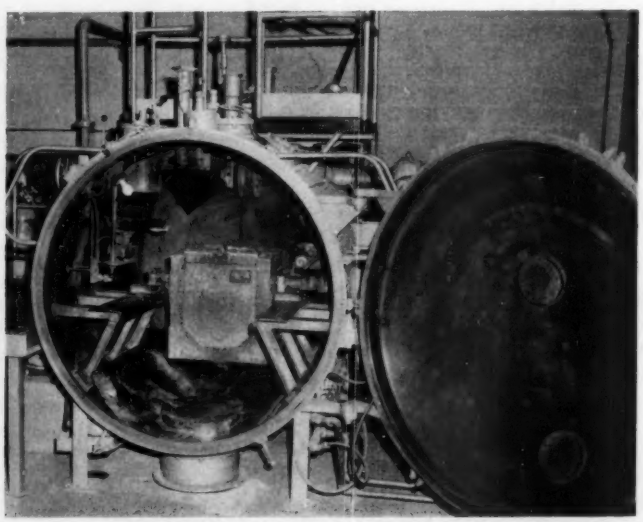


Fig. 2 — 1000 lb (uranium) capacity vacuum induction melting unit.

$$2\ \text{Gas (in solution)} = \text{Gas}_2\ \text{(vapor)} \qquad (14)$$

The inductive stirring mechanism continuously exposes new molten surfaces to vacuum and facilitates mass transfer of these elements from solution to their gaseous states. However, this mechanism is not as effective as in metals with lower density where a meniscus forms having a greater surface area exposed to vacuum.

### *Zircon Sand Mold*

A mold material which is comprised of a phosphate-bonded zircon system of the proportions shown in Table 3 is used in these vacuum processing investigations.

TABLE 3 — ZIRCON MIXTURE USED IN THESE INVESTIGATIONS

| Material | Mesh | Weight, % |
|---|---|---|
| | **Dry Mix** | |
| Zircon sand ($ZrO_2 \cdot SiO_2$) | 120 AFS | 80 |
| Zircon flour | 325 | 20 |
| | **Wet Mix** | |
| Aluminum hydroxide ($Al_2O_3 \cdot 3\ H_2O$) | 325 | 22.75 |
| Water | | 47.94 |
| Ortho Phosphoric Acid ($H_3PO_4$) | | 29.31 |

The mold mixture is prepared by the batch procedure:

1. 160 lb of zircon sand are mulled with 40 lb of zircon flour for 5 min in a high speed muller.
2. 3000 cc of water are added to 1424 gms. of aluminum hydroxide in a separate container.
3. 1834 gms. of phosphoric acid are stirred into the alumina-water mix until all ingredients are thoroughly combined in a smooth slurry.
4. This slurry is then added in portions to the dry mix and mulled for 3 min each.

A mono-aluminum phosphate binder is formed as a product of the reaction of phosphoric acid with aluminum hydroxide according to the equation:

$$^{*}Al_2O_3 + 6\,H_3PO_4 = 2\,Al\,(H_2PO_4)_3 + 3\,H_2O \qquad (15)$$

In this case 1424 grams of $AlO(OH) \cdot nH_2O$ represents 926 grams of $Al_2O_3$ or approximately 9 moles of alumina, and 1834 grams of $H_3PO_4$ solution represents 1592 grams or 18.75 moles of ortho phosphoric acid. Inasmuch as a nonionic species is formed as a product of the chemical reaction, the reaction can be considered to go to completion, thereby producing 6.25 moles of aluminum phosphate, $Al(H_2PO_4)_3$ or 2063 grams of binder. The reasons for the use of excess alumina are to make certain that all the phosphoric acid has reacted, and to promote the fastest possible combination of ingredients by allowing the acid to be exposed to a greater amount of alumina than is necessary to complete the reaction.

The mixture is hand-rammed to a density of approximately 3.15 g/cc over the pattern in a steel flask. After removal of the pattern, the mold is preheated for 8 hr at 500 F in a gas-fired oven. It is then transferred to a semi-muffle gas-fired furnace which is brought up to a temperature of 2000 F at a rate of approximately 250 F/hr and held for 4 hr at temperature, then furnace cooled. The slow heating and cooling cycle is used to avoid thermal stress and rupture of the mold.

### *Casting Procedure*

After the desired refining time and final chamber pressure of less than 50 microns are attained, the melt is poured at 2650 F into a bottom-gated phosphate-bonded zircon sand mold, which has been preheated at 500 F and allowed to cool down to room temperature under reduced pressure. The 2650 F pouring temperature represents a superheat of 420 F above the liquidus, which is as low a temperature as deemed feasible considering the gating system used.

When the molten alloy is cast into a mold, much of the low density insoluble impurity content segregates to the surface and remains in the top portion of the riser during solidification, where the impurities can be removed prior to remelting. Up to certain limitations, the greater the number of remelting cycles, the higher the purity of the uranium alloy that can be produced.

Three possible combinations of casting and solidification conditions are available:

1. Pour and solidify under vacuum.
2. Pour under vacuum and solidify under inert atmosphere.
3. Pour and solidify under inert atmosphere.

If dynamic vacuum conditions exist during pouring, that is, with the pumping system operating, a slight stream degassing effect is realized, as a large molten metal surface per unit volume is exposed to the vacuum, resulting in a slightly lower gas content.[7] If solidification also is accomplished under vacuum, the equilibrium partial pressures of the gases above the molten uranium in the mold are less than their actual partial pressures associated with the metal, favoring further gaseous evolution, possibly creating porosity in the casting.

This evolution is caused by the higher solubility of gases in liquid metal relative to the solid, provided the concentration exceeds the solubility limit in the liquid upon solidification. In a stagnant system, that is, with the vacuum chamber isolated and a positive pressure of an inert gas such as argon introduced, stream degassing will not be effected. Due to the increased barometric pressure created by the argon, and the pressure exerted by a column of molten metal at a given depth, the formation and nucleation of a gas bubble is unfavorable, thereby preventing evolution and promoting sound castings. The conditions finally decided upon are to pour under dynamic vacuum and allow solidification to occur under a positive pressure of an inert atmosphere.

* $Al_2O_3$ represents 2 $AlO(OH) \cdot nH_2O$.

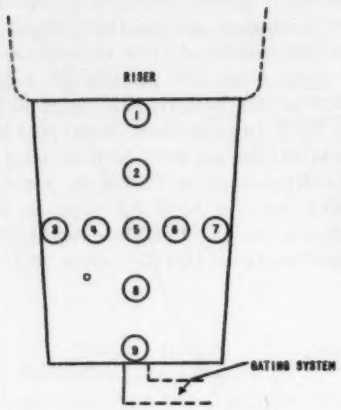



Fig. 3 — Location of specimens obtained for segregation studies from a vertical casting section.

| | O (ppm) | H (ppm) | N (ppm) | C (ppm) | Mo (w/o) | Fe (ppm) |
|---|---|---|---|---|---|---|
| 1 | 29 | 0.5 | 13 | 45 | 8.33 | 87 |
| 2 | 22 | 0.7 | 10 | 40 | 8.33 | 84 |
| 3 | 25 | 1.7 | 8 | 45 | 8.05 | 78 |
| 4 | 19 | 0.6 | 10 | 45 | 8.12 | 90 |
| 5 | 27 | 1.2 | 14 | 50 | 8.35 | 85 |
| 6 | 24 | 0.7 | 14 | 45 | 8.18 | 91 |
| 7 | 21 | 1.2 | 4 | 55 | 8.05 | 91 |
| 8 | 22 | 0.9 | 10 | 40 | 8.33 | 94 |
| 9 | 18 | 0.7 | 7 | 45 | 8.27 | 91 |

Four Uranium - 8 w/o Molybdenum heats ranging from 350 to 950 lb were cast using the aforementioned melting, refining, casting and solidification conditions. One casting in particular was subjected to thorough metallurgical evaluation in order to assess the overall effectiveness of the procedure. The method of testing was:

1. Nine samples were removed from the casting in accordance with the sketch shown in Fig. 3. These were subjected to conventional analysis for alloy content and vacuum fusion analysis for determination of residual gas content.
2. As-cast mechanical properties were determined in both longitudinal and transverse directions from bars removed from the lower section of the casting.
3. Microstructures were examined to assess the general cleanliness level of the cast material.
4. Special attention was given to the mold for this casting. For example, during mold firing a curve of per cent weight loss of the mold as a function of time was developed in order to insure that moisture had been completely removed from the mold prior to casting operations. In addition visual examination of the casting surface was conducted in order to determine in a general way whether or not a mold-metal reaction occurred during casting and solidification.

## RESULTS AND DISCUSSION

### *Residual Gas Content and Segregation*

The chemical analysis results are presented in Fig. 3, and these can be compared with the analysis of the charge materials shown in Table 3. Final gas concentrations of oxygen, hydrogen and nitrogen in the casting average 23, 0.9 and 10 ppm, which represent a reduction of 47.7, 67.3 and 35.8 per cent, respectively, from the charge materials. Variations of oxygen, hydrogen and nitrogen content with location in the casting is not considered significant. The reduction in hydrogen and nitrogen can easily be explained by the vacuum environment present during the melting and refining cycles, and by the stream degassing effect obtained during the pouring operation.

However, the reduction of oxygen content, although especially significant, is more difficult to rationalize. Since a decrease in oxygen does result, it can be concluded that both the melt-crucible and metal-mold reactions do not occur. Therefore, the crucible and the mold material are entirely satisfactory for containing and receiving molten uranium. If either of the aforementioned reactions had taken place, an oxygen increase would have resulted. The actual decrease in oxygen concentration cannot be explained by the reduction of uranium dioxide as this was deduced to be thermodynamically impossible under the existing conditions in the vacuum system.

However, this reduction can be interpreted on the basis that uranium dioxide has an opportunity to float to the melt surface, thereby lowering the overall oxygen content. This point is substantiated by the presence of an oxide crust on the surface of the melt during refinement.

The iron content existing in the final casting is the same as that of the charge material, indicating that the wash heat does not contaminate the uranium with iron after the initial heat. The molybdenum throughout the casting, although within the prescribed range for this alloy, does exhibit a perceptible tendency for gross segregation. Finally, the slight increase in carbon content is unexplained, for presumably carbon did not come into contact with molten uranium during any stage of processing.

### *Mechanical Properties*

As-cast mechanical properties were determined in both longitudinal and transverse specimens from the lower portion of the casting, and these results are presented in Table 4.

TABLE 4 — MECHANICAL PROPERTIES OF U-8 W/O MO ALLOY IN AS-CAST CONDITION

| 61RL121 | Y.S. 0.1%, psi | Y.S. 0.2%, psi | UTS, psi | Elong., % | Red. Area, % | Impact (−40 F), ft-lb |
|---|---|---|---|---|---|---|
| longitudinal | 122,500 | 127,000 | 128,000 | 5.7 | 13.0 | 3.0 |
| | 121,500 | 126,000 | 126,500 | 7.9 | 13.5 | 3.4 |
| | 126,500 | 128,000 | 128,400 | 9.5 | 21.8 | 3.2 |
| transverse | 117,500 | 123,000 | 123,500 | 5.7 | 10.9 | 3.4 |
| | 125,000 | 126,500 | 126,800 | 7.5 | 15.6 | 3.2 |

At the as-cast strength level of 126,000 psi, average ductility, measured by reduction of area, is approximately 15 per cent indicating that the material is free from gross porosity due to gaseous evolution or solidification shrinkage. The properties are reason-

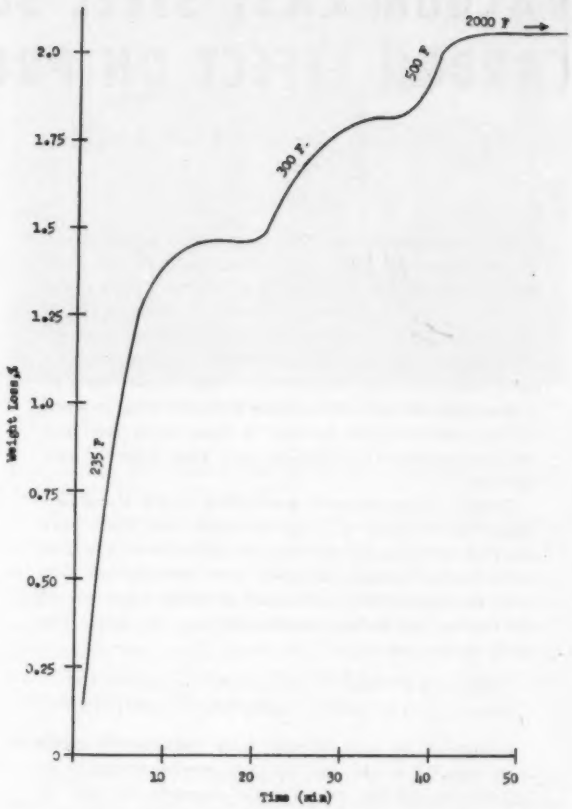



Fig. 4 — Weight loss of zircon mold as a function of time.

ably uniform, demonstrating that a general soundness level was obtained throughout the casting and this was substantiated by visual observation of the casting. During the course of experimentation, it was determined that the uranium alloy freezes in a manner similar to steel, as gross shrinkage is approximately three per cent and the mode of solidification is by wall growth rather than by the development of a mushy solidification zone.

### *Microstructure and Examination for Cleanliness*

Metallographic examination revealed the general cleanliness level of the cast alloys. Spherical nonmetallic inclusions present probably did not detract from mechanical properties, for the stress raiser effect of such inclusions is not as marked as the stringer-type that often occur at cast grain boundaries. However, the size and frequency of these spherical inclusions indicated an opportunity for further alterations in the melting and casting procedures, in order to improve cleanliness and thus obtain a moderate increase in properties.

### *Evaluation of Mold Performance*

The presence of moisture and organics in the mold at the time of vacuum pouring can easily cause mold-metal reaction. Accordingly, the experiment to determine mold weight loss as a function of firing time is considered significant; these results being presented in Fig. 4. Preheating the mold to 500 F results in a loss of 2.06 weight per cent. At a temperature of 2000 F, a cumulative loss of 2.36 per cent is achieved, but overall mold density increases because of sintering occurring at high temperatures.

Now a ceramic-type body, the mold does not react to any measurable extent with the molten uranium. This is apparent from visual observation of the condition of the mold wall, the casting surface and the residual oxygen content near the casting surface. From a casting point of view, zircon sand is a high temperature refractory which does not undergo phase transformation within the temperature limits of this cast metals procedure. Zircon is also beneficial to surface finish because of its naturally occurring round grain structure which promotes tight, smooth mold walls and uses less binder than angular or subangular materials.

### *Future Work*

These results infer a number of opportunities for future improvement of the vacuum technique and the further examination of the uranium alloys produced with this successful furnace lining and molding material. Attention can be given to improving the general cleanliness level of the alloys, and then conducting formalized experiments on mechanical property evaluation of both as-cast and heat-treated uranium alloys.

## ACKNOWLEDGMENTS

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# VACUUM CAST STEEL SULFUR AND CARBON EFFECT ON PROPERTIES

by J. B. Dabney, M. C. Flemings and H. F. Taylor


## ABSTRACT

An investigation was conducted on the effect of sulfur and carbon on the tensile properties of cast 4300 series vacuum melted high strength steel. Final sulfur contents of heats poured ranged from 0.004 to 0.020 per cent; carbon contents were from 0.40 to 0.60 per cent. Most effective desulfurization was obtained in heats melted at less than $5 \times 10^{-3}$ mm Hg pressure, in a magnesia crucible with a lime fluorspar plug sintered to the inside crucible bottom. In these heats, final sulfur concentrations of 0.004 per cent were achieved consistently.

Tensile properties were determined of the sound cast metal, heat treated to a high strength level. Data were analyzed statistically, and the analysis showed the ultimate tensile strength and per cent reduction in area could be satisfactorily estimated as linear functions of the carbon and sulfur concentrations in the steel. The relationships are:

$$UTS = [2.73(\%C) + 10.76(\%S) + 1.44] \times 10^5$$
$$\%RA = [-1.37(\%C) - 13.56(\%S) + 0.99] \times 10^2$$

Reduction in area of the fully desulfurized 4340 steel was 37 to 40 per cent at tensile strengths of 250,000 to 260,000 psi. Tensile strengths as high as 317,000 psi were obtained at higher carbon contents.


## INTRODUCTION

Recent work has indicated that many factors are important in determining the ductility of high strength, low alloy steel under tension and impact testing. Factors which limit the properties of air melted steel include microporosity, as-cast grain structure, the shape and number of inclusions and the relative amounts of impurities such as sulfur and phosphorous.[1-5] Significant improvement in ductility of steel can be obtained by vacuum melting and refining, and this may be attributed to:

1. A cleaner steel due to lower oxygen content.
2. Less microporosity due to reduced concentration of residual gases.
3. A lower concentration of certain impurities as a result of volatilization.

Present commercial vacuum melting techniques do not remove significant amounts of sulfur;[6,7] to achieve optimum quality, the initial charge material must be as free from sulfur as possible.[6]

Increased ductility of air melted, high strength, low alloy steel has been reported to accompany desulfurization from 0.023 to 0.008 per cent,[3] and from 0.010 to 0.003 per cent.[5] It should also be possible to improve vacuum melted steel by desulfurization to extremely low sulfur levels. This investigation was undertaken with the dual objectives of (1) investigating a method of removing sulfur from steel under vacuum and (2) studying the effect of this desulfurization on the properties obtained thereby.

The method of desulfurization employed a lime fluorspar mixture as part of the crucible material. The method was similar to one previously used for desulfurization of steel melted by induction in air.[8,9] In the air melted steel sulfur contents as low as 0.001 per cent were reported with the achievement of adequate deoxidation. Under vacuum, excellent deoxidation is achieved by small amounts of carbon,[10,11,12] so the only other addition needed is the desulfurization agent.

## LITERATURE SURVEY

Purification during vacuum refining results primarily from the distillation of impurities from the surface of the liquid metal. However, in the case of sulfur, distillation is not rapid enough under ordinary conditions to accomplish substantial desulfurization.[10,18] Additions of carbon and/or silicon increase the rate of desulfurization. There are two likely reasons for this; first, additions of carbon lead to the formation of carbon-sulfur compounds which have higher vapor pressures than gaseous sulfur,[14] and second, both carbon and silicon increase the activity of sulfur in iron.[15a] There is considerable disagreement among individual investigators,[7,16] with regard to the extent to which desulfurization in vacuum (by distillation) proceeds. In view of published data, it is unlikely that more than 50 per cent desulfurization can be achieved readily by vacuum refining steel melts having low initial sulfur contents. Furthermore, 0.005 per cent sulfur appears to be the minimum concentration attainable by vacuum refining, and this level of purification requires that an electrolytically refined charge material be used.

The effect of sulfur in vacuum melted steels is greater than in most air melted steels because of the tendency for film-like, Type II sulfides to occur in the former.[2] This is due to the low oxygen concentration in these steels, and to the absence of substantial quantities of deoxidizers such as aluminum which cause the less harmful, Type III sulfides to form.[2,17,18]

To attain optimum mechanical properties and reproducibility, steel must contain minimum amounts of inclusion forming elements such as sulfur and oxygen. This is particularly true at high strength levels. Other factors, such as microporosity and shrinkage, are also important, but the real advantages of their elimination cannot be fully realized unless the effects of inclusions are also eliminated or reduced to a minimum. While oxygen concentration

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can be readily reduced to low values by vacuum, desulfurization presents certain difficulties which are not encountered in air melting.

Slags, as such, are difficult to employ effectively under vacuum because of handling and slag removal problems; also the presence of a slag cover may hinder distillation of gases and volatile elements. A solution to this problem lies in using a refractory material for the crucible, or as a crucible lining or wash, which has a strong desulfurizing power.[8,9,19]

### *Methods of Producing Low Sulfur Steel*

Other methods of producing low sulfur steel in vacuum have been attempted. Additions of cerium in the form of mischmetal have been reported to reduce sulfur from 0.017 to 0.006 per cent, and calcium can be added for the same purpose in the form of a master alloy of calcium iron.[6] However, the rate of vaporization of these desulfurizing elements is violent, so a partial pressure of inert gas is required; also, the master alloy is expensive, and the reaction product remains in the liquid metal. Currently, the most widely used practice for producing vacuum melted, low sulfur, steel is to employ electrolytically pure materials, or materials which have been previously desulfurized in air; in the latter case, sulfur contents as low as 0.003 per cent are obtainable, but essentially no desulfurization takes place under vacuum and the cost of the initial charge material adds considerably to the cost of producing the steel.

## EXPERIMENTAL PROCEDURE

Two series of vacuum melted heats were produced in this study. One series was carried out in magnesia crucibles, while the other was carried out in magnesia crucibles containing a ½-in. layer of a CaO-10 per cent $CaF_2$ mixture sintered to the crucible bottom. It was anticipated that desulfurization of heats melted in the magnesia crucibles would occur only by distillation, whereas in the case of the series melted in crucibles with the lime layer, significant desulfurization was expected from slag-metal reaction. The calcium fluoride addition to the lime has been reported necessary to promote desulfurization;[8,9] it also facilitates sintering the lime layer.

Several heats were made in each type crucible using armco iron base metal; the remaining heats were made using electrolytic iron. A vacuum induction melting and casting unit was employed (Fig. 1); heats were approximately 10 lb for each experiment.

The pattern for the test casting was parted vertically, and consisted of a central sprue or downgate from which projected two pairs of arms (Fig. 2). Each arm was designed to yield one tensile or impact specimen giving a total of four test specimens. The test bars were heavily risered and strongly chilled to promote good feeding by directional solidification. The mold material (Table 1) was mullite bonded with ethyl silicate. Molds were fired at 1800 F (960 C) 3 hr before inserting steel chills and assembling. Before each heat a mold was preheated to 400 F (205 C), placed into the vacuum furnace and immediately



Fig. 1 — Vacuum melting and casting furnace.

evacuated. No difficulties were encountered with gas evolution from the mold.

Alloying additions were in the form of high purity ferro-alloys, and were the same for all heats on both series. The electrolytic iron was first air melted and cast into ingots to facilitate charging. The armco iron was charged as squares which had been sand blasted to remove the oxide scale. All materials were charged in the crucible except ferromanganese, which was added late in the refining cycle.

### *Melting Practice*

When the metal began to melt, the vacuum pumps were closed and helium was bled into the system until the pressure reached 50 mm Hg. This prevented violent boiling, and, after meltdown, the pressure in the chamber was reduced gradually to maintain a steady, moderate, boiling action. This boiling

TABLE 1 — MOLD MATERIAL

| Ethyl-Silicate Solution | |
|---|---|
| Ethyl-silicate, % | 52.5 |
| Ethyl-alcohol, % | 42 |
| Water, % | 5 |
| HCl (conc.), % | 0.5 |

All per cents by volume.

**Molding Slurry**

1000 cc Ethyl-silicate solution.
5 lb 100 mullite refractory.
9.3 grams MgO powder.

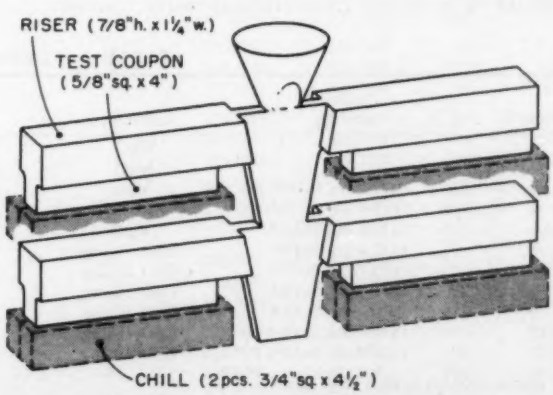



Fig. 2 — Test casting showing four test coupons.

due to deoxidation by carbon usually ceased after several minutes at a pressure of about 0.2 mm Hg, and two heats were poured at this pressure.

The remainder of the heats were evacuated to a refining pressure of $1 \times 10^{-3}$ to $5 \times 10^{-3}$ mm Hg and held at this pressure for approximately 20 min. After the refining period, the pressure was again raised to 50 mm Hg and the ferromanganese added. The pressure was then reduced to 0.1 mm Hg and held for 3 min. At this point, helium was bled into the chamber and the metal was cast. The helium pressure was allowed to reach 200 mm Hg to minimize any possibility of gas absorption by the metal from the mold, and the test castings were allowed to solidify under this pressure.

Radiographs of the unmachined test bar blanks and of ⅛-in. machined cross-sections were taken to determine the degree of soundness. Microradiographic analyses of the castings were carried out according to the method described by Uram, Flemings and Taylor[1] (x-raying of specimens approximately 0.005 in. thick). In all cases, the microporosity observed was nil. Chemical analyses for sulfur, phosphorus and alloying elements were obtained for all heats. Chemical analyses are presented in Table 4 for the various heats. Oxygen and nitrogen analyses were carried out on several preliminary heats, and since the results were uniform and agreed well with published data for vacuum melted steels, this step was discontinued.

Tensile test blanks were heat treated to develop high strength properties before being machined into 0.252 in. diameter tensile bars. Details of the heat treatment are presented in Table 2. The bars were pulled at room temperature in a standard tensile test machine. Yield strength (0.2 per cent offset), ultimate tensile strength, per cent elongation and per cent reduction in area were measured.

## RESULTS AND DISCUSSION

### *Desulfurization*

The first series of heats were melted in magnesia crucibles, without addition of lime; desulfurization of these heats was studied primarily to verify data reported in the literature, which indicated that limited desulfurization should result from the volatilization of sulfur and sulfur-containing compounds.[9,10,13,20] In this case, no desulfurization would be expected from crucible-metal reaction.

Heats in this series poured at 0.2 mm Hg (heats 1 and 2) had final sulfur concentrations that agreed well with the estimated sulfur contents of the initial charges, indicating that essentially no desulfurization took place at this pressure. The final analyses of the two heats were 0.020 per cent sulfur (for the heat which was formulated using armco iron) and 0.013 per cent sulfur (for the heat formulated using electrolytic material); estimates of amount of desulfurization obtained in subsequent heats were made using these figures as reference points (Table 3).

Desulfurization of the remaining melts in the series, using magnesia crucibles, ranged from 30 to slightly over 50 per cent. These heats were refined at pressures of $1 \times 10^{-3}$ to $5 \times 10^{-3}$ mm Hg for ½-hr. Minimum sulfur concentrations of 0.010 per cent and 0.006 per cent were found, respectively, for the heats using armco and electrolytic iron as charge material. Results agree well with those in the literature, and indicate that the mechanism of desulfurization involved in this case is the distillation of sulfur and carbon-sulfur compounds.[10] The final sulfur content of 0.006 per cent, obtained for the heat formulated with electrolytic iron, is in agreement with sulfur contents obtained commercially in vacuum melted 4340 steel, using an electrolytic charge.[20,21]

### *Magnesia Crucible Melting Practice*

The second test series was melted in a magnesia crucible which had a mixture of CaO-10 per cent $CaF_2$ sintered to the crucible bottom. Final sulfur concentrations of all heats were 0.004 to 0.005 per cent after refining at $1 \times 10^{-3}$ to $5 \times 10^{-3}$ mm Hg. The majority of values were 0.004 per cent, and the rest were within the experimental error of the analysis. Differences in initial sulfur contents resulting from use of electrolytic iron for some heats and

TABLE 2 — HEAT TREATMENT

| Temp., F (C) | Time, hr | Cooling Process |
|---|---|---|
| 2200* (1205) | 2 | air cool** |
| 1750* ( 955) | 2 | air cool |
| 1600 ( 871) | 2 | Furnace cool to 1450 F (788 C) |
| 1450 ( 788) | 3 | oil quench |
| 400 ( 204) | 2 | water quench |
| 400 ( 204) | 2 | water quench |

* Carried out in argon atmosphere.
** Cooled to room temperature unless otherwise noted.

TABLE 3 — DESULFURIZATION DATA

| Heat | Pressure, mm Hg | Refractory | Charge | Sulfur initial | Sulfur final | Desulfurization, % |
|---|---|---|---|---|---|---|
| 1 | 0.2 | MgO | armco | 0.020 | 0.020 | 0 |
| 2 | 0.2 | MgO | electro | 0.013 | 0.013 | 0 |
| 3 | 1x10-3 to 5x10-3 | MgO | armco | 0.020 | 0.010 | 50 |
| 4 | 1x10-3 to 5x10-3 | MgO | armco | 0.020 | 0.012 | 40 |
| 5 | 1x10-3 to 5x10-3 | MgO | electro | 0.013 | 0.006 | 54 |
| 6 | 1x10-3 to 5x10-3 | CaO lining | armco | 0.020 | 0.004 | 80 |
| 7 | 1x10-3 to 5x10-3 | CaO lining | armco | 0.020 | 0.004 | 80 |
| 8 | 1x10-3 to 5x10-3 | CaO lining | armco | 0.020 | 0.004 | 80 |
| 9 | 1x10-3 to 5x10-3 | CaO lining | electro | 0.013 | 0.004 | 70 |
| 10 | 1x10-3 to 5x10-3 | CaO lining | electro | 0.013 | 0.005 | 62 |
| 11 | 1x10-3 to 5x10-3 | CaO lining | electro | 0.013 | 0.005 | 62 |

armco denotes armco iron base.
electro denotes electrolytic iron base.

armco iron for others apparently had no effect on the final sulfur content.

Lower resultant concentrations might be expected if the area of the lime mixture in contact with the melt were increased and if the magnesia, which is less stable and may therefore oxidize the liquid metal at a greater rate than lime, were isolated from the liquid metal. However, even with these modifications, it is not likely that sulfur concentrations below 0.001 per cent could be attained with refining times of reasonable length. Also, it is not likely that further desulfurization would occur in the system investigated with moderate increases in refining times. The time required for completion of the reaction under similar conditions is extremely rapid.[8,9]

Polished sections of several castings were examined to determine the quantity and types of inclusions per cent. At sulfur concentrations greater than 0.010 per cent, Type II sulfide inclusions were found. This was expected because vacuum melted steels are particularly prone to the formation of these film-like inclusions. The number and severity of these sulfides increased with increasing sulfur concentrations. In castings containing less than 0.010 per cent sulfur, only spherical inclusions were observed. There was no distinguishable difference in either the number, size or shape of the inclusions observed in this material. It is believed that more of these inclusions are nearly pure silica.

### *Mechanical Properties*

Results of tensile tests are recorded in Table 5. As the material tested covered a range of carbon and sulfur concentrations, the data were analyzed by statistical procedures.[22] The method of least squares was used, and equations were developed which estimated the per cent reduction in area and the ultimate tensile strength in terms of the two independent variables, carbon and sulfur. These equations define planes which relate carbon and sulfur, with (1) tensile strength and (2) per cent reduction in area.

By holding carbon or sulfur constant, the expected value of these properties can be plotted against either variable. This is equivalent to intersecting the plane defined by the equation by a second plane. Thus, straight line relationships must be obtained. Plots of the deviations of experimental data from the estimated values were constructed for both the ultimate tensile strength and the per cent reduction in area against both carbon and sulfur. These plots showed only random scatter and no apparent relationships other than linear. Data from three test bars were not used in the analysis because of apparent defects.

It should be noted that the relationships developed by the statistical analysis are limited to the ranges of compositions studied. These relations may, and in some instances must, break down outside these ranges. For instance, negative values of reduction in area are not possible, and so the straight lines of Fig. 5 cannot continue below the abscissa.

A determination of the effect of carbon on the tensile strength was carried out for carbon concentrations ranging from 0.40 to 0.60 per cent. The equation relating the estimated tensile strength to the concentrations of carbon and sulfur was derived:[22]

$$UTS_{(est)} = [\,2.73\,(\%C) + 10.76\,(\%S) + 1.44\,] \times 10^5$$

Plots of the estimated tensile strength with respect to carbon and sulfur are illustrated in Fig. 3. As is indicated in the above equation and in Fig. 3, the sulfur content has a small positive effect on the tensile strength. The agreement of the estimated tensile strength with the experimental data is extremely good, as is shown in Fig. 4. Statistically, carbon alone is associated with 76 per cent of the variance of the tensile strength. Carbon and sulfur together are associated with 82 per cent. Therefore, only 18 per cent is attributable to randomness and variables such as grain structure, microporosity and tensile testing.

**TABLE 4 — CHEMICAL ANALYSIS**

| Heat | Analysis, % | | | | | | | |
|---|---|---|---|---|---|---|---|---|
| | C | Si | Mn | Ni | Cr | Mo | P | S |
| 1 | 0.44 | 0.35 | 1.08 | 1.79 | 0.76 | 0.26 | 0.008 | 0.020 |
| 2 | 0.42 | 0.30 | 0.93 | 1.83 | 0.79 | 0.29 | 0.008 | 0.013 |
| 3 | 0.43 | 0.31 | 0.81 | 1.83 | 0.68 | 0.28 | 0.006 | 0.010 |
| 4 | 0.48 | 0.32 | 0.83 | 1.83 | 0.79 | 0.28 | 0.006 | 0.012 |
| 5 | 0.46 | 0.30 | 0.99 | 1.87 | 0.77 | 0.27 | 0.005 | 0.006 |
| 6 | 0.49 | 0.30 | 0.93 | 1.83 | 0.74 | 0.28 | 0.005 | 0.004 |
| 7 | 0.62 | 0.30 | 0.96 | 1.83 | 0.78 | 0.25 | 0.005 | 0.004 |
| 8 | 0.55 | 0.30 | 0.93 | 1.79 | 0.82 | 0.26 | 0.005 | 0.004 |
| 9 | 0.40 | 0.25 | 0.87 | 1.83 | 0.83 | 0.23 | 0.007 | 0.004 |
| 10 | 0.40 | 0.29 | 0.37 | 1.83 | 0.80 | 0.26 | 0.005 | 0.005 |
| 11 | 0.50 | 0.30 | 0.90 | 1.83 | 0.82 | 0.27 | 0.004 | 0.005 |

**TABLE 5 — RESULTS OF TENSILE TESTS**

| Heat | Specimen | Yield Strength (0.2% Offset), psi | Tensile Strength, psi | Red. in Area, % | Elong., % |
|---|---|---|---|---|---|
| 1 | 3 | 247,000 | 287,000 | 10.0 | 10.0 |
| 2 | 1 | 230,000 | 270,000 | 17.4 | 9.0 |
| | 2 | 200,000 | 270,000 | 16.6 | 8.0 |
| | 3 | 227,000 | 287,000 | 13.0 | 9.0 |
| 3 | 1 | 245,000 | 281,000 | 29.4 | 10.0 |
| 4 | 1 | 213,000 | 279,000 | 27.4 | 12.0 |
| | 2 | 217,500 | 277,500 | 28.0 | 11.0 |
| | 3* | 217,000 | 277,000 | 4.8 | — |
| 5 | 1 | 211,000 | 272,000 | 30.2 | 13.5 |
| | 2 | 203,000 | 263,400 | 33.6 | 11.5 |
| | 3 | 208,000 | 273,000 | 30.2 | 12.5 |
| 6 | 1 | 220,000 | 299,000 | 23.2 | 11.5 |
| | 2 | 211,500 | 289,000 | 30.2 | 12.5 |
| | 3 | 211,000 | 291,000 | 28.0 | 9.0 |
| 7 | 1* | 204,000 | 284,000 | 3.2 | 3.0 |
| | 2* | 206,000 | 286,000 | 4.0 | 6.0 |
| | 3 | 207,000 | 317,000 | 8.6 | 7.0 |
| 8 | 1 | 197,500 | 296,500 | 8.6 | 6.0 |
| | 2 | 203,000 | 299,000 | 7.0 | 5.5 |
| | 3 | 205,000 | 292,000 | 22.4 | 7.5 |
| 9 | 1 | 198,500 | 250,000 | 38.0 | 12.0 |
| | 2 | 221,000 | 258,000 | 39.4 | 11.5 |
| | 3 | — | 247,000 | 38.6 | 10.0 |
| 10 | 1 | 222,500 | 255,000 | 37.2 | — |
| | 2 | 213,000 | 253,000 | 40.4 | 12.5 |
| | 3 | 192,000 | 262,000 | 37.0 | 12.0 |
| 11 | 1 | 229,000 | 289,000 | 24.0 | 10.5 |
| | 2 | 219,000 | 284,000 | 19.6 | 7.0 |
| | 3 | 210,500 | 284,000 | 29.6 | 10.5 |

*Premature brittle fracture.

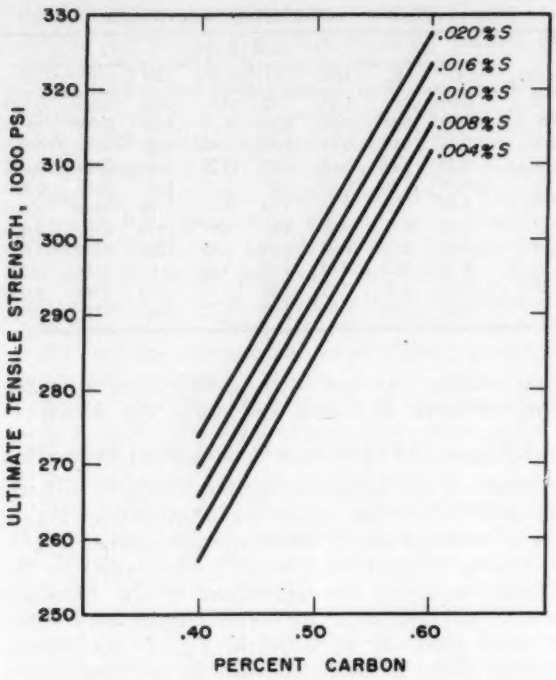



Fig. 3 — Estimated ultimate tensile strength as a function of carbon and sulfur, as based on experimental data.

### *Sulfur Effect*

The effect of sulfur may possibly be a consequence of the least squares procedure coupled with this particular experiment. Repeated controlled experiments would be necessary to establish the correctness of the weight given to sulfur. An equation relating carbon and sulfur concentrations to the per cent reduction in area was similarly determined:

$$\%\,RA_{(est)} = [-1.37\,(\%\,C) - 13.56\,(\%\,S) + 0.99] \times 10^2$$

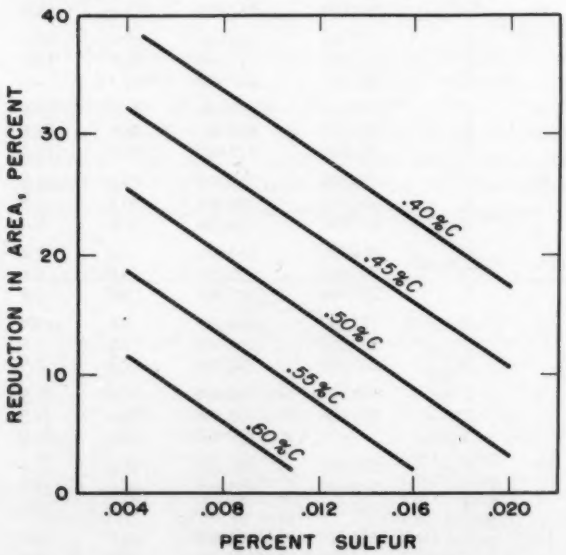



Fig. 5 — Estimated per cent reduction in area as a function of sulfur and carbon contents, based on experimental data.

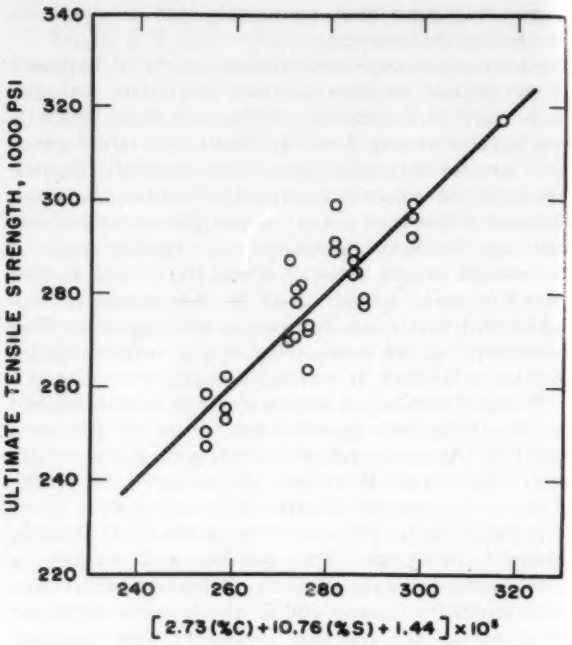



Fig. 4 — Correlation of estimated ultimate tensile strengths with experimental data.

This is plotted in Fig. 5 against per cent sulfur for several carbon concentrations. Both carbon and sulfur are shown to have pronounced effects on the reduction in area. There is more scatter in this data than was found for the relation to tensile strength, as is evident in Fig. 6. In this case the carbon and sulfur together can be related to 68 per cent of the variance in the values for reduction in area. Forty per cent can be attributed to the carbon alone, and 32 per cent must be accounted for by other variables.

Yield strengths, as measured by the 0.2 per cent offset method, were determined, and are included in

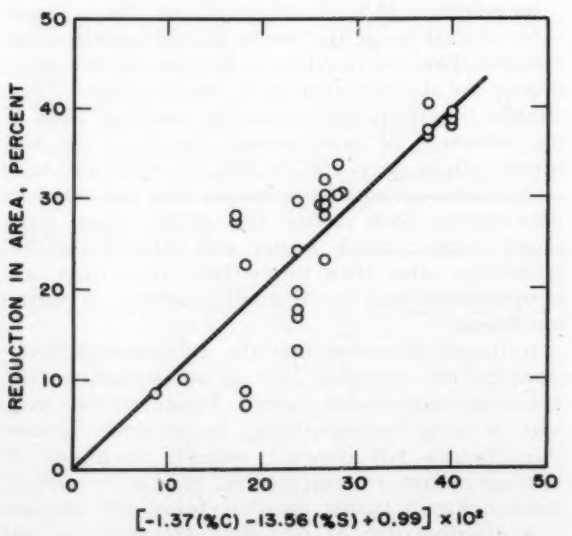



Fig. 6 — Correlation of estimated per cent reduction in area with experimental data.

Table 5. No marked correlation with either carbon or sulfur content or with tensile strength was observed. This agrees with published data concerning air melted (4300 series) steels.[20]

Elongation values were measured as the per cent elongation in one in. These varied from a low of 3.0 per cent to a high of 13.5 per cent, and were more or less proportional to the reduction in area values in that high values for the latter were accompanied by high elongations. Because of the range of the reduction in area values, this measurement was felt to be a more sensitive index of the effects of chemical variables on tensile strength and tensile ductility.

## CONCLUSIONS

In a series of alloys melted in magnesia crucibles, desulfurization was negligible before refining, but after holding at $1 \times 10^{-3}$ to $5 \times 10^{-3}$ mm Hg for 30 min, sulfur contents were reduced 40-54 per cent. In one heat, sulfur was reduced from 0.020 to 0.010 per cent; in another, sulfur was reduced 0.013 to 0.006 per cent (both after refining at $1 \times 10^{-3}$ to $5 \times 10^{-3}$ mm). These sulfur reductions are attributable to distillation of sulfur and compounds of carbon and sulfur.

A sintered layer of a CaO-10 per cent $CaF_2$ mixture on the bottom of a magnesia crucible was found to give significantly greater desulfurization than was obtained in magnesia crucibles alone.

In a series of alloys refined in the partially lined magnesia crucibles, final sulfur concentrations of 0.004 to 0.005 per cent were obtained consistently. The final sulfur content was independent of the initial sulfur (in the range 0.013 to 0.020 per cent) and carbon content (in the range 0.40 to 0.60 per cent).

Ultimate tensile strengths as high as 317,000 psi were obtained in the desulfurized, vacuum melted metal. All test bars from fully desulfurized heats containing up to 0.55 per cent carbon exhibited ductile fractures. Values of reduction in area for desulfurized specimens of 4340 composition ranged from 37 to 40 per cent. Tensile strength of the same specimens was 250,000 to 260,000 psi; elongation was 10.0 to 12.5 per cent in the one in. gage length.

Type II sulfide inclusions were observed in castings containing more than 0.010 per cent sulfur, and globular silicates were present in all castings. Size type, and distribution of these inclusions are a major factor in determining the ductility at a given strength level.

The test castings were completely sound when examined by microradiography; hence microporosity is not believed to be a significant factor affecting mechanical properties of the specimens tested.

Statistical analysis of the tensile test data showed that ultimate tensile strength and reduction in area of the castings, heat treated to high strength levels, can be expressed linearly in terms of carbon and sulfur contents within the composition limits of this investigation.

The per cent reduction in area of the material investigated can be estimated by the equation:

$$\%\,RA_{(est)} = [-1.37\,(\%\,C) - 13.56\,(\%\,S) + 0.99] \times 10^2$$

This indicates that a linear decrease in the per cent reduction in area results from increases in either carbon or sulfur concentrations.

The ultimate tensile strength of these steels can be estimated as a linear function of the carbon and sulfur concentrations:

$$UTS_{(est)} = [2.73\,(\%\,C) + 10.76\,(\%\,S) + 1.44] \times 10^5$$

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# DIE CASTING TRIM DIE PROCEDURE

by Duncan Hannah

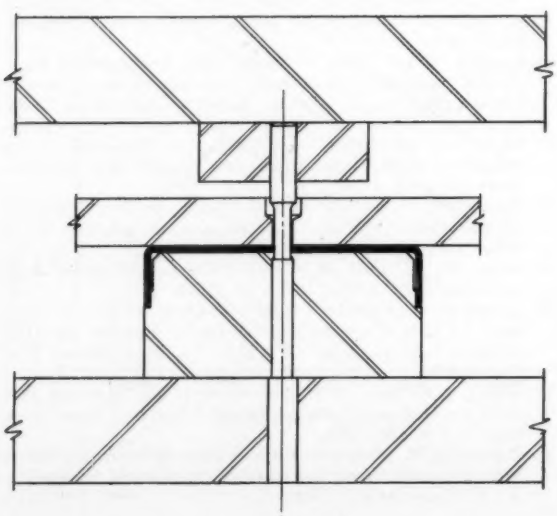

## ABSTRACT

**Trim dies for zinc, aluminum and magnesium die castings perform a basic function — that of trimming from the castings all flash in an economical manner. This trimming operation can be performed in various manners, with various types of equipment, depending on the specific problem at hand. For best results the trim die should be designed at the time the die casting die is designed, and any changes made in the die casting die design should be incorporated into the trim die design at the same time. This will make the trim die a completely functional tool giving as satisfactory a performance as the die casting die itself.**


## FLASH TRIMMING

Trim dies for trimming zinc, aluminum and magnesium die castings have a basic function. These dies must trim the casting of all flash in the most economical and expedient manner.

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The trimming of the flash from the casting may necessitate pierce punches (Fig. 1) to clean flash from holes, cam pierce operations (Fig. 2) to clean flash which is not accessible in the cast position and occasionally, secondary operations such as drilling, tapping, broaching, or any other secondary operation which may economically be added to the trimming die. These methods will be enlarged upon later in the article.

## BASIC TRIM DIE TYPES

### *Push Through Trim Die*

The first, and most common type, is the push through trim die (Fig. 3). In this die the part is loaded into the cutting steels (A) and pushed through the steels by the punch (B). The part then falls through the die shoe and press bed, into a bin or onto a conveyer, or the die set is mounted on risers (C), which permits removal of the casting from beneath the die set. This type of trim die is the most inexpensive to build, is fast in operation and has a low maintenance cost.

It has the disadvantages of nicking or marking a part when the casting is being loaded into the trim steels. Also, in the case of cylindrical parts, the parts may tend to roll in the die while being pushed through because varying gate, overflow and flash thicknesses tend to give an uneven trim force.

Several methods are used to prevent the marking or nicking of castings in the push through trim die.

**Fig. 1 — Trimming flash with pierce punches to clean castings holes.**



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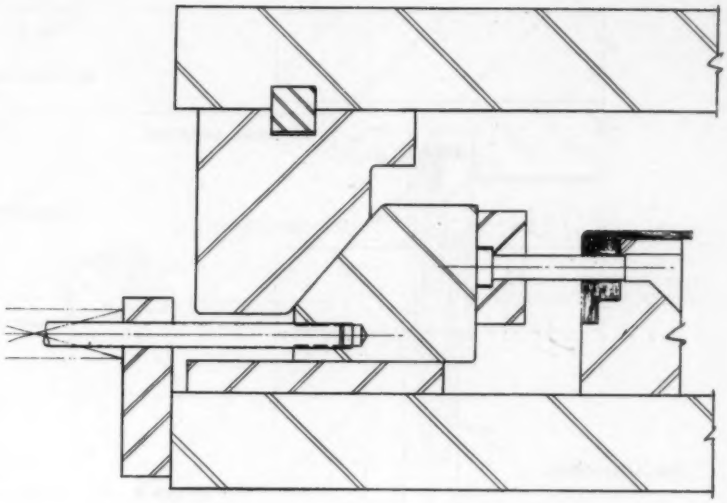


Fig. 2 — Cam pierce operation to clean flash not accessible in the cast position.

Figure 4 shows locators which are used to position the casting before the casting drops into the trimming steels. These locators insure that the part cannot be loaded incorrectly. When locators are used in a push through trim die, the locators are left in the trim steels at the end of the trimming cycle. This can be overcome by the use of ejectors mounted under the locator holes (Fig. 5) or the locators can be pierced through (Fig. 6). Figure 6 shows a scrap or flash stripper mounted on a push through trim die.

A scrap stripper is necessary when a casting is flashed completely around the entire parting line. A stripper will push the flash off the punch in the same manner as a sheet metal blanking or piercing die. If a stripper is not used, scrap cutters may be used (Fig. 7). A scrap cutter will cut the flash in two or more areas to allow its easy removal.

Another method of loading castings into a push through trim die is by the use of a loading tray. Figure 8 shows a casting in place in a loading tray after the press has closed. The loading tray is suspended from the upper shoe and therefore has a

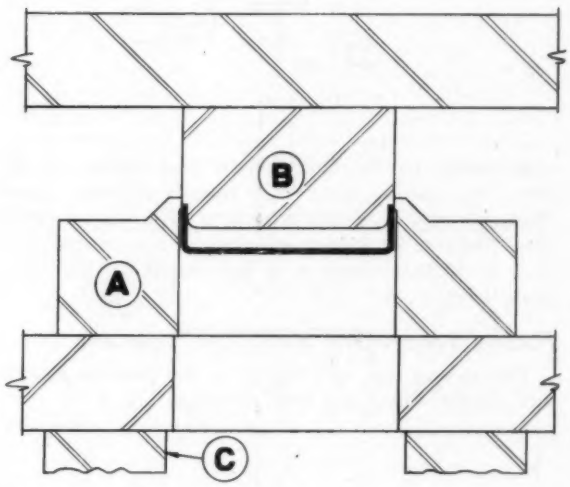



Fig. 3 — Push through trim die.

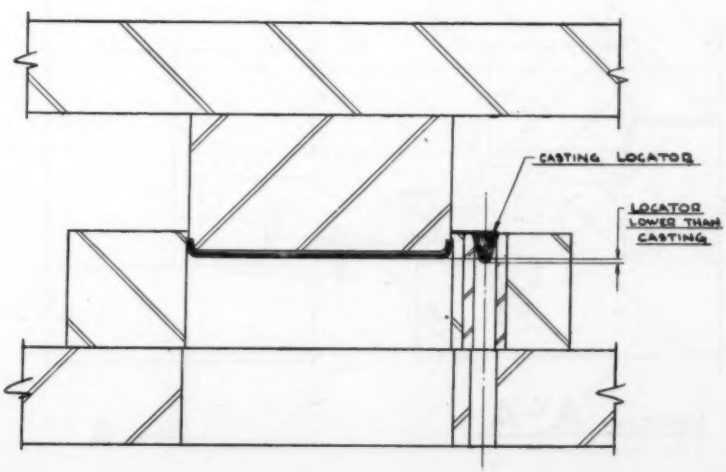



Fig. 4 — Locators used to position casting in push through die.

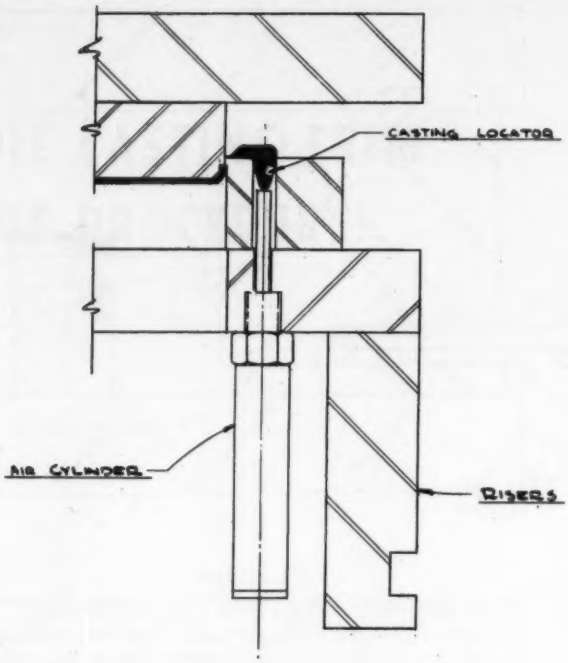



Fig. 5 — Ejectors mounted under locator holes.

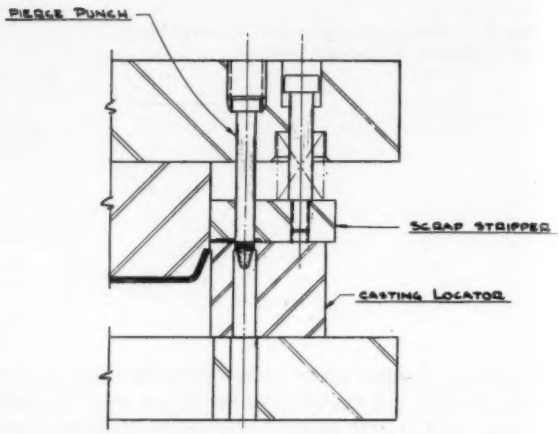



Fig. 6 — Scrap or flash stripper mounted on push through die.

relationship to the shut height and stroke of the press. A loading tray will usually operate more efficiently in a hydraulic trimming press than a crank press, because in a high speed press the casting may tend to bounce when it is positioned in the trimming steels.

### *Positive Locator and Knock-Out Trim Die*

The second type of trim die is the positive locator and positive knock-out type, shown in Fig. 9.

In this type of die construction, the part is loaded onto a locator (A) and the trim steels (B) trim the outside edge of the casting. On the upstroke of the press, the part is ejected by the stripper pad (C) when the knockout pins (D) contact a stationary bar in the press itself. In this type of trim die the part must be caught by the operator, used on an inclinable press which allows the casting to fall out the rear of the press, or a tray or chute may be used to catch the part.

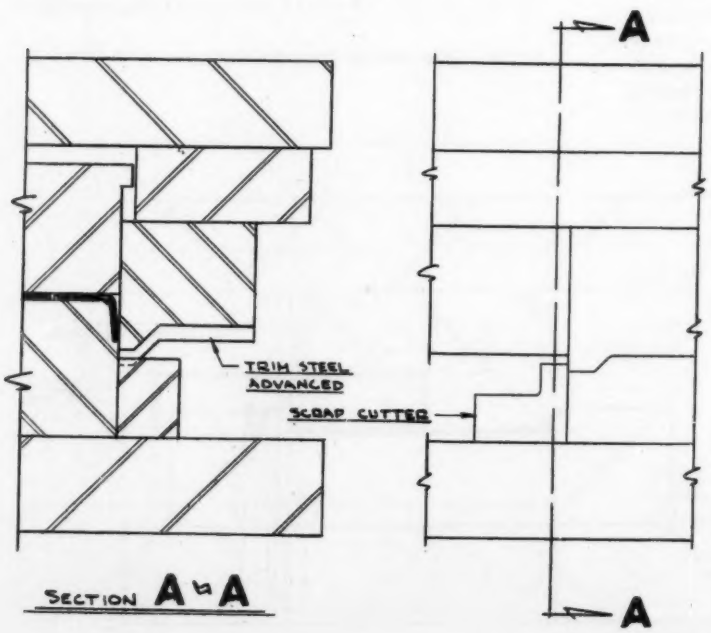



Fig. 7 — Scrap cutter on a push through die.

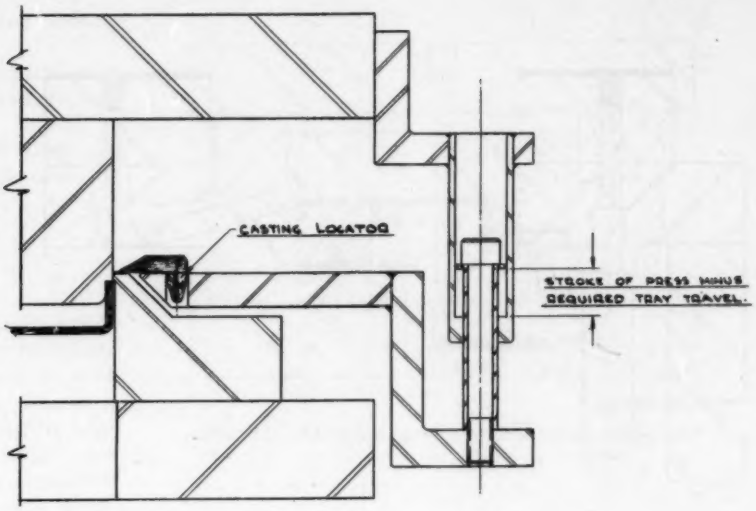



Fig. 8 — Casting in place on push through die after press has closed.

The tray or chute must enter into the die on the upstroke of the press, and must retract before the trim die closes. It must therefore be actuated by a pinion and rack which is tied into the press stroke, or a cylinder may be used which is tied into the stroke of the press by switches or valves.

### *Positive Locator and Manual Knock-Out Trim Die*

The third type of trimming die is basically the same as the second type. The difference is that instead of a positive knock-out, springs are mounted above the stripper or pressure pad (Fig. 10). This type of die leaves the casting sitting on the locator (A) from where the operator must pick it off. Air cylinders may be used to lift the casting off the locator, or this same operation may be performed manually (Fig. 11).

When holes are to be cleaned of flash, pierce punches are used, as shown in Fig. 12. A few methods of retaining the punches are shown on this layout. Type (A) is the common ball seated punch which is removable from the face of the die while the die is in the trimming press. This type of punch is readily changed with little downtime of the die. The next are headed punches which can be retained in any of the methods shown, (B) shows a standard retainer.

The third drawing (C) shows the same type of retainer with a set screw behind the punch. This

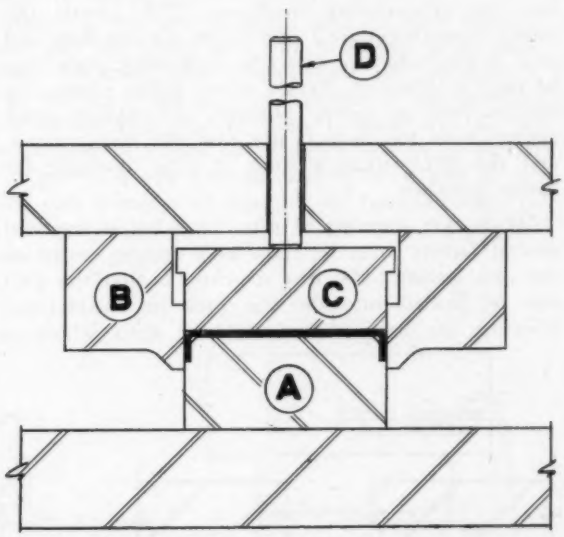



Fig. 9 — Positive locator and positive knock-out trim die.

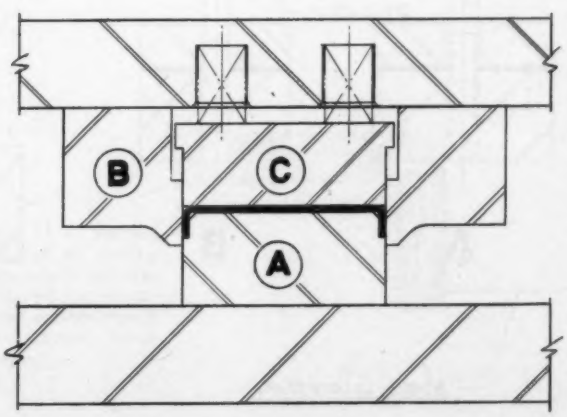



Fig. 10 — Positive locator and manual knock-out trim die.

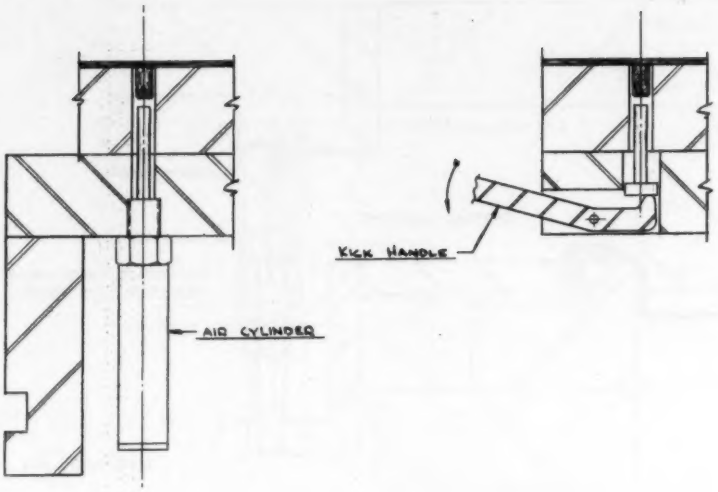



Fig. 11 — Air cylinder and manual lift on positive locator and manual knock-out die.

allows the punch to be removed from the die without any dismantling necessary. The fourth (D) shows a punch retained directly in the die shoe and held in place with a set screw. A backing plate may be used in place of the set screw. Many casters use ejector pins as pierce punches. All casters stock ejector pins for normal die cast die maintenance, and this gives them a stock of pins available for pierce punches.

When cam piercing or cam trimming is required several factors must be considered in the design of the cam station. The first requisite is that the part can be loaded into the die with no interference between the cam station and the flash which is inherent in every die casting. The slide must have enough travel to allow for the removal of the scrap after trimming, and the cam slides should be protected against small particles of flash which could jam and break the cam itself. In place of a cam driver, an air cylinder may be used to actuate the slide. This is useful when a long stroke is required.

Air cylinders are also used to perform drilling, tapping or broaching operations on the casting. A dwell in the stroke of the press is useful if the above operations are to be used. A good functional trim die has several requisites which it must possess:

1. It must locate the casting quickly and easily and give repetitive location.

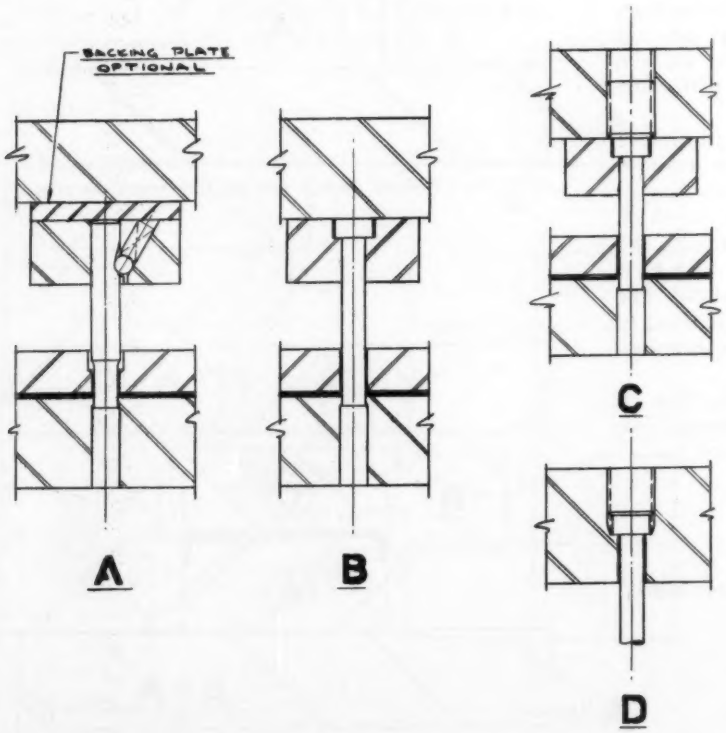



Fig. 12 — Pierce Punches used with die in Fig. 10. Type A — common ball seated punch. B, C and D are retaining methods for punches — B, a standard retainer; C, same type as B, only with a set screw behind the punch; D, punch retained directly in the die shoe and held in place with a set screw.

2. It must trim all flash, gates, overflows and runners cleanly and evenly.
3. It must strip the casting from the trimming steels without marking or deforming the part in any way.
4. The castings must be easily removable from the die after trimming, and the die itself must be easily maintained.

*Spring Loaded Pad Die*

Figure 13 shows a cross-section of a typical spring loaded pad die. The locator (A) has two functions. It must back up all trimming edges, and it must locate the casting. You will note that the locator is cleared away in all nonessential areas. A casting edge may be trimmed without support for a small area but a burr will result. This condition is sometimes necessary due to the design of the casting itself. Locators may be made of any flame hardening steel, or a steel prehardened to Rockwell C of 30-32. If severe service is required on the die O-1 or W-10 may be used.

The trimming steels may be O-1 hardened to Rc of 58-60 W-10 steel. Air hardening steel of the A-3 type or soft steels may be used having a welded trimming edge of tool steel or stellite (Fig. 14). On this type of trimming steels an edge of about $\frac{3}{16}$-in. square is recommended.

The pressure pad may be mild steel, flame hardening steel, polished or even resins and epoxys. Nylon buttons are sometimes used when a part may be nicked or scratched in trimming. The buttons should extend out of the pad no more than $\frac{1}{16}$-in.

For a die cast part which contacts a painted or enameled surface, the parting line of the die casting die should be stepped about $\frac{1}{32}$-in. (Fig. 15). The trim die must actually shave this edge for a distance of from 0.010 to 0.030 in.

To insure a clean trim edge or biting edge, a caster may put a flash or trim edge around the entire casting. This is a deliberate flash, cut into the die casting die itself approximately 0.010 in. deep and 0.250 in. wide. If a die has a light flash of 0.007 in. or less, the flash will tend to roll or fold leaving a burr on the casting.

## TRIM DIE BASIC FEATURES

A commercial or functional castings, which would include most aluminum and magnesium parts, will stay between 0.005 and 0.010 in. away from the casting itself. Certain basic features should be followed in designing and building a trim die for die castings.

On the design layout, the entire parting line should be shown along with the trim conditions that occur in the die cast die. This layout should also include the gates, runners, overflows and ejector pins. Screws and dowels should also be shown. This will prevent interference possibly between the above mentioned items. If a locator is cleared for bosses or studs, the clearance holes should extend through the locator and die shoe. This will eliminate any blind holes which would pack with dirt and flash and eventually keep the casting from registering properly.

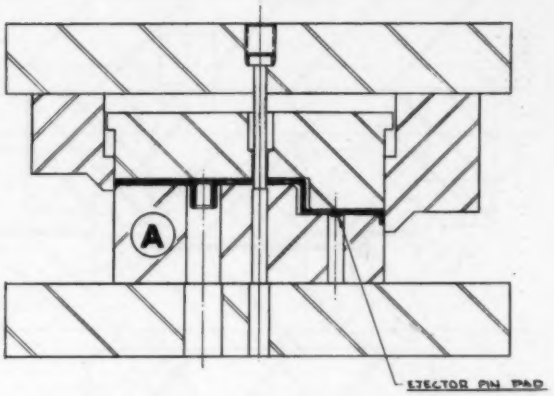



Fig. 13 — Cross-section of a typical spring loaded pad die.

Steel die sets are preferred and on push through trim dies the lower shoe should always be of steel. Some hydraulic trim presses have no top bolster. The die set is mounted on the end of the cylinder rod. When this type of press is used, the die set should not be a back post style. This type of die set will put a severe load on the cylinder rod if the pins and bushings of the die set become sticky or worn and scored. In this press use a center post set or a three or four post set.

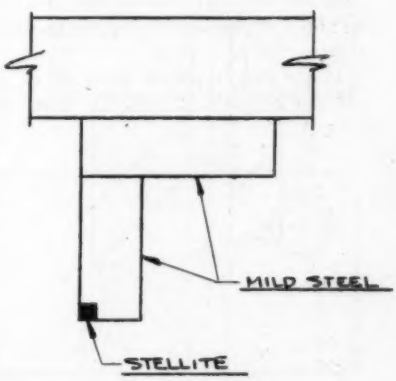



Fig. 14—Air hardened or soft steel with a welded trimming edge of stellite.

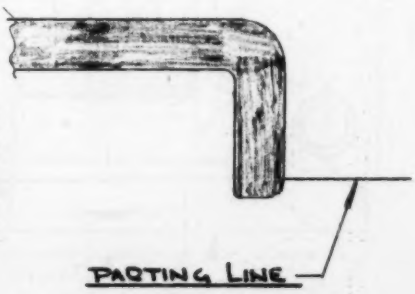



Fig. 15 — Parting line of die casting die which contacts painted or enameled surface should be stepped about $\frac{1}{32}$-in.

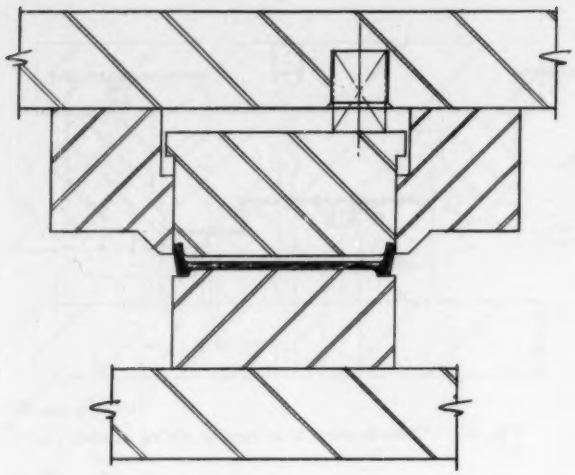


Fig. 16 — Part being trimmed in the reverse or upside down position.

Any type of trimming press will trim die castings, but most casters prefer the hydraulic presses because of the greater sheet height and adjustment, and because of more flexability with any secondary operations which may be included in the trim die. For long run jobs, crank presses are slightly faster, can be inclined in many instances, and they have no hydraulic maintenance problem.

## REVERSE TRIM

One trim not mentioned so far is the reverse trim. Figure 16 shows a part being trimmed in the reverse or upside down position. The casting is loaded into the die face down, and the trimming steels pass by the castings in the opposite direction of a normal trim.

With this type of trim it is possible to send a casting directly from the trim die to the plating department with no buffing or polishing. The casting itself must be rigid enough to support the trimming operation without buckling or deforming, and the locator of the die itself must be kept free of flash to prevent marking of the casting, as the casting is loaded face down on the locator.

The trim edge of the cutting steels should be smooth, and the finish marks (grinding or polishing) should be in the direction of trim to keep the steels from picking up metal from the casting and subsequently galling.

## SAFETY CONSIDERATIONS

Safety must be considered at all times when designing, building and running a trim die. Care should be taken to insure that the press operator at no time puts his hands directly over or under the cutting edges of the die. In many cases the castings may be loaded into the die with tongs. If space and construction permits, a loading handle may be cast in the die casting die (Fig. 17). This handle is used to load the part into the die and also to remove the scrap or flash.

The common types of safety features may be used such as sweep arms, bracelets or electric eyes; or the die may have a screen guard on its face with an opening which is just large enough to allow loading of the casting in the trim die.

As die castings become larger, and therefore heavier, methods are necessary for handling casting weighing 10 lb and over. The part must be automatically unloaded or ejected from the die casting die. It must then be moved to the trimming press. At this time most casters are loading the casting into the trim die manually and unloading the trimmed casting by means of the previously mentioned methods of pinion or cylinder operated chutes or slides.

On a casting with holes and flash which must be trimmed from various directions a trim machine may be built to do the entire operation in setting of the casting. Air and hydraulic cylinders are used to drive the punches and trim steels. The tonnage

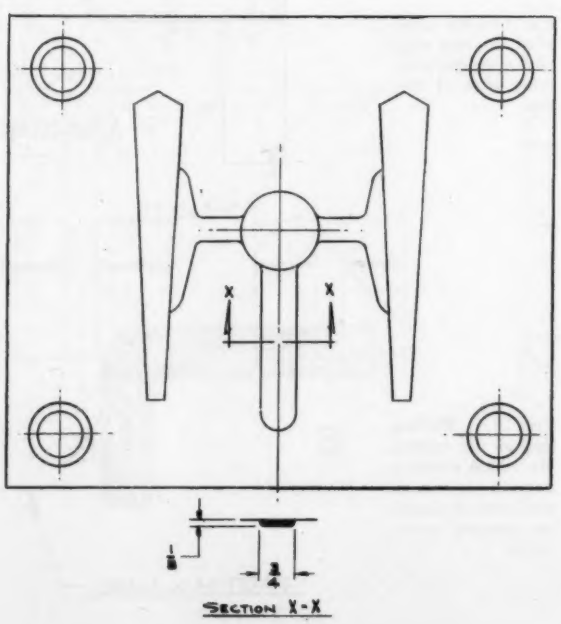



Fig. 17 — Loading handle cast in the die casting die.

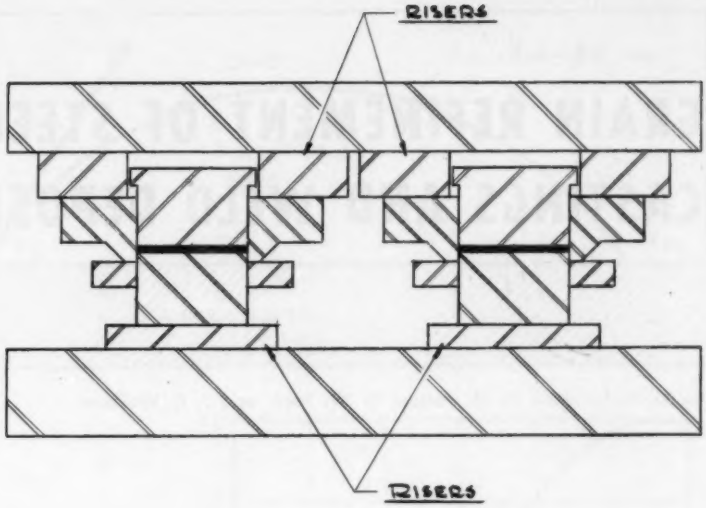



Fig. 18 — In multiple trim dies there should be cutting steels and locators or punches mounted on risers.

required to trim a die cast part should be treated in much the same manner as a sheet metal die. One method used is the length of cut times the thickness of gate and flash times 10. The answer will be a trimming pressure in tons.

The commonly used factor is 30 lb of spring pressure for every one in. of cutting edge, i.e., a casting with a periphery of 10 in. would require a spring pressure of 300 lb to release the casting from the trim steels. Any multiple trim die, i.e., a die trimming more than one casting per stroke of the press, should have the cutting steels as well as the locators or punches mounted on risers (Fig. 18).

## DIE VARIATIONS

A die caster may vary his gating, venting and overflows for a period after he has received his die cast die and trim die in order to produce sound platable castings, or sound dense castings in the case of functional parts. These changes in the die casting die may vary the shrinkage between the castings, which results in a mismatch in the trim die. Risers under the trim steels and locators or punches will allow these sections to be moved with a minimum of tool room time.

If trouble is anticipated in a section of a trim die, such as an engineering change or possible draft revisions on the die casting die are contemplated, double dowel holes may be installed in a trimming section which will allow the section to be repositioned. Another method more commonly used is shown in Fig. 19. Soft plugs are installed in the riser or the shoe. If a section must be moved, new plugs are installed and the section is redoweled.

A third method is shown in Fig. 20. The cutting steels are pocketed in the riser or the shoe and no dowel holes are required. Shims are used to move the die sections if repositioning is required. This method is more expensive, but it allows rapid resetting of trim die sections. At all times, punches, sections and replaceable items should be installed for easy replacement without dismantling of the die.

## CONCLUSION

The final conclusion is that the trim die should be designed at the time the die casting die is designed. Any changes made in the die casting die design should be incorporated into the trim die design at the same time. This rule applies also to any engineering part print changes.

The trimming die should be treated as a functional tool which must give as satisfactory a performance as the die casting die itself.

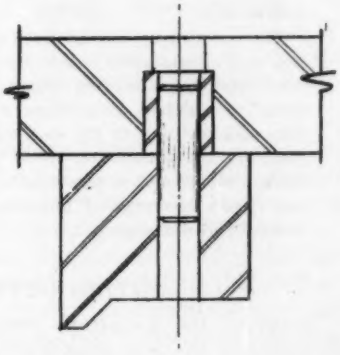


Fig. 19—Soft plugs installed in riser or shoe of die.

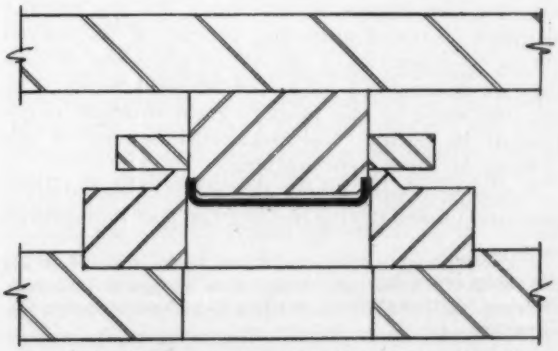


Fig. 20 — Cutting steels pocketed in riser or shoe. No dowel holes are required.

# GRAIN REFINEMENT OF STEEL CASTINGS AND WELD DEPOSITS

by G. K. Turnbull, D. M. Patton, G. W. Form and J. F. Wallace

## ABSTRACT

The refinement of the as-cast grain size of small A.I.S.I. C1018 steel ingots and the as-deposited dendritic arm or Widmanstatten ferrite structure spacing of steel welds containing approximately 0.14 per cent carbon was investigated. Additions were made to the melt before pouring the ingots or to the center of a tubular electrode utilized for depositing the weld metal by the submerged arc welding process. These additions were selected as sources of foreign nuclei or for grain growth restriction. The metallographic structure and mechanical properties of the ingots and deposited welds were determined.

The as-cast grain size of the ingots and the dendritic arm or Widmanstatten structure spaces of the deposited steel welds were markedly refined by selected additions, including titanium, columbium and titanium carbide. The strength of both the cast metal after heat treatment and the as-deposited welds was increased by this refinement. However, the ductility of the cast metal and impact resistance of the deposited welds were reduced by the treatment.

## INTRODUCTION

Many of the important properties of metals are determined by their microstructure. A control over the structure is, therefore, an important means in obtaining a satisfactory performance of a given material. One microstructural factor that has recently acquired increased attention, because of its marked effects on properties, is the as-cast grain size of castings or as-deposited grain size of weld beads. The size and shape of this primary grain structure is important in castings and weldments because of the absence of subsequent working.

It is generally thought[1-5] that a fine equiaxed grain structure formed during freezing can lead to improved mechanical properties, more uniform response in heat treatment, reduced anisotropy and better casting and welding behavior compared to large columnar grains in the casting or deposited weld metal. For these reasons, the possibility of refining the as-cast structure of metals has been extensively studied. Considerable success has been achieved with aluminum,[6-12] magnesium,[13-16] copper[17-19] and eutectics.[20-25] Some articles have been published concerning the methods of achieving a fine, equiaxed, as-cast structure in steel.[26-29] The information on this subject is limited, possibly because of difficulties encountered when attempting to refine and evaluate the primary steel structure.

Some recent work with steel,[30-31] however, has indicated that the mechanical properties, such as tensile ductility, are superior when columnar crystals are obtained. It is indicated that the large columnar crystals that result from an essentially unidirectional solidification contain little microporosity or microshrinkage between the dendritic arms and crystals, because these have solidified under steep thermal gradients. These steep gradients promote good directional solidification and feeding. In addition, the columnar type of unidirectional solidification does not contain any appreciable macrosegregation.

According to this work,[30-31] the equiaxed grain structures contain more microporosity, chemical segregation and resulting lower ductility. The poor properties of the equiaxed grain, however, do not appear to be intrinsic for this mode of solidification, but are probably associated with the lower thermal gradient existing in these regions during solidification. It may well be that joint employment of a satisfactory grain refiner and higher thermal gradients would show equiaxed grains with improved properties.

This investigation was directed towards a systematic study of the possibilities of grain refining low carbon steel castings and weld deposits by means of appropriate additions. The basic principles on which such attempts must be based have been published in detail.[32-33] Therefore, only a summary will be given.

G. K. TURNBULL is Rsch. Asst., G. W. FORM is Asst. Prof. and J. F. WALLACE is Prof. Met. Engrg., Case Institute of Technology, Cleveland. D. M. PATTON is Welding Engr., Lincoln Electric Co., Cleveland.

(Thesis submitted by D. M. Patton for M.S. degree at Case Institute of Technology)

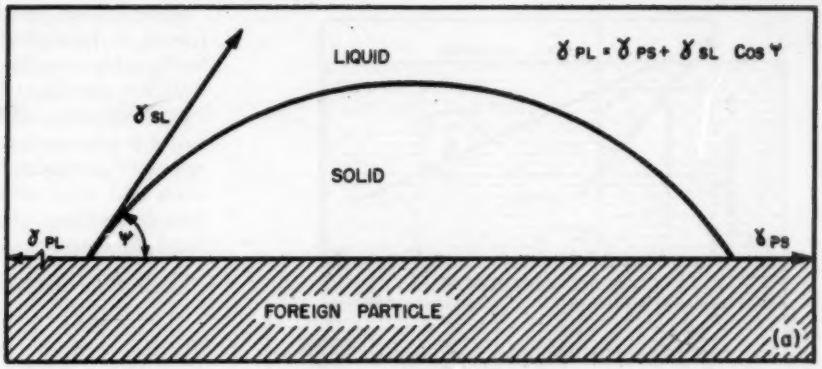
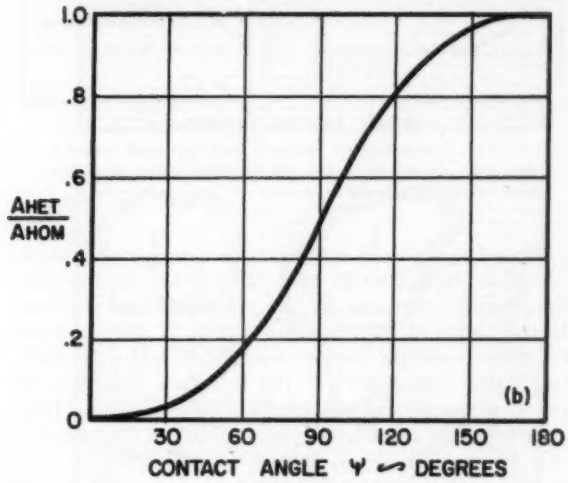



Fig. 1 — Contact angle effect on work of nucleation.

$$AHOM = \frac{C \gamma SL^2}{\Delta FV^3}$$

$$AHET = \frac{C \gamma SL^2}{\Delta FV^3} \quad \frac{(2 + \cos \psi)\ (1 - \cos \psi)^2}{4}$$




## BASIC ASPECTS OF GRAIN SIZE CONTROL

The achievement of a fine, equiaxed, as-cast structure requires nucleation to occur at a large number of uniformly distributed sites in the melt, and the restriction of excessive growth of any one crystal. The principal difficulty in achieving this dual objective originates from two sources – 1) a positive temperature gradient normally exists in the liquid toward the center of the casting and 2) the temperature at which growth of crystallites can proceed is higher than the nucleation temperature.

Since solidification starts at the mold wall in castings, the nuclei formed at this location in a pure metal tend to grow toward the center without interference, forming a columnar structure. This occurs, irrespective of how many more potential sites of easy nucleation may be contained in the melt, because at any point ahead of the advancing solid-liquid interface, the growth temperature is reached before the melt has cooled down to the nucleation temperature. While the availability of regions of easy nucleation is an essential feature of grain refinement, it is, in itself, insufficient to promote a fine, equiaxed structure.

Additional measures must be provided to make these potential nuclei become effective. This latter prerequisite is the grain growth restriction aspect of the grain refinement problem. Rapid nucleation from many sites can occur as a result of either a chill action, mechanical vibration or appropriate inoculation. In the present investigation grain growth restriction and inoculation were studied.

Inoculation is defined as the addition to the melt of a substance that is present in form of/or produces solid and stable particles above the solidification temperature range on which the parent melt can readily solidify. The mechanism by which inoculants can reduce the work of nucleation is best rationalized in terms of the interfacial energies acting on a flat-sided particle.

### *Heterogeneous Nucleation Theory*

The theory of heterogeneous nucleation proposes that nucleation with a small degree of undercooling (indicating a marked reduction in work of nucleation) is favored by a particle that produces a large particle-liquid surface free energy compared to the particle-solid interfacial free energy, as shown schematically in Fig. 1. It has been suggested[34] that good matching between the crystal structure of parent solid and inoculating particle is a necessary prerequisite of an effective inoculant. However, some exceptions to this rule are known.[35]

Grain growth restriction does not merely involve the slowing down of the advancing solid-liquid interface, but requires actual interruption of the growth of existing crystallites. The requirement of the latter prerequisite is best illustrated by the relative insensitivity of the length of the as-cast grains in pure metals to the changes in rate of heat removal, although the rate of crystal growth is varied within wide margins. During the growth of nuclei in alloy metals, however, a concentration gradient is generally set up in the liquid just ahead of the solidified portion, because freezing usually occurs faster than compositional equilibrium can be maintained leading to a local enrichment of the liquid with solute atoms.

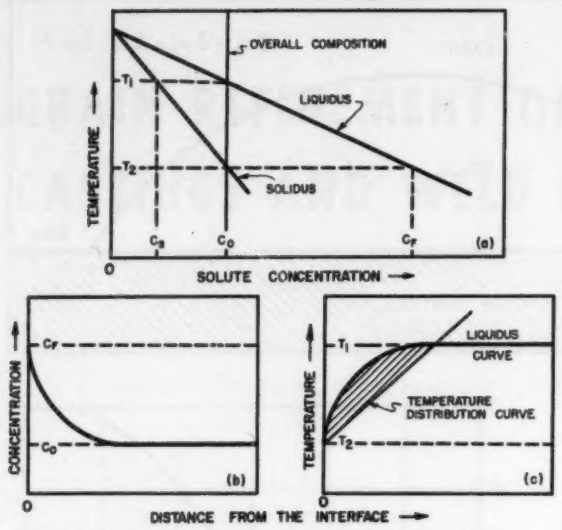



Fig. 2 — Concentration buildup and liquidus distribution next to interface for single phase alloy solidification (steady state growth).

Since the liquidus temperature varies with composition, it must change near the advancing interface in a manner dictated by the concentration gradient. If the slope of the liquidus curve at the interface is steeper than the temperature gradient in the melt, the liquid metal next to the interface is said to be constitutionally supercooled.[36] This condition is illustrated by the schematic curves in Fig. 2. In order to achieve appreciable constitutional supercooling, it is necessary that a solute addition be contained in the melt that produces a small distribution coefficient (k) or ratio of solute in the solid to solute in the liquid. When a sufficiently shallow thermal gradient is obtained, independent nucleation in the melt ahead of the solid-liquid interface can occur.

Thus, the addition of an appropriate solute to the melt to achieve appreciable constitutional supercooling may serve as an effective means of assisting in the independent formation of new grains in the melt by preventing the existing grains from further growth. If, by this means, the growth of crystallites is interrupted early enough, the process of repeated independent nucleation will lead to the desired fine equiaxed grain. Based on the above considerations, the two avenues of approach in the present investigation involve the addition to molten steel of potential inoculants and of solutes promising a strong ability for constitutional supercooling.

## MATERIALS AND PROCEDURE

The principal objective of this study was the development of feasible methods to grain refine steel castings and weld deposits by means of suitable additions. The investigation was undertaken in two phases – 1) relatively small melts of steel were treated with various inoculants, poured into ingot molds and tested to determine the effectiveness of these additions in grain refinement, and 2) grain refining additions were made to welding electrodes and their effect on the deposited weld metal was determined.

The initial portion of the investigation encompassed a theoretical evaluation of possible grain refining additions, based on crystallography, density and stability considerations. Based on these results, an extended series of closely controlled tests with 20 lb castings was initiated to evaluate the substances which appeared favorable. The most promising of these additions and some other materials were then introduced into a welding electrode, and deposited in a weld bead on a steel plate by arc welding techniques.

### *Selection of Additions*

The additions were selected on the basis of the principal aspects outlined in the introduction. Accordingly, a given addition was placed into either one of two categories: an inoculant or source of foreign nuclei, or a grain growth inhibitor. Some additions, however, were potential grain refiners for both reasons.

Potentially suitable inoculants were originally chosen on the basis of good lattice registry with the base metal. The solidification of low carbon steel involves the selection of grain refiners with close crystallographic matching with either austenite or delta iron. Since the delta phase constitutes the primary crystallites, however, the attention was focused on the nucleation of this structure. The effect of the initial delta particles on the resulting solidified structure is shown schematically in Fig. 3.

It was assumed that the development of many fine delta particles in the liquid was desirable in itself, and would result in a finer as-cast grain size during the subsequent peritectic reaction. Among the crystallographically favorable substances, only those which were sufficiently stable in the parent liquid were feasible. This excluded additions which melt at a temperature substantially lower than steel or which dissolve easily in molten steel.

Although most inoculants are soluble to some extent in liquid steel, the time dependency of dissolution will permit an interval for effective inoculation. This requires, of course, a close control over the time elapsing between the introduction of the addition and actual solidification. This latter problem is, of course, less serious in weldments. While the dissolution aspects limit the effectiveness of an addition when large castings or high pouring temperatures are involved, it has the advantage that it provides for a clean surface which otherwise cannot readily be attained.

The chemical reaction of a given inoculant with steel could be assessed to some extent from known phase relationships and known thermodynamic data. It was not possible, however, from the limited information available, to determine the reaction of such a particle with other solutes present in the melt. Some restriction on the selection of suitable inoculants was also imposed by density consideration. A large difference in density between addition and steel may lead to marked segregation by gravity, particularly when the particle size is coarse.

### *Grain Growth Inhibitors*

The grain growth inhibitors were chosen on the basis of their potency to attain a large degree of constitutional supercooling per atomic per cent of

addition. A small value of the distribution coefficient (k), as read from the respective equilibrium diagram, and a large freezing range was considered favorable. Unfortunately, many of the iron-solute binary diagrams have not yet been completely established for low percentages of solute close to the melting range, thus preventing an accurate determination of the distribution coefficient.

Additions for the purpose of grain growth inhibition are not subject to many of the limitations imposed by an inoculating addition. The melting point of the solute becomes unimportant. The time interval elapsing between introduction of the addition and actual solidification is considerably less critical for grain growth inhibitors than for inoculants, although long holding times increase the possibility of oxidation of the solute and tend to reduce the effective concentration of the additions.

### *Cast Ingots*

Each melt was prepared individually by melting 20 lb of steel in a high frequency induction furnace. A charge consisting of A.I.S.I. C1018 steel bar stock was melted down before making any alloying or inoculating additions. Deoxidation was accomplished in the furnace with a 0.25 per cent addition of aluminum. More aluminum than usually employed in commercial practice was utilized to minimize oxidation of the inoculants. A preliminary series of ingots was cast to determine the influence of pouring temperature and cooling rate on the grain size.

Additions were introduced into the remaining series of ingots by plunging the inoculant attached to a graphite or steel plunger below the surface of the melt and stirring. The additions employed, identified as either inoculants or grain growth inhibitors, are listed in Table 1. Sulfur additions were made to some melts to aid in the identification of the as-cast grain. The slag was then removed from the melt and the casting poured. The pouring temperature was maintained at 2975 F (1635 C) ± 25 F.

After treatment, the molten steel was poured into a cast iron chill mold with an inside diameter of 5 in. and a wall thickness of $\frac{5}{16}$-in. seated on a 1½-in. thick steel chill block. This arrangement permitted a fairly rapid directional solidification from the base and sides. A 3 in. square by ¼-in. thick steel plate was placed upon the bottom block within the mold cavity to prevent the casting from welding to the chill block. The castings were permitted to cool in the mold and were stripped by breaking away the cast iron cylinder.

Metallographic examinations were conducted on longitudinal cross-sections of each casting. Sulfur prints of this section gave an indication of the degree of segregation encountered. Each ingot was subsequently re-ground and macroetched with either cupric ammonium chloride (Humfrey's reagent) or a 50 per cent aqueous solution of nitric acid. Specimens were cut from selected castings for microscopic examination.

Tensile tests were conducted on a series of castings containing additions which proved to be effective grain refiners. Individually cast bars 2 in. in diameter

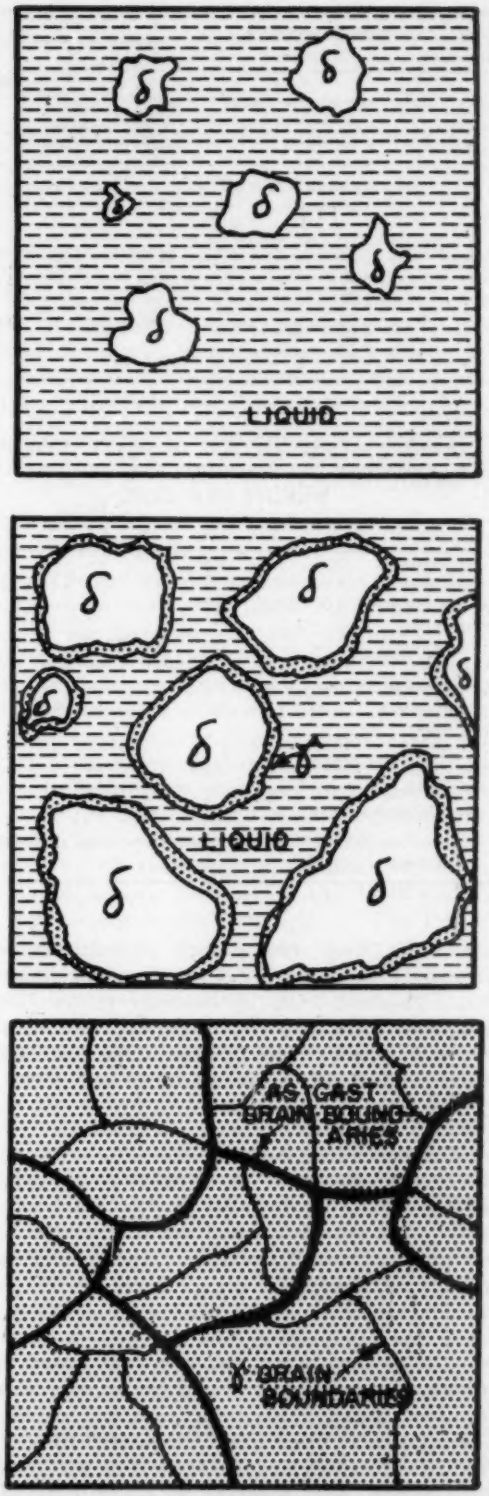



**Fig. 3 — Solidification sequence for a 1018 steel (schematic). *Top* — T > T peritectic (L + δ). *Center* — T = T peritectic (L + δ + γ). *Bottom* — T < T peritectic (γ).**

TABLE 1 — ANALYSIS AND AMOUNT OF ADDITION ADDED TO CAST INGOTS

| Addition | Amount Added, % by wt. | Elements, % by wt. C | Si | Al | Zr | Cb | Ti | B | Ni | V | Co | Fe |
|---|---|---|---|---|---|---|---|---|---|---|---|---|
| **Grain Growth Inhibitor** | | | | | | | | | | | | |
| FeCb | 0.2, 0.4, 0.9, 1.0 Cb | 0.05 | 2.2 | — | — | 57.0 | — | — | — | — | — | bal. |
| FeTi | 0.05, 0.50 Ti | 0.06 | 22.0 | 1.0 | — | — | 29.0 | — | — | — | — | bal. |
| FeB | 0.10, 0.50 B | 0.06 | 1.13 | 1.04 | — | — | — | 16.76 | — | — | — | bal. |
| Zr | 0.50 | — | — | — | 100 | — | — | — | — | — | — | — |
| TiC | 0.45 and 0.20, 0.30, 0.45 with Ti | 20 | — | — | — | — | 78 | — | — | — | — | — |
| **Nuclei Former** | | | | | | | | | | | | |
| Fe,V,B | 0.50 | 0.085 | 1.03 | 2.07 | — | — | — | 6.09 | — | 40.0 | — | bal. |
| FeV | 0.50V | 0.20 | 1.44 | — | — | — | — | — | — | — | — | bal. |
| W | 1.00 | — | — | — | — | — | — | — | — | — | — | — |
| AlCo | 1.00 | — | — | 31.0 | — | — | — | — | — | — | 69 | — |
| AlNi | 1.00 | — | — | 24.70 | — | — | — | — | 75.30 | — | — | — |
| ZrC | 0.30 with Ti | 12 | — | — | 88 | — | — | — | — | — | — | — |

TABLE 2 — COMPOSITION OF MATERIALS AND ADDITIONS TO WELDING ELECTRODES IN WEIGHT PER CENT*

| Material | C | Mn | Si | Al | Zr | Ti | Cb | Fe |
|---|---|---|---|---|---|---|---|---|
| Iron powder | 0.16 | — | 0.35 | — | — | — | — | 96.5 |
| Ferrosilicon | 0.10 | — | 50.0 | — | — | — | — | bal. |
| Ferromanganese | 7.0 | 80.0 | 1.0 | — | — | — | — | 13.0 |
| Aluminum-zirconium | — | — | 2.5 | 38.7 | 55.0 | — | — | — |
| Aluminum | — | — | — | 99.0 | — | — | — | — |
| Ferrocolumbium | 0.05 | — | 2.2 | — | — | — | 57.0 | bal. |
| Ferrotitanium | 0.06 | — | 22.0 | 1.0 | — | 29.0 | — | bal. |
| Titanium Carbide | 20.0 | — | — | — | — | 78.0 | — | — |
| Base Plate | 0.17 | 0.51 | 0.08 | — | — | — | — | bal. |
| Tubular Strip for Electrodes | 0.10 | 0.24 | 0.09 | — | — | — | — | bal. |

*Sulfur less than 0.025% and phosphorous less than 0.015% in all cases.

and 5 in. long were poured in sand molds after treatment with additions of titanium and titanium carbide. Standard 0.505 in. diameter tensile bars were machined 0.020 in. oversize from each casting, austenitized at 1650 F (899 C), water quenched, drawn at 350 F for one hr, finish machined and tested.

### *Weld Deposits*

The weld beads made for the purpose of obtaining the tensile specimens were deposited on the surface of a ½ x 2 x 18 in. long, A.I.S.I. C1015 hot rolled steel plate. The actual composition of the steel plate is shown in Table 2 along with the other data. The surface of the plate was prepared for welding by milling two slots in the plate approximately 3/32-in. deep and 1½-in. long. The slots were located in the middle of the short dimension and started approximately 4½-in. from either end, so that they were centered about 6 in. from each end of the plate. The slots were made to get the desired penetration of the weld bead below the surface of the plate. This penetration was necessary in order that the gage length of the tensile specimen would consist of all weld metal.

The welds were deposited on the prepared plates with a tubular, automatic electrode. The tubular electrode was hot rolled A.I.S.I. C1015 steel strip formed into a tube with an outside diameter of ⅛-in. and 0.022 in. wall thickness. The tube was filled with the addition agents and iron powder filler. The weld was struck through a ¾-in. thick, 2 in. wide layer of a neutral $MnSiO_3$ flux, and the arc was maintained below this flux during the weld as is customary in the submerged arc process.

The weld was started approximately one in. from one end of the plate and deposited, without breaking the arc, in the center of the long dimension to about one in. from the other end. A constant potential, direct current, welding generator operated fully automatically at 32 volts, 600 amps and 25 in./min of travel was employed. Figure 4 illustrates the relative position of the various components during welding. The resulting weld bead was approximately ⅝-in. wide at the top, tapered to about 5/16-in. at the bottom, ½-in. deep and 16 in. long.

The Charpy V-notch impact specimens were obtained from a weld bead deposited in a V-groove that was ⅜-in. deep and had a 50 degree included angle between the edges of two plates. The tubular electrode ($MnSiO_3$ flux) base plate and power source used in depositing the weld bead were similar to those used in making the welds for tensile specimens. The weld bead was deposited under a 1 x 2 in. wide layer of flux in a continuous operation. The bead was approximately ¾-in. wide at the top and tapered down to ⅜-in. wide at the bottom by about 16 in. long. The welding machine was operated automatically at 37 volts, 800 amps and 15 in./min travel speed throughout the weld.

### *Metal and Alloy Additions*

Various metals and alloy additions were made to the iron powder filler material, and these electrodes were tested to determine the influence of additions on the as-deposited grain size and mechanical properties of the resultant weld deposit. The control electrode for this investigation contained the iron powder filler, 0.30 per cent by weight of silicon in the form of a ferrosilicon (50% Fe, 50% Si), and 0.60 per cent manganese in the form of a high carbon ferromanganese (7% C, 13% Fe, 80% Mn). All of the subsequent electrodes contained this amount of silicon and manganese plus various additives.

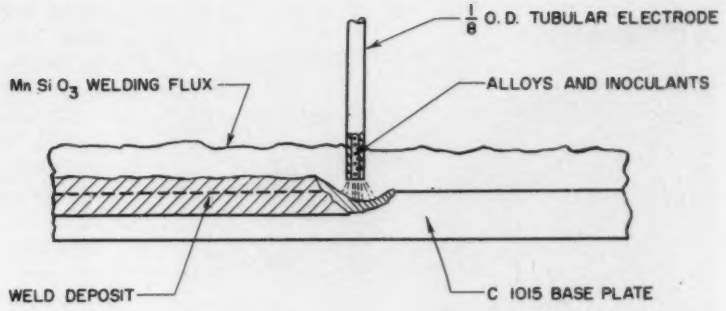



Fig. 4 — Submerged arc welding process employed for grain refining weld deposits.

The composition of the additions are shown in Table 2, and the overall composition of the welding electrodes resulting from these additions in Table 3. An aluminum-zirconium alloy (39% Al, 55% Zr) was added in amounts from 0.5 to 2.0 per cent. In order to differentiate the effects of aluminum and zirconium, only aluminum powder was added to some electrodes in amounts ranging from 0.77 to 1.50 per cent. The effect of columbium was investigated by the use of 0.88 to 2.74 per cent ferrocolumbium (57% Cb, 41% Fe) in the control electrode.

TABLE 3 — COMPOSITION OF TEST ELECTRODES RESULTING FROM ADDITIONS

| | Elements, % by wt. | | | | | | | |
|---|---|---|---|---|---|---|---|---|
| Additions, % | Mn | Si | Al | Zr | Cb | Ti | TiC | Fe |
| Control | 0.60 | 0.30 | — | — | — | — | — | bal. |
| 2.00 Al-Zr | 0.60 | 0.30 | 0.78 | 1.10 | — | — | — | bal. |
| 1.00 Al-Zr | 0.60 | 0.30 | 0.39 | 0.55 | — | — | — | bal. |
| 0.50 Al-Zr | 0.60 | 0.30 | 0.19 | 0.28 | — | — | — | bal. |
| 0.77 Al | 0.60 | 0.30 | 0.77 | — | — | — | — | bal. |
| 1.50 Al | 0.60 | 0.30 | 1.50 | — | — | — | — | bal. |
| 0.88 Fe-Cb | 0.60 | 0.30 | — | — | 0.50 | — | — | bal. |
| 1.82 Fe-Cb | 0.60 | 0.30 | — | — | 1.00 | — | — | bal. |
| 2.74 Fe-Cb | 0.60 | 0.30 | — | — | 1.50 | — | — | bal. |
| 1.72 Fe-Ti, 0.50 TiC | 0.60 | 0.34 | — | — | — | 0.50 | 0.50 | bal. |
| 3.45 Fe-Ti, 1.00 TiC | 0.60 | 0.78 | — | — | — | 1.00 | 1.00 | bal. |
| 3.00 Fe-Mn | 2.40 | 0.30 | — | — | — | — | — | bal. |
| 1.72 Fe-Ti, 0.50 TiC, 3.00 Fe-Mn | 2.40 | 0.34 | — | — | — | 0.50 | 0.50 | bal. |

It was desired to study the effects of equal parts of titanium and titanium carbide in the same electrode on the weld deposit. A ferrotitanium (29% Ti, 68% Fe) was added in amounts from 1.72 to 3.45 per cent. In conjunction with the ferrotitanium, 0.5 to 1.00 per cent commercially pure titanium carbide was used. On the second last electrode the manganese content of the control electrode was increased to a total of 2.40 per cent in the form of a ferromanganese (80% Mn, 20% Fe). Finally, to this latter high manganese electrode, 0.50 per cent titanium carbide and 1.72 per cent ferrotitanium (29% Ti, 69% Fe) were added. All powder additions were finer than 40 mesh.

Two 0.250 in. diameter tensile test specimens with one in. gage lengths were machined from the weld deposits in each plate. The small tensile specimen was chosen in preference to the larger sizes, because of the difficulty in obtaining weld metal for the entire test specimens in a single pass weld if larger tests were employed. Multiple pass welds were not desired, because of the influence of reheating on the as-deposited structure. Samples for determining the hardness of the weld were obtained. These samples were taken midway between the two tensile specimens. Strips ½-in. wide and transverse to the long axis of the weld were saw cut from the plate, polished and tested to obtain an average Rockwell B hardness for the weld metal.

### *Charpy V-Notch Test*

Several Charpy V-notch specimens were machined from each weld bead. The notch was placed at the bottom of the weld so that the fracture occurred entirely through weld metal. The energy required for fracture of the impact specimens was an average obtained from testing a minimum of six specimens at room temperature. All mechanical property data were obtained from as-deposited weld metal.

The line intercept method was used to determine the size of the arms on the dendritic grains. The planes of polish were selected parallel to the longitudinal axes of the dendritic grains. Welds are dendritic in nature because of the high thermal gradients existing during solidification. The line intercept method was developed by Snyder and Graff to determine the austenitic grain size.[37] However, it is effective in determining the relative size of the as-deposited microstructure of single pass welds. The system consists of averaging the number of grains intercepted by a 0.005 in. line at ten random locations in the sample.

These measurements were conducted at 1000×. Further reference to the size of the dendritic arms or Widmanstätten ferrite in this paper will mean the as-deposited size of these arms or the dendritic spacing as identified by the intercept method. It was difficult to distinguish the dendritic arms from the Widmanstätten ferrite in some of the welds. Photomicrographs of the polished and etched weld deposits were taken at 500×. These photomicrographs were located at approximately the center of the weld bead and transverse to the longitudinal dimension.

## RESULTS AND DISCUSSION

The identification of the as-cast grain size requires considerable study. This difficulty occurs because the iron solidifies first as delta iron that is converted to gamma at the peritectic temperature. Subsequent solidification of the remaining one quarter or less of the liquid occurs as gamma iron, as shown in Fig. 3. Then, the austenite undergoes a transformation to

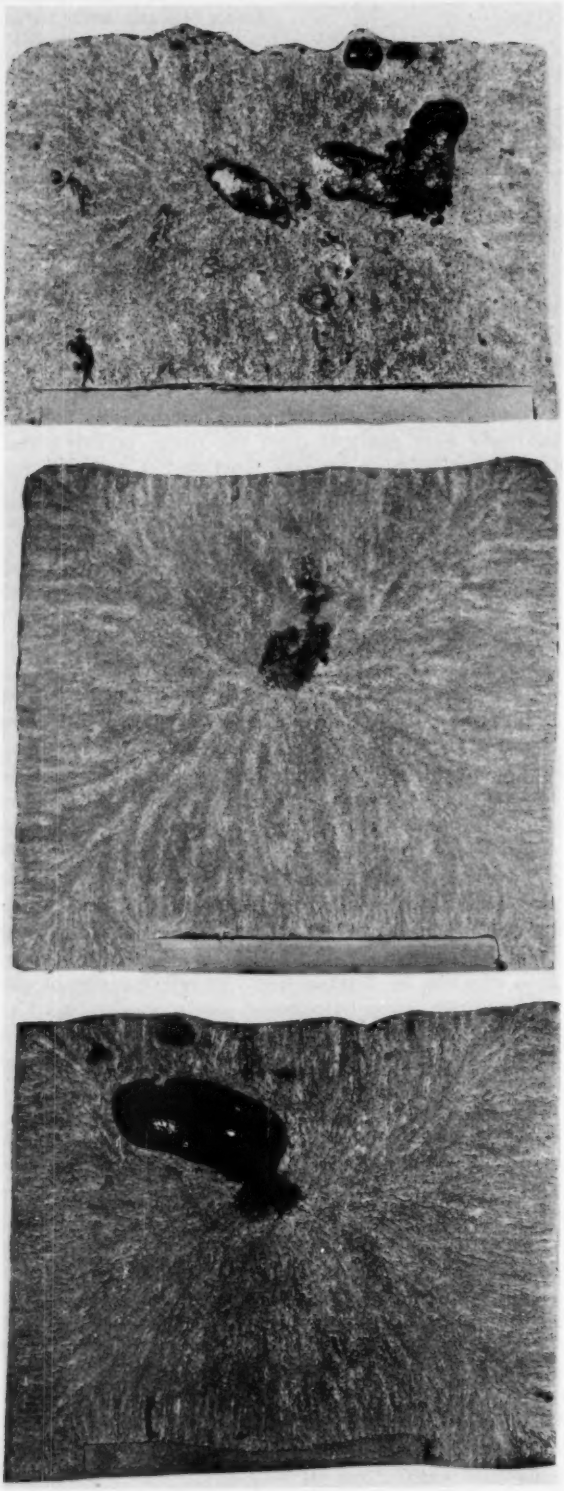


Fig. 5 — Small steel castings poured at various temperatures. *Top to bottom* — 2800 F, 2980 F and 3080 F.

alpha iron and cementite as it cools to room temperature. The grain size of the delta iron as it solidifies in these low carbon materials is, however, the important consideration.

The effective refinement of the as-cast grain, usually consists of altering the structure from columnar to equiaxed grains. The grain refinement of weldments is primarily accomplished by reducing the spacing of the dendrites of each grain. The actual grains in welds are columnar and many times (50 or more) as large as the dendrite size.[38] The last areas to solidify are generally higher in alloying elements and nonmetallics. Accordingly, appropriate etching can usually be employed to reveal this structure.

In some cases, however, positive identification of grain refinement in castings was not possible. The as-cast grain size of the ingots was classified as either columnar, partially columnar and partially equiaxed, or equiaxed when these grains could be identified. When identification was not possible, the grain structure was referred to as questionable. The more rapid solidification of the as-deposited weld metal produced a columnar dendritic pattern in all cases, and made it advisable to use only the intercept method of identification.

While it was feasible to observe the relative fineness of the welds, it was not always possible to distinguish the dendritic arms from the Widmanstätten ferrite. For this reason, the fineness of the weld structure was classified both on the basis of the dendritic arms and Widmanstätten ferrite. Since a relation may well exist between the fineness of these two structures, the use of both to determine relative fineness is believed to be justified.

### *Cast Ingots*

The preliminary series of tests indicated that no significant difference in the grain size of the ingots occurred with large variations in pouring temperature and cooling rates with the untreated or uninoculated metal. The pouring temperature was varied from 2800 to 3080 F (1538 to 1693 C), and the cooling rate was representative of a chill mold, a green sand mold and a preheated crucible that was furnace cooled after pouring. The similarity of structure with variations in the pouring temperature and cooling rate are shown by the photomacrographs of longitudinally sectioned and etched ingots in Figs. 5 and 6. It was concluded from these results that unintentional slight variations in pouring temperature and cooling rate would not affect the structure significantly.

The influence of the various additions on the structure of the 20 lb ingots of 1018 steel are shown in Table 4. This table also lists the ingots to which sulfur was added. Solute additions of columbium and titanium were effective, apparently because of their grain growth restriction effect. Small amounts of these additions exerted a grain refining influence, and these metals were effective in completely removing the columnar grains when 0.50 per cent by weight or more was added. It must be appreciated that nearly all the additions investigated are readily oxidizable.

Even though a small amount of these materials is sufficient to obtain marked grain refinement, consider-

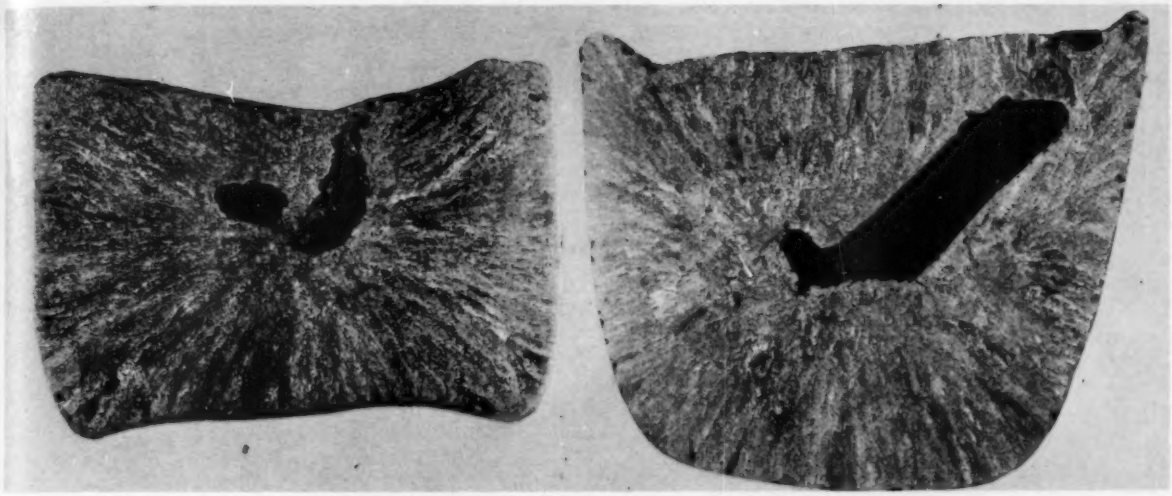


Fig. 6 — Macrostructure of small steel castings solidified at different rates. *Left* — cooled in green sand mold. *Right* — poured into refractory crucible and furnace cooled.

ably more may have to be added because much of the addition may be lost by oxidation. Additions of boron and zirconium did influence the macrostructural appearance to a considerable degree, but a positive identification of the as-cast grain size by microscopic examination was not possible in these ingots.

Titanium carbide, added as a nuclei former, was effective in reducing the grain size of the ingots. This addition in relatively small amounts reduced the length of the columnar zone sharply. Tungsten, AlCo and AlNi did not exert any noticeable grain refining effect. Combinations of effective nuclei formers with grain growth restriction agents also produced marked grain refinement, as shown by various combinations of titanium and titanium carbide or zirconium carbide.

A typical example of the combined action of an inoculant and a grain growth inhibiting agent is shown in Fig. 7a. The casting represented in this figure contained additions of 0.30 per cent titanium and 0.30 per cent zirconium carbide. A pronounced reduction in the length of the columnar structure compared to the control ingots in Fig. 5 can be observed.

### *Sulfur Prints*

Sulfur printing proved particularly successful with castings containing sulfur additions, when employed with some structures in which the as-cast grain size could not be identified by a macroscopic etching technique. The results were less conclusive when deliberate sulfur additions were not made and the sulfur content in the castings was low. Figure 7b shows a casting to which 0.50 per cent V-B was added for the purpose of inoculation. On the basis of the initial macroscopic examination, this casting was classified as having a questionable structure.

The sulfur print, however, brings out a transition from columnar to equiaxed solidification, and permits classification of this casting as 75 per cent equiaxed thereby indicating appreciable grain refinement.

TABLE 4 — GRAIN REFINING ADDITIONS INFLUENCE ON STRUCTURE OF INGOTS

| Addition | | | |
|---|---|---|---|
| Type | Amount, % by wt. | Sulfur Added | Macrostructure |
| **Growth Restriction** | | | |
| Cb as FeCb | 0.20 | No | Questionable |
| Cb as FeCb | 0.40 | Yes | 15% columnar, 85% equiaxed |
| Cb as FeCb | 0.90 | Yes | 100% equiaxed |
| Cb as FeCb | 1.00 | No | 100% equiaxed |
| Ti as FeTi | 0.10 | No | 35% columnar, 65% equiaxed |
| Ti as FeTi | 0.50 | No | 100% equiaxed |
| Ti as FeTi | 0.60 | No | 100% equiaxed |
| B as FeB | 0.10 | No | Questionable |
| B as FeB | 0.50 | No | Questionable |
| Zr metal | 0.50 | No | Questionable |
| **Nuclei Formers** | | | |
| TiC | 0.45 | No | 30% columnar, 70% equiaxed |
| V as FeV | 0.50 | No | 100% columnar |
| Fe, V, B | 0.50 | Yes | 25% columnar, 75% equiaxed |
| W | 1.00 | Yes | 100% columnar |
| AlCo | 1.00 | Yes | 100% columnar |
| AlNi | 1.00 | Yes | 100% columnar |
| **Growth Restriction and Nuclei Formers** | | | |
| Ti Metal + TiC | 0.20 each | No | 20% columnar, 80% equiaxed |
| Ti metal + TiC | 0.30 each | No | 100% equiaxed |
| Ti metal + TiC | 0.45 each | No | 100% equiaxed |
| Ti metal + ZrC | 0.30 each | No | 20% columnar, 80% equiaxed |

Since simultaneous additions of titanium and titanium carbide produced consistent grain refinement, a series of cylindrical bars was cast to determine the effect of these additions on the tensile properties. The results of these tests after quenching the tempering are:

| Additions, % | | Tensile Str., | Elong., | Red. in Area, |
|---|---|---|---|---|
| Ti | TiC | Ksi | % | % |
| 0 | 0 | 69.0 | 24.5 | 58.6 |
| 0.1 | 0.1 | 79.6 | 3.0 | 6.6 |
| 0.1 | 0.5 | 95.3 | 3.0 | 3.7 |
| 0.3 | 0.1 | 85.2 | 3.0 | 4.9 |
| 0.3 | 0.5 | 86.7 | 2.0 | 1.2 |

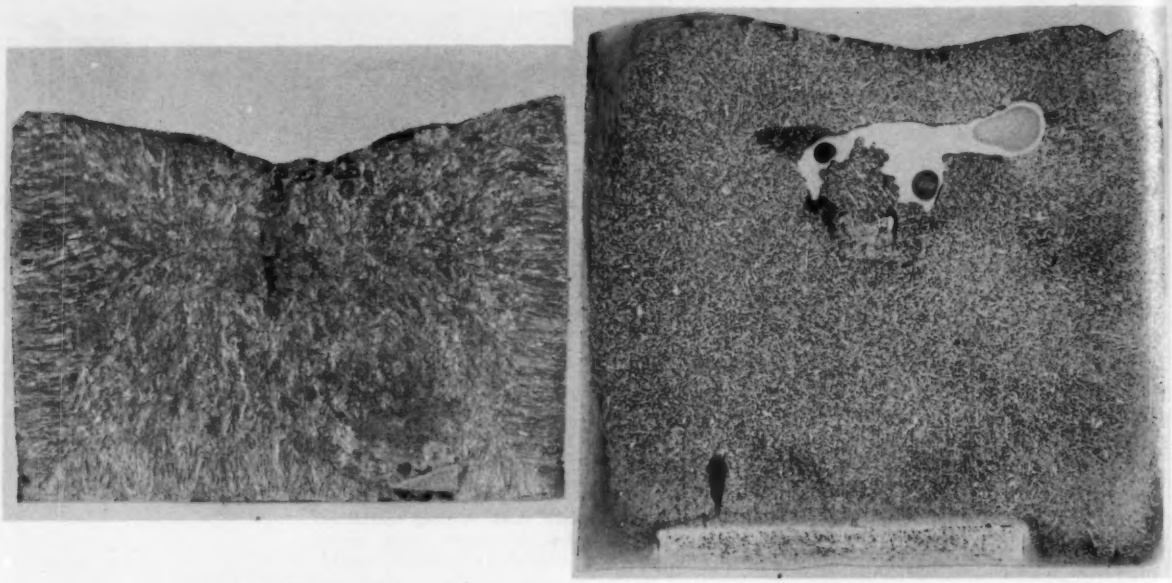


Fig. 7 — Grain refining addition influence on macrostructure. *Left* — Macrostructure of small steel casting containing 0.30 per cent Ti and 0.30 per cent ZrC. *Right* — sulfur print of small steel casting containing 0.50 per cent vanadium boron addition.

Although the tensile strength was increased by additions of titanium and titanium carbide, the ductility decreased sharply. The low ductility obtained with grain refining additions indicated that the strength values reported are actually fracture strengths. This overall effect of these additions was thought to be an alloying effect as well as the result of variations in grain size. The presence of precipitates, such as TiC, along the grain boundaries may well have contributed to the embrittlement tendency.

### *Weld Deposits*

The influence of the various additions on the weld structure of the weld deposits is listed, along with the chemical analysis of each deposited weld bead, in Table 5. The control sample in this investigation has an intercept number of 5, a relatively coarse size. Additions of 2.0, 1.0 and 0.5 per cent aluminum-zirconium alloy produced only slightly smaller dendritic arms or Widmanstätten ferrite, with intercept numbers of 6, 7 and 9, respectively. The fine aluminum powder added to the tubular electrode in amounts of 0.77 and 1.50 per cent was also not very effective and caused a slight refinement to 9 and 11.

However, additions of effective grain growth inhibitors such as 0.88, 1.82 and 2.73 per cent of ferrocolumbium resulted in considerable refinement to intercept numbers 8, 22 and 24, respectively. Equal amounts of titanium metal in the form of a ferrotitanium and titanium carbide were added to the electrode as 1.72 per cent ferrotitanium and 0.5 per cent titanium carbide. This combination of a nuclei former and growth restriction agent produced an incept spacing of 16. The finest weld structure obtained during this investigation was 36, with an electrode containing 3.45 per cent ferrotitanium and 1.0 per cent titanium carbide.

TABLE 5 — COMPOSITION AND GRAIN REFINEMENT OF SINGLE PASS WELD DEPOSITS WITH REFINING ADDITIONS

| Electrode Additives, % | Per Cent | | | | | | | Intercept Spacing |
|---|---|---|---|---|---|---|---|---|
| | C | Mn | Si | Al | Zr | Ti | Cb | |
| Control | 0.116 | 0.62 | 0.25 | — | — | — | — | 5 |
| 2.00 Al-Zr | 0.166 | 0.85 | 0.36 | 0.10 | 0.16 | — | — | 6 |
| 1.00 Al-Zr | 0.154 | 0.75 | 0.38 | 0.048 | 0.08 | — | — | 7 |
| 0.50 Al-Zr | 0.162 | 0.69 | 0.34 | 0.03 | 0.025 | — | — | 9 |
| 0.77 Al | 0.164 | 0.68 | 0.19 | 0.033 | — | — | — | 9 |
| 1.50 Al | 0.139 | 0.79 | 0.27 | 0.100 | — | — | — | 11 |
| 0.88 Fe-Cb | 0.128 | 0.65 | 0.23 | — | — | — | 0.07 | 18 |
| 1.82 Fe-Cb | 0.120 | 0.69 | 0.25 | — | — | — | 0.10 | 22 |
| 2.73 Fe-Cb | 0.120 | 0.68 | 0.29 | — | — | — | 0.21 | 24 |
| 1.72 Fe-Ti, 0.50 TiC | 0.144 | 0.69 | 0.37 | — | — | 0.09 | — | 16 |
| 3.45 Fe-Ti, 1.00 TiC | 0.176 | 0.76 | 0.54 | — | — | 0.19 | — | 36 |
| 3.00 Fe-Mn | 0.149 | 1.03 | 0.23 | — | — | — | — | 9 |
| 1.72 Fe-Ti, 0.50 TiC, 3.00 Fe-Mn | 0.184 | 0.95 | 0.37 | — | — | 0.08 | — | 26 |

NOTE: Sulfur and phosphorous were approximately 0.025 and 0.015%, respectively, in all cases.

TABLE 6 — GRAIN REFINING ADDITIONS INFLUENCE ON DENDRITIC SPACING AND MECHANICAL PROPERTIES OF AS-DEPOSITED WELD METAL

| Electrode Additives, % | Intercept Spacing | Ultimate Str., psi | Elong., % | Red in Area, % | 70 F Impact Str., ft-lb | Rockwell Hardness, Rb |
|---|---|---|---|---|---|---|
| Control | 5 | 68,700 | 32.5 | 65.2 | 78 | 76 |
| 2.00 Al-Zr | 6 | 72,300 | 30.0 | 56.3 | 20 | 80 |
| 1.00 Al-Zr | 7 | 70,700 | 29.0 | 59.0 | 39 | 82 |
| 0.50 Al-Zr | 9 | 81,000 | 30.0 | 65.0 | 54 | 81 |
| 0.77 Al | 9 | 76,000 | 28.0 | 62.2 | 75 | 81 |
| 1.50 Al | 11 | 84,200 | 28.0 | 60.5 | 55 | 82 |
| 0.88 Fe-Cb | 18 | 94,000 | 21.0 | 50.5 | 10 | 96 |
| 1.82 Fe-Cb | 22 | 99,000 | 17.0 | 43.6 | 3 | 98 |
| 2.73 Fe-Cb | 24 | 114,000 | 19.0 | 48.0 | 2 | 100 |
| 1.72 Fe-Ti, 0.50 TiC, 3.00 Fe-Mn | 26 | 110,500 | 14.0 | 29.0 | 8 | 95 |
| 3.45 Fe-Ti, 1.00 TiC | 36 | 118,000 | 20.0 | 57.7 | 2 | 99 |
| 3.00 Fe-Mn | 9 | 86,000 | 27.0 | 61.6 | 60 | 79 |
| 1.72 Fe-Ti, 0.50 TiC | 16 | 99,700 | 23.0 | 52.0 | 5 | 93 |

NOTE: Tensile properties average of two tests
Impact properties average of six tests
Hardness readings average of four readings.

With a simple deoxidizer, and an alloy with considerable solid solubility such as an addition of 2.4 per cent manganese, the spacing was refined slightly to 9. The same high level of manganese was maintained and 1.72 per cent ferrotitanium plus 0.50 per cent titanium carbide were included in another electrode. These additions resulted in a weld structure size of 26. In general, additions of titanium, titanium carbide and columbium produced dendritic arms or Widmanstätten ferrite finer than 15. Titanium, titanium carbide and columbium were much more effective in this respect than aluminum-zirconium, aluminum or manganese. These results are generally in agreement with the refinement of the cast ingots.

The seven photomicrographs in Fig. 8 indicate the influence of composition on the weld structure of the deposited bead. The control weld, made with an electrode containing no grain refiners, is shown in Fig. 8a. The slight refining effect of aluminum and manganese are illustrated in Figs. 8c and 8e. The dendritic nature of these welds is in evidence, and the dendritic arms and Widmanstätten ferrite can be observed. The refining effect of aluminum on weld metal has been found previously.[39] Zirconium additions appear to reduce this refinement somewhat when added with aluminum (Fig. 8b).

Additions of titanium and columbium reduce the size of the weld structure, as shown in Figs. 8d, f and g. The fine size obtained by additions of both titanium and titanium carbide with and without manganese is illustrated by Figs. 8f and g. Both the final composition of the weld deposit and additions are shown in Fig. 8. The titanium content of the weld deposits, as listed in Fig. 8 and Table 5, is a total figure resulting from the ferrotitanium and titanium carbide additions.

### *Size Influence on Properties*

The influence of the size of the as-deposited structure on the ultimate strength, elongation, reduction in area, room temperature impact properties and hardness are shown in Table 6. The size of the weld structure is plotted versus the tensile properties in Fig. 9. It is apparent that a continuous increase in tensile strength occurs with increasing fineness of the structure. The specimens containing 0.19 per cent titanium increased 72 per cent in ultimate strength over the control tests. However, the elongation decreased from 32.5 to 14.0 per cent with decreasing size.

Many of the tensile properties of refined deposits shown in Table 6 and Fig. 9 are above 20 per cent elongation, and are acceptable for many welding applications. The reduction in area also decreased with a finer structure. The specimen with FeTi, TiC and FeMn additions that yielded an analysis of the deposited weld of 0.08 per cent titanium and 0.95 per cent manganese exhibited the lowest ductility. The most highly refined steel deposit with additions of both FeTi and TiC had a high ductility (20 per cent elongation and 57 per cent reduction of area) for an average of the two tensile tests.

The embrittling effect of grain refining additions, however, is apparent with the impact data of Table 6, as plotted, together with hardness versus the size of the weld structure, in Fig. 10. The impact energy for fracture of the Charpy V-notch test in those bars from weldments to which Al-Zr, Al and Mn were added varied from 78 to 20 ft-lb. The dendritic spacing was only slightly influenced by these additions. However, when the effective grain refiners, such as titanium, titanium carbide or columbium were added to the weld metal, the impact properties decreased sharply.

All of the welds containing these latter materials exhibited a breaking energy of 10 ft-lb or less. The effect of the size of the dendritic arms or Widmanstätten ferrite on the impact properties was not as uniform as in the case of tensile properties. The sharp loss in impact energy after refinement was apparently the result of a brittle fracture at room temperature, and may have been due to an embrittling film at the dendritic boundaries.

The hardness followed the same trend as the tensile strength. A decrease in the fineness of the weld structure was accompanied by an increase in hardness. Columbium, titanium and titanium carbide produced hardness values as high as 24 Rockwell B above that

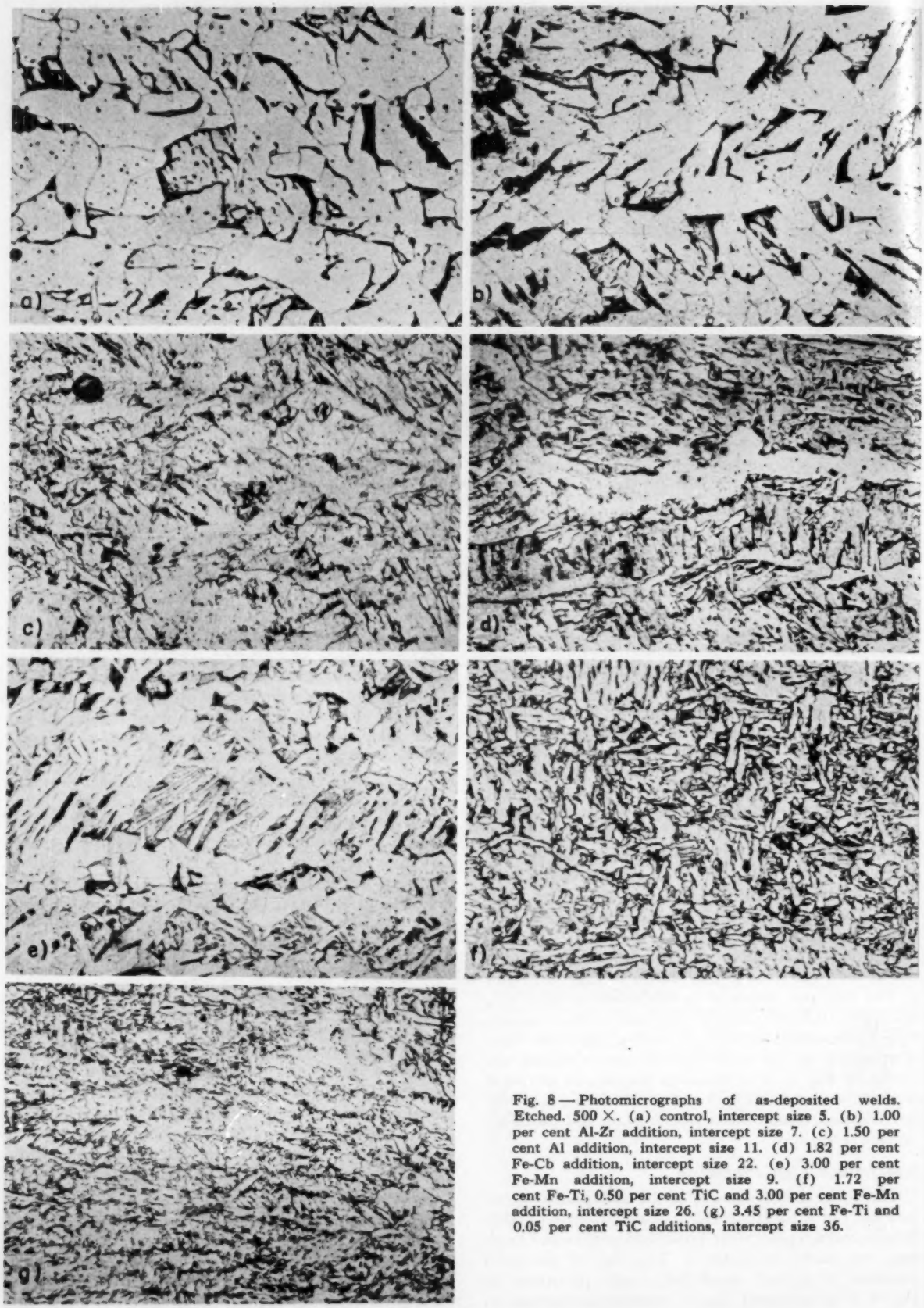



Fig. 8 — Photomicrographs of as-deposited welds. Etched. 500 ×. (a) control, intercept size 5. (b) 1.00 per cent Al-Zr addition, intercept size 7. (c) 1.50 per cent Al addition, intercept size 11. (d) 1.82 per cent Fe-Cb addition, intercept size 22. (e) 3.00 per cent Fe-Mn addition, intercept size 9. (f) 1.72 per cent Fe-Ti, 0.50 per cent TiC and 3.00 per cent Fe-Mn addition, intercept size 26. (g) 3.45 per cent Fe-Ti and 0.05 per cent TiC additions, intercept size 36.

Fig. 9 — Tensile properties of deposited weld metal as a function of weld structure size.

of the control electrode. The other additions that did not produce much grain refinement resulted in less than a 10 point Rockwell B increase in hardness that was probably at least partially an alloying effect.

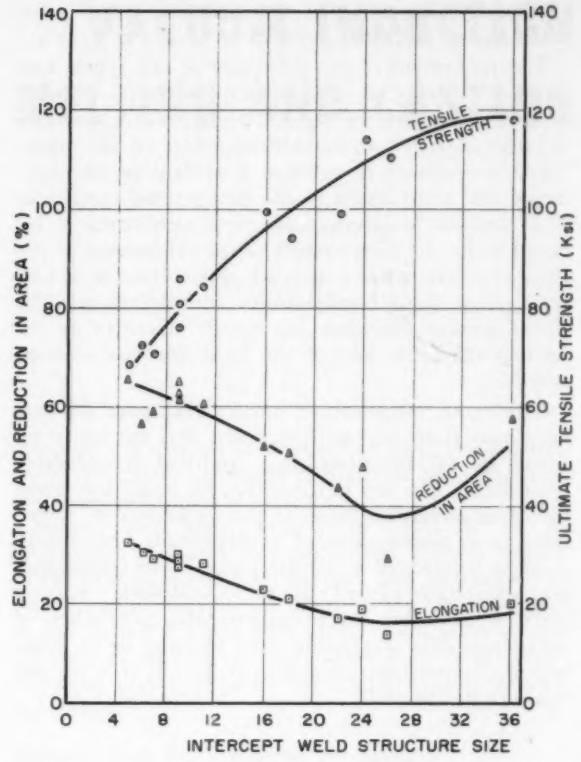
## DISCUSSION AND SUMMARY

This investigation indicates some of the major difficulties encountered when refining the grain size or dendritic spacing of as-cast or as-weld-deposited steel. The identification of the grain prevalent at the end of freezing requires a detailed and thorough analysis of all those facets, which are inherited from the as-cast grain boundary or dendritic structure, and which persist during the allotropic transformation occurring on cooling to ambient temperature. The relatively high melting point of steel restricts the number of potentially suitable inoculants because of the stability requirement imposed on such substances.

Many of the additions oxidize readily at the high temperatures involved and their effectiveness is lost or reduced. This problem, as well as the gradual dissolution of insufficiently stable inoculants, introduces a time factor into the problem of grain refinement, particularly in large slowly solidifying castings. This latter problem is not a serious one in welding, but oxidation of the additions during weld deposition obviously occurs.

The results have shown that small additions to the steel melt or weld deposit can produce considerable refinement. This refinement is thought to be the result of either the introduction of effective foreign nuclei or constitutional supercooling. The degree to which the various additions produced this refinement depended on their effectiveness in one or the other mechanisms and their remaining active in the solidifying steel.

The results of the grain or dendritic arm refining work with both steel ingots and weld deposits have shown that titanium and columbium added as ferro alloys are effective refining agents, probably because of their constitutional supercooling and grain growth inhibiting influence. Titanium carbide and vanadium boron were also both effective refiners, but their effect is attributed to providing a source of foreign nuclei for heterogeneous solidification. It is unlikely that the titanium and columbium when added as metals form carbides and serve only as inoculants in these low carbon steels. However, combinations of additions thought to restrict grain growth (Ti and Cb), together with materials believed to be nuclei formers



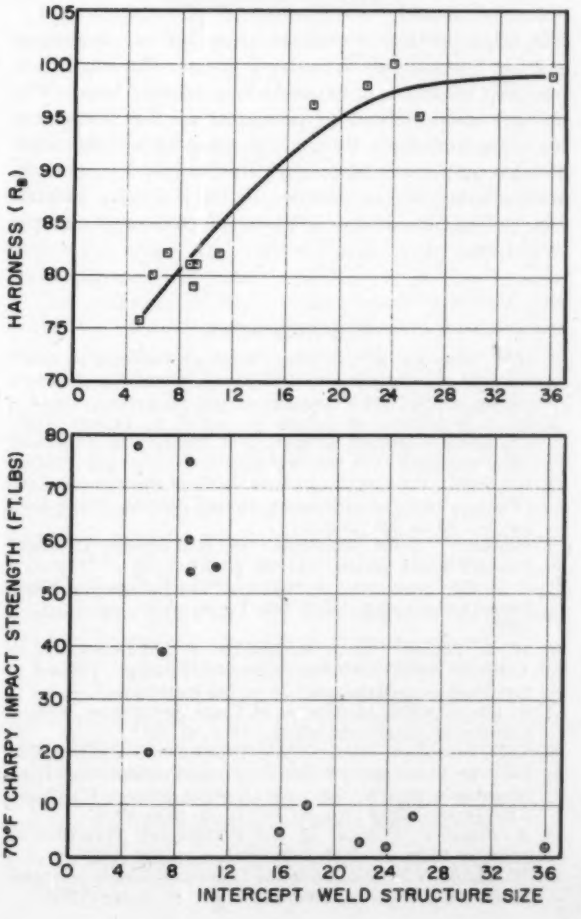




Fig. 10 — Impact and hardness properties of deposited weld metal as a function of weld structure size.

(TiC and V-B), were the most effective refiners for both castings and weld deposits.

The refinement of the grain size of cast ingots and dendritic spacing or Widmanstätten ferrite of weld deposits increased the tensile strength and hardness, but also imparted an embrittling effect on the metal. The improvement in strength is evident in the weldments that were tested in the as-deposited condition. The increase in strength in these weldments is believed to be the direct result of the refinement of the weld structure, since it is much larger than would be anticipated from hardenability considerations. The embrittlement decreases the tensile ductility of the castings markedly and of the weld deposits to some extent.

The room temperature impact resistance of bars machined from the weld deposits was markedly reduced by all additions that produced considerable refinement. The loss in ductility and toughness limits the commercial feasibility of this procedure of refinement in its present stage of development. The embrittlement is believed to be the result of an intergranular precipitate caused by these additions. Further work is necessary to investigate the possibility of preventing this precipitate from forming or of methods of eliminating this embrittlement if it is formed during solidification.

## ACKNOWLEDGMENT

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# VACUUM INDUCTION MELTING AND CASTING

by W. A. Matejka


## ABSTRACT

**Vacuum melting has proved to be a desirable process for the production of consistently clean, gas free, closely controlled alloys of castings of various sizes, shapes and weights. Success with high temperature alloys led to the adaption of the vacuum induction melting process to the production of high strength and bearing steels. Further applications were found in the electronic and atomic industry where high purity alloys are required. This process is and will remain a valuable production tool for the fabrication of materials requiring the highest degree of metal integrity. The various aspects of this process are detailed by the author.**


## INTRODUCTION

The rapid developments in aircraft engine design during and after World War II have posed many problems to the metallurgical industry. The gas turbine requires high strength, high temperature alloys in both cast and wrought forms. As the thrust of the engines increases, the property requirements of the materials become more and more stringent. Where, in the early days of jet design, only the buckets and turbine discs required high temperature alloys, now even compressor components, such as discs and shafts, demand the same high grade material.

To the metal producing industry this means that clean, gas free and closely controlled alloys in larger and larger ingots have to be provided. These attributes must be consistently present regardless of the size, shape or weight of the metal part. Vacuum melting has proved to be the desirable process for the production of such alloys. Vacuum induction melted alloys show improved workability and highly consistent properties from heat to heat. Forging, pressing or other conversion processes can be more readily performed.

In wrought parts the directionality of mechanical properties become less pronounced; especially notable is the improvement in transverse ductility. Transition temperature from the ductile to brittle state is also beneficially affected. Figure 1 shows an example of the improvement possible over air melting. These results may be ascribed to close chemical control of the metal composition, low gas content of the bath, the possibility of using larger amounts of titanium and aluminum and a higher degree of cleanliness.

Success with high temperature alloys led to the adaptation of the vacuum induction melting process to the production of high strength and bearing steels. Further applications were found in the electronic and atomic industry, where high purity alloys are required. The use of vacuum melted material for forging and extrusion dies is another application that is steadily increasing.

### *Vacuum Induction Melting Advantages*

The outstanding advantages of vacuum induction melting have led to a tremendous growth in this branch of the industry. Figure 2 illustrates the increase of vacuum induction melting capacity over the past nine years and its projected growth. Special attention is directed to the steady climb of the maximum possible ingot weight. By far the majority of all vacuum induction melted alloys are cast into ingots for subsequent conversion into billet or barstock. The demand for larger and larger ingots poses substantial technological problems.

The majority of high temperature materials produced by vacuum melting are of nickel base alloyed with chromium, cobalt, molybdenum in varying amounts and with additions of aluminum, titanium, boron, zirconium and tungsten. The high strength at elevated temperatures is obtained by a precipitation hardening process, the hardners being aluminum, titanium, boron and zirconium. These highly alloyed compositions solidify over a wide mushy range, and therefore tend to severe shrinkage and are prone to segregation. A further complicating characteristic is their relatively low thermal conductivity. These facts immediately point out the difficulties to be encountered in producing satisfactory ingot products.

To the foundryman accustomed to the production of complicated cast shapes, the problem of pouring an ingot seems rather simple. It will be difficult to convince him that an ingot actually is a complicated casting, especially when poured under vacuum. These requirements have to be met in an ingot to make it suitable for the subsequent conversion processes:

1. The chemistry has to be within narrow specifications.

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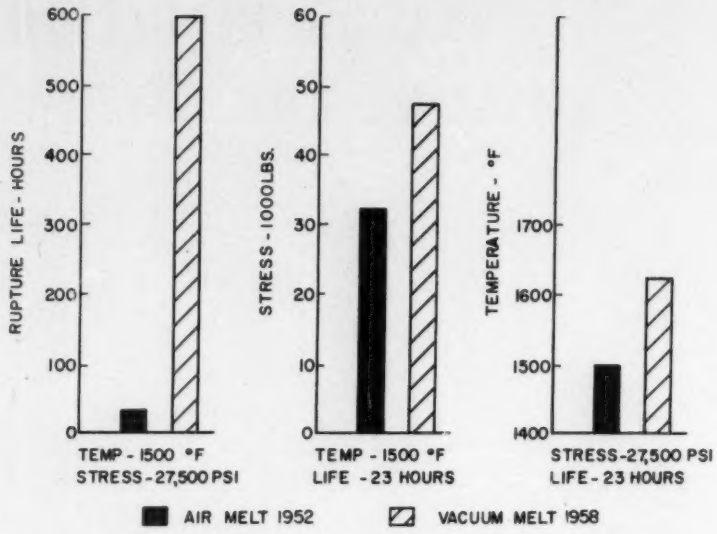



Fig. 1 — Specification for air-melted material in 1952 compared with vacuum melted specifications in 1958.

2. There must be no segregation.
3. The ingot must be dense.
4. Nonmetallic inclusions are not permitted.
5. The grain structure should be uniform.

These points shall be considered in more detail later.

## THE PROCESS

For those not familiar with vacuum melting, a short process description is given. More detailed information is available in the literature.[1,2] The induction heated crucible is contained within a water-cooled tank. The induction furnace assembly is similar to the air melting units, but special attention has to be paid to the maximum voltage on the uninsulated coil to prevent corona and arcing in the 10-500 micron pressure range.[3] Modern vacuum furnaces are provided with a secondary low frequency induction coil to induce stirring of the melt to increase the reaction rate at the metal/vacuum interface.[4]

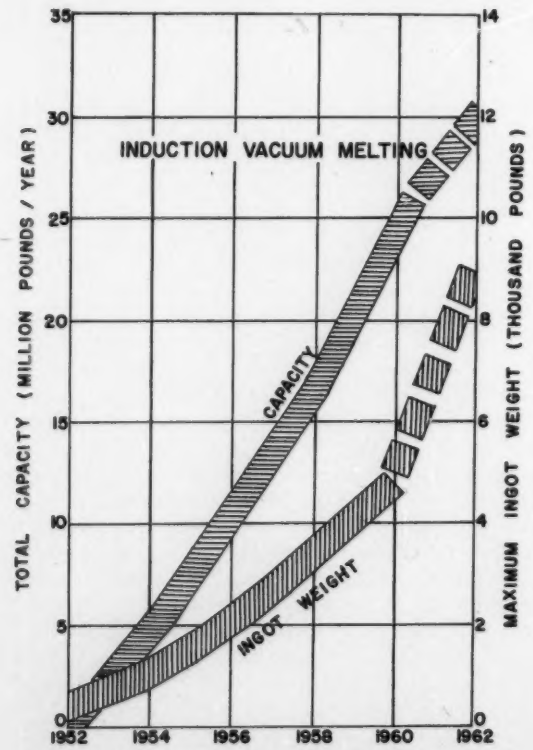



Fig. 2 — Growth rates of vacuum induction melting capacity and maximum ingot weight.

The tank contains fixtures to accept one or a multiple number of ingot molds. Special addition cups within the unit permit alloying operations during the melting process without breaking the vacuum. It is also possible to take dip samples from the bath and make immersion temperature measurements through a vacuum lock system. The pump system consists, depending on the furnace size, of a number of mechanical pumps for the medium pressure range and oil diffusion pumps for the low vacuum range.

The operating pressures of modern vacuum induction furnaces are between one and 10 microns. One micron = $10^{-3}$ mm of Hg = 1.32 $0.10^{-5}$ atmosphere (Fig. 3). Some vacuum furnaces have been equipped with centrifugal casting machines making it possible to cast shapes at speeds resulting in forces 75 times gravity without loss of vacuum.

### *Typical Melt Cycle*

Carefully selected raw material is weighed on an electronic scale, which will indicate acoustically any possible mistake in the charge composition. The raw material is deposited in a determined sequence in the charge can, and additions of reactive elements are placed separately in the addition cups. After lowering the cold charge into the crucible, the tank is closed and the vacuum pump system is connected to the furnace. After the entire system is pumped down, the electrical power is put to the coil.

As shown in Fig. 4, the pressure will decrease until gases contained in the raw materials are set free as the

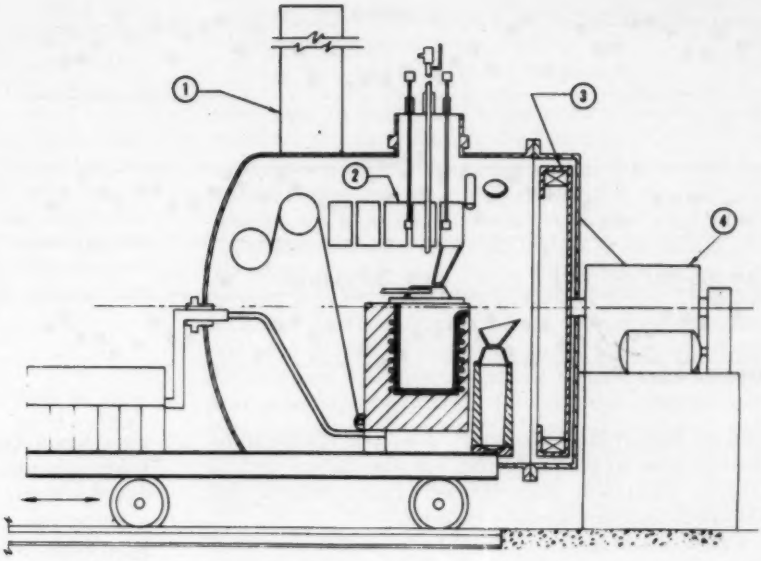



Fig. 3 — Cross-sectional view of a 1000 lb vacuum furnace. (1) control panel, (2) addition cups, (3) centrifugal casting mold and (4) centrifugal casting machine.

charge begins to melt. To prevent excessive boiling the pumps are blanked off and the pressure is permitted to build up in the tank until the major reactions subside. At this time full vacuum is applied, and the stirring power is switched on thus starting the refining period. This refining process is accomplished by following tightly specified procedures. Only then are the addition of reactive elements such as titanium and aluminum made. An additional mixing cycle insures proper distribution of the added constituents.

Now a dip sample is taken without breaking the vacuum. While the sample is submitted for spectrographic analysis, the molten metal is held approximately 50 F above the liquidus temperature. If necessary, corrections of the melt composition are now made, and after another mixing period the analysis is recontrolled. Only after the release from the spectro department is the metal poured into the molds. The ingots are permitted to cool down before the vacuum is broken.

### *Chemistry Control*

By following the melt procedures as outlined it is possible to control the chemistry of the alloys within narrow limits. In Fig. 5, the fluctuation of boron and aluminum contents in 40 consecutive heats of M-252 are illustrated. The bands outline the 26 limits. By far the majority of results are well within the scatter band determined by the possible accuracy of the spectrographic analyses. In the lower part of the diagram are shown the elevated temperature tensile properties for these heats. A maximum spread over 8000 psi is maintained; in other words reproducibility within ± 4000 psi is commercially possible. These results satisfy an earlier mentioned requirement for a satisfactory ingot product, namely, that close chemical control is necessary.

The other four requirements, absence of segregation, high density, absence of inclusions and uniform grain structure shall be discussed together since they are closely related. To better explain the problems of casting large ingots under vacuum, a few remarks about the special conditions existing in a vacuum furnace will be made.

### *Physical Conditions in a Vacuum Furnace*

Because of the low pressures a specific thermodynamic system is established, which means that certain chemical reactions not encountered in standard air melting procedures now enter into the picture. To predict such reactions the vapor pressure curves for pure metals and free-energy-of-formation diagrams become essential data.[5,6,7] These point out reactions that will take place, but they do not disclose the rate at which such reactions may proceed. Without going further into the physical-chemical aspects of vacuum melting, one important reaction will be mentioned — the metal/crucible reaction.[8,9,10]

This reaction is of great importance because all of the beneficial effects promised by the process can be nullified at the metal/refractory interface. In addition to the crucible, the runners, nozzles and mold washes are also possible sources of contamination. The proper selection of refractory materials therefore becomes of

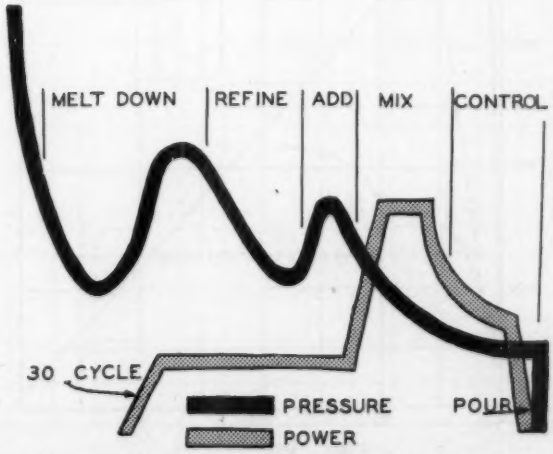



Fig. 4 — Pressure and mixing power changes during a typical vacuum melting cycle.

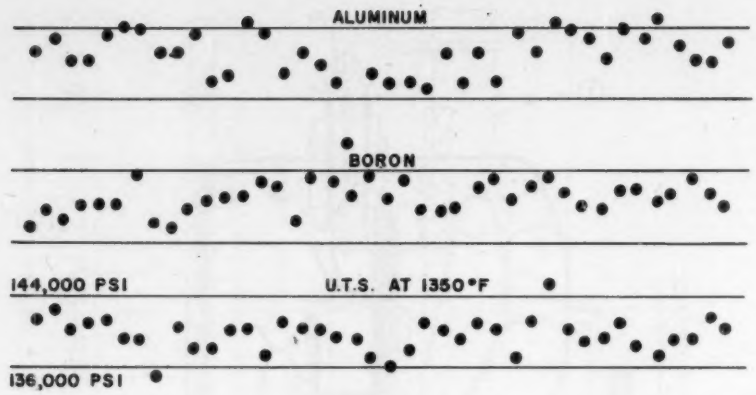



Fig. 5 — Fluctuations of aluminum and boron contents and ultimate tensile strength (1350 F) in 40 consecutive heats of alloy M-252.

crucial importance. Not only does the chemical reactivity have to be considered, but also factors such as erosion resistance, heat transfer characteristics and the coefficient of thermal expansion.

Crucible life is an important factor in the economics of vacuum melting. The inability to produce satisfactory linings for large crucibles has long been a limiting factor in the development of larger induction melting units. By proper combination of chemical composition, grain size distribution, ramming and sintering procedures it has been found possible to prepare 5000 lb crucibles with adequate life.

Most vacuum induction crucibles are rammed with sintered and crushed $MgO$-$Al_2O_3$ refractories. The mix, usually 70:30 per cent by weight, is dry rammed without binder or fluxes. At the low pressures of one-10 microns some metal oxides of the crucible refractory are reduced by the carbon dissolved in the metal. The reaction products are absorbed by the bath. In the case of magnesia crucibles, the reduced magnesium has a high vapor pressure and is pumped off. When alumina is used as crucible material the reduced aluminum is dissolved in the melt.

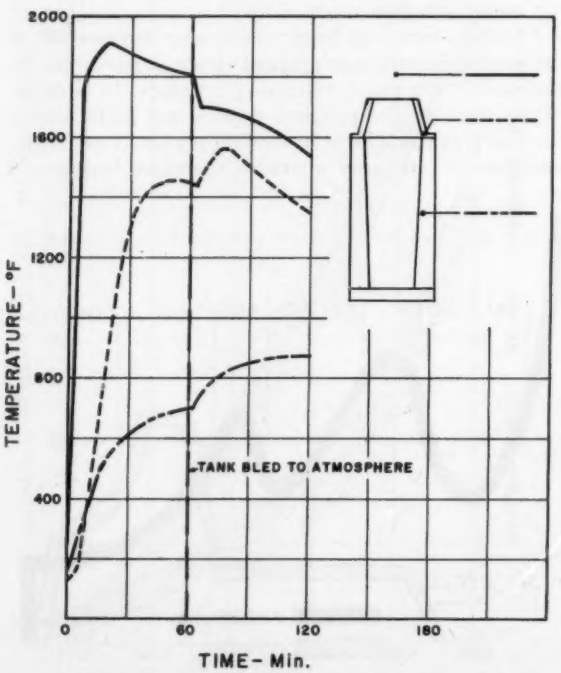



Fig. 6 — Influence of vacuum on the heat transfer characteristics in an ingot casting setup.

Since this element is normally present as a high percentage alloying constituent, the metal/crucible reaction is not detrimental to the process. The kinetics of this reaction depend on the carbon and oxygen levels in the bath, the temperature, the area and length of contact time between metal and crucible. Silica is not suitable for vacuum melting crucibles since the reduction reaction is so favored that violent boiling results.

### *Metal/Mold Reaction Problem*

The metal/mold reaction problem was mentioned to show that a careful selection of refractory materials must be made when shaped castings are to be poured under vacuum. The metal/mold reaction becomes of great importance in heavy castings where fast freezing is not possible.

In addition to the thermodynamic considerations, the low pressure alters other physical conditions within the furnace. The lack of atmospheric pressure (one atmosphere equals 14.7 psi) impairs pressure feeding of castings. One atmosphere is equivalent to the ferrostatic pressure exerted by a 51½-in. high feeder head or riser. On the other hand, the vacuum eliminates back pressure in the mold cavities, thus permitting easy mold filling of thin cross-sections.

Since the help of atmospheric pressure in the feeding process is nonexistent, dense ingots can be obtained only by inducing strictly directional solidification. In this process the establishment of proper heat gradients is not always simple. Since the vacuum surrounding the ingot mold is a good insulator, the extraction of heat from the solidifying casting must therefore depend to a high degree on radiation. Early in the pouring, the solidifying ingot shrinks away from the mold wall thereby limiting the direct metal-to-metal contact to the stool area.

All other surfaces are insulated by vacuum. The influence of this vacuum layer on the heat transfer from ingot to mold is shown in Fig. 6. For this experiment, three thermocouples were installed in a mold set up for the production of ingots weighing 2400 lb. The first thermocouple is located at mid-height in the ingot mold 1¼-in. from the metal/mold interface. The second one is installed in the hot top. To determine

the heat losses from the hot top surface a third thermocouple was suspended 7 in. over the free radiating surface. The diagram shows a marked temperature increase as soon as the furnace tank is bled to atmosphere.

The temperature differential in the hot top is more pronounced than in the mold. This is due to the distance of the hot top junction from the heat source. The sharp drop of the curve showing the suspended thermocouple temperature is explained by the inrushing cold air which lowered the ambient tank temperature. The temperature increase in mold wall and hot top after bleeding is over 100 F.

### *Alloy Solidification Time*

Another important factor to be considered in these investigations is the solidification time of the alloy. Since for high temperature alloys no such information was available in the literature, it had to be determine experimentally. Its aim was to determine the approximate rate of solidification of the alloy 1002B. This nickel-rich material, less complex in its composition than other nickel base alloys, has the following analysis – C 0.05, Cr 13.00, Ni 40.00, Mo 6.00, Ti 2.50, Al 0.20, B 0.01 – 0.05, Fe Bal.

This material is used in turbine and compressor discs and parts where high tensile, creep and rupture strength are needed in the 1000-1400 F range. For the experiment a well insulated cylindrical mold was placed on a heavy copper chill plate. The metal was cast under vacuum with a superheat of 150 F and the mold was covered immediately after end of pour. A series of Pt-Pt/Rh thermocouples were imbedded in the mold sidewall as close as possible to the metal/mold interface, and the couples connected to a 16 point temperature recorder. The results are plotted in Fig. 7, which shows the well known parabolic relationship between the freezing rate and time, as described in the literature. On the basis of this graph an equation is offered for the vertical freezing rate of alloy 1002B:

General equation: $D = q\sqrt{t} - C$ (in./min$^{1/2}$)
Equation for proprietary alloy:

$$D = 1.11\sqrt{t} - (1.70) \text{ (in./min}^{1/2}\text{)}$$

Additional information on the general solidification equation will be found in the references.[11,12,13,14]

When considering the graph in Fig. 7, it should be observed that the plot was established for vertical freezing when intimate contact with the chill plate was maintained. To relate the values to horizontal or transverse solidification rates, the formation of the vacuum gap has to be introduced as a rate limiting variable. A determination of the horizontal freezing rate of a particular ingot, by using the values from the diagram, will therefore result in minimum values. Also to be considered is the fact that the temperatures were measured through a thin layer of insulator, and that as a result the plotted line may need a parallel displacement to compensate for this temperature differential. For this reason the constant *C* is placed in parenthesis.

### *The Ingot Mold Setup*

These observations are useful in setting up proper conditions in the ingot/mold system to promote directional solidification. The first requirement is to establish the necessary heat gradients. For practical and logical reasons it is advisable to split the mold system into three subcomponents – the stool, the mold and the hot top. By doing so, it is possible to evaluate the influence of each segment separately by changing only one variable at a time.

A large ingot casting is preferably cooled from the bottom up. A heavy chill block of copper is therefore used as a stool. Graphite, with its good heat transfer and thermoshock characteristics, would be equally suitable except that it would be severely eroded by the impact of the pouring stream. Since the free drop of the metal is well over 50 in., special precautions must be used even in connection with copper stools.

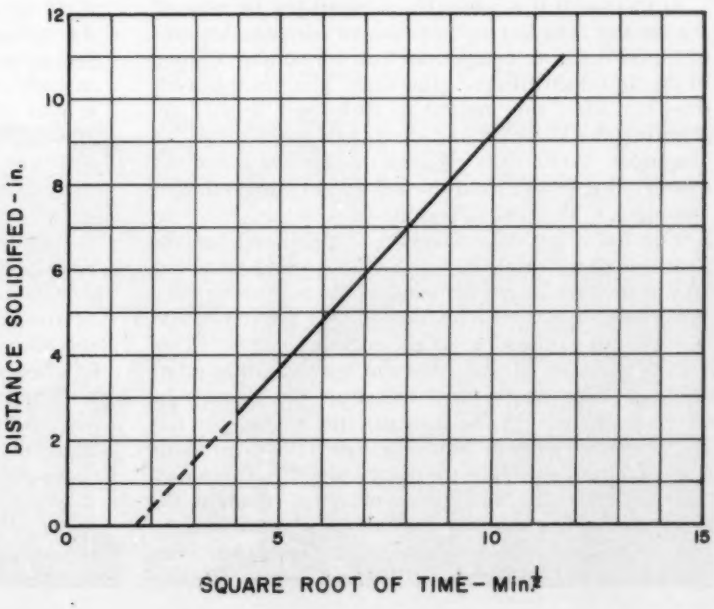



Fig. 7 — Solidification time of alloy 1002 B.

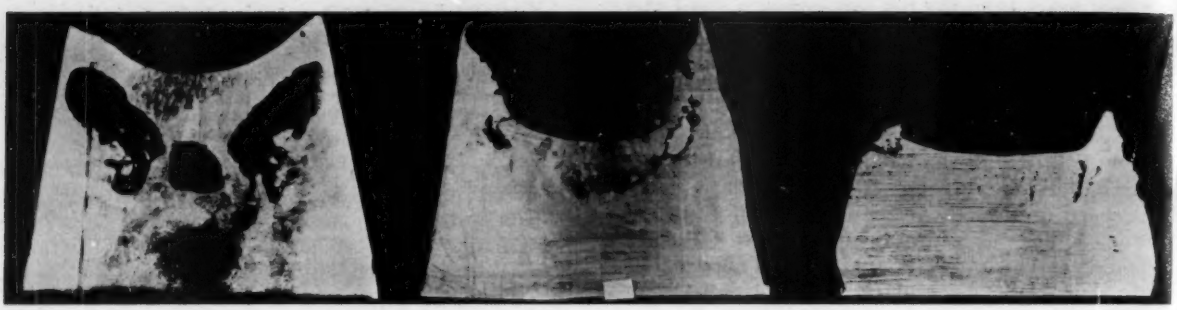


Fig. 8 — Improvement of hot top feeding characteristics.

The mold, usually made from high grade cast iron, is placed in intimate contact with the stool to prevent run outs and assure heat flow from mold to stool. The mold to ingot weight ratio is kept well over 1.5 to provide enough mass for heat absorption. The cross-section of the mold increases from top to bottom to establish a thermal gradient. Where it is necessary or desirable a mold coating can be applied; spray coatings of magnesia or alumina-based material have proved to be satisfactory. In the cast of split, as opposed to solid molds, plasma or metal spray coats are possible, although the economics of this technique are not as yet favorable since the adhesion of the applied layer to the mold wall is not under full control. Spalling due to thermoshock could result in inclusions in the solidified ingot.

### *The Hot Top*

In vacuum melting where metal recovery due to the premium metal price is of prime importance, hot top design is especially critical. Unlike the practice with killed steels, the ingot hot top of high temperature alloys is usually not cropped prior to the conversion process. It is entirely possible that sound segregation free material even above the hot top line can be forged into good parts.

At this point it is advisable to recall the purpose of the hot top. The hot top provides an adequate amount of liquid metal to compensate for the volume change of the liquid shrinkage of the alloy. The hot top itself does not affect the amount of shrinkage; it only determines the shape of the cavity and its location in the ingot. To be fully efficient the hot top has to retain its heat longer than the last solidifying section of the ingot.

The following data have been established for the heat loss distribution in steel hot tops when using fire clay as insulators – 15 per cent conducted through the refractory, 60 per cent retained by the refractory and 25 per cent radiated from the top surface.[15,16] This clearly points to the fact that the heaviest losses occur through heat absorption; therefore, the lower the thermal capacity of the insulator the higher the hot top efficiency. Because of the limited stability of many commercially available insulators under vacuum-and-metal contact conditions other means of retaining the heat in the hot top have been investigated.

One method, the application of exothermic compounds, is not advisable in vacuum casting of large ingots, because the thermal movement induced by the exothermic reaction increases greatly the chance of entrapping foreign material in the ingot. Moreover, there is diffusion of undesirable reaction products into the cast metal. Arc hot topping with nonconsumable electrodes has been used successfully, but it also carries the danger of possible contamination. Particles of the electrode can drop into the liquid pool. In the case of tungsten, such particles will sink until they are entrapped in the mushy zone.

### *Process Improvement*

If graphite electrodes are used, carbon pickup can result. Figure 8 shows the cross-section through the hot tops of three large ingots cast under vacuum; the obvious improvement in feeding characteristics was the result of an intensive development program. The hot top on the left is a typical example of the shrink cavity appearance before the start of the program, when there was no thermal barrier between the hot top and the mold and only an insufficient amount of insulation. The improvement shown in the other samples was due to redesign of the setup and proper adjustment of pouring temperature and pouring time.

The ingot weight limit is still 2400 lb of cast weight. The cross-section through the top half of such an ingot is shown in Fig. 9. The dark area is an optical effect created by the converging machine marks from the grinder. Three different grain structures can be distinguished. The pronounced columnar structure extends over 40 per cent of the entire cross-section. It goes over into a zone of uniformally small equiaxed grains. In the upper quarter of the ingot the equiaxed grains grow larger.

During the solidification process of many alloys these three distinct zones are found. At the metal/mold interface a solid layer is immediately formed. Adjacent to the solid is a zone of mushy material composed of solids and liquid material. Towards the thermal center of the casting all is liquid. Depending on the alloy composition and the established heat gradients, the width of the mushy zone will vary within definite limits. In Fig. 9, the columnar zone represents the area that solidified under a steep heat gradient.

As the effect of the chill mold becomes less pronounced and as the superheat of the melt is dissipated, the heat transfer characteristics of the solidified metal take over as the rate limiting factor, and the temperature gradient becomes less steep. The solidi-

fication front gets much wider and relatively slow equiaxed grain growth is induced. The theory for columnar and equiaxed grain growth is explained in great detail in the literature. The study of this valuable information is highly recommended to every foundryman.[11,12,13,14]

### *Larger Equiaxed Grain Theories*

Two possible explanations for the larger equiaxed grains in the upper quarter of the ingot are suggested.

It can safely be assumed that in an adequately hot topped ingot the thermal center is located at the widest point of the casting or, in this particular case, at the mold/hot top line. This area will be the last to solidify. Therefore the crystals have the longest time to grow. Another possible explanation for the large grains is based on the theory that a skin is formed at the upper hot top face. This again is a region where steep heat gradients exist and columnar grain growth prevails. It is quite possible that some dendrites are broken off due to convection currents or external shocks to the mold. These crystallites start to settle due to gravity, and being exposed to supercooled liquid they will be in a favorable condition to grow.

There are several ways to change or influence the grain growth characteristics of alloys. The application of vibration to the mold or to the solidifying metal has been used successfully for grain refinement.[17,18] Oscillation of the mold during and after pour will lead to the same result.[19] Inoculation with crystallographically similar materials has been suggested as a nucleation procedure.[20] The question still remains, not fully answered, as to which grain formation is most suitable for a given required purpose.

The latest work conducted by MIT strongly supports the theory that unindirectionally solidified ingots will deliver the most advantageous cast product. Whatever the final answer to the problem will be, the vacuum inducting melting and casting process has the flexibility to follow any of the suggested approaches. The outlook is bright. With the availability of larger and larger units the maximum charge weight has been pushed up to 5000 lb. Special shapes have been cast in a weight range of 4600 lb. The furnace that makes the casting of such loads possible is shown in Fig. 10.

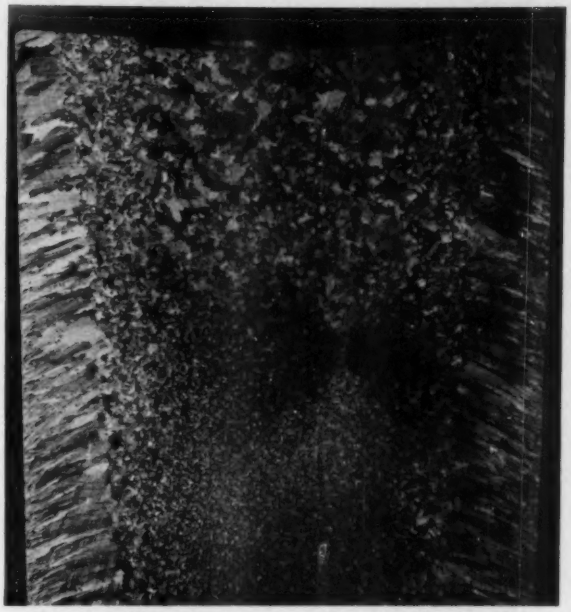


**Fig. 9 — Macrostructure of a typical 2400 lb alloy 1002 B ingot (only upper half of ingot is shown).**

To illustrate the ability of such a large vacuum induction furnace to maintain the same high quality requirements as smaller units on a production basis, the data of $x$ heats of converted alloy 1002B are assembled in a block diagram, Fig. 11. The favorable comparison of the specification limits with properties actually obtained points to the fact that, by using sound melting and casting techniques, the vacuum induction process is and will remain a valuable production tool for the fabrication of materials requiring the highest degree of metal integrity.

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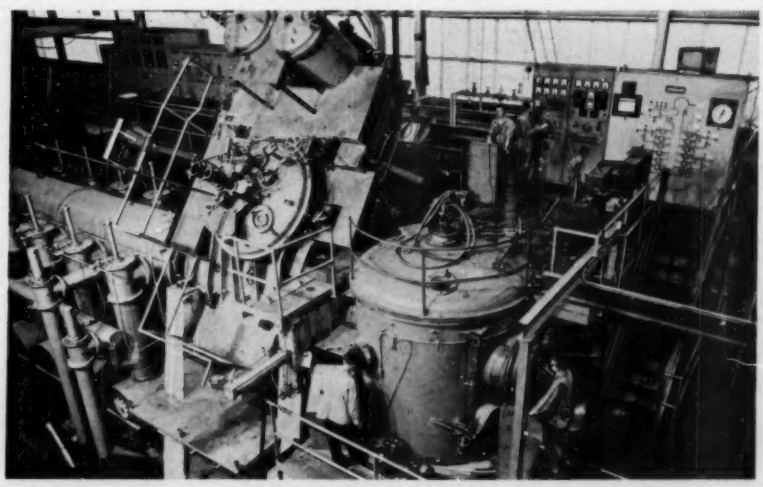



**Fig. 10 — 5000 lb vacuum induction melting furnace (pouring position).**

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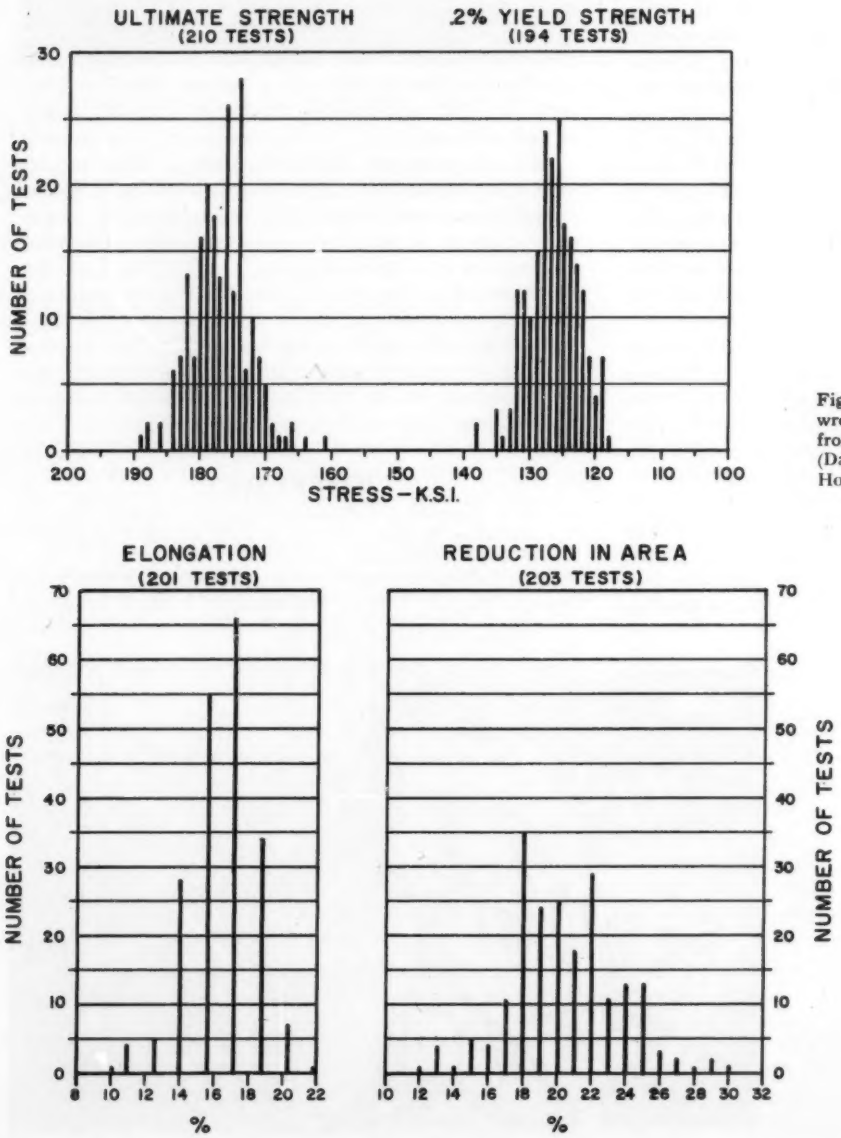




**Fig. 11 — Mechanical properties of wrought alloy 1002 B parts converted from vacuum induction melted material.** (Data courtesy Cameron Iron Works, Houston, Texas)

# HIGH INTEGRITY STEEL CASTINGS FOUNDRY PROCEDURE DEVELOPMENT

by R. J. Ely and J. E. Fries


## ABSTRACT

An extensive research program in both a process and a technique for production of high integrity, high strength steel castings larger in size than heretofore found in the casting industry has been developed. This report covers the three phases of this program:

1. Research methods and results.
2. Application of results to structural casting production.
3. Summarization of current state of casting art and suggested areas for further research and development.

High surface quality, high degree of soundness, reduction of cracks and hot tears and a greater reliability factor have resulted from this program at the authors' company. Case histories of castings which have been improved through the work of this extensive program are presented.


## INTRODUCTION

It is the intent to discuss three major areas related to high strength, high integrity steel castings. The first of these areas will concern some of the research methods used and the results obtained in the course of continuing investigations being performed in this field. The second will be devoted to a discussion of the application of these results to the production of structural castings. The final area will be an attempt to summarize briefly the current state of the art, and to suggest some fruitful areas for further research and development.

As a starting point it seems necessary and desirable to define what is meant by high strength and high integrity, and also to consider whether these two qualities must always go together. High strength steel castings, to the authors, mean castings in which mechanical properties of the order of 180,000 psi or better are specified. This type of casting normally carries a high integrity specification in addition. High integrity on the other hand does not necessarily mean a concurrent high strength requirement.

There are many applications where service loadings and design considerations are such that substantial ductility or impact strength may be the determining parameters. Under such conditions a high degree of structural integrity, that is, freedom from hot tears, shrinkage or surface imperfections, may be just as important in a 125,000 psi specification as in one which calls out 260,000 psi.

It must be admitted that at the current time the majority of applications which have involved high strength steel castings have been associated with military requirements in both airborne and ground support equipment, where weight and/or space saving has been of a sufficiently critical nature to warrant the premimum costs incurred in producing castings of the quality level required. In the long run it is felt that sufficient experience and capacity will be available to reduce these costs and attract a larger segment of the commercial market.

From the point of view of both convenience and logic, research and development in the casting field may be divided and discussed under two major headings. These are:

1) physical metallurgy research wherein the bulk of the work is performed using simple test bars and where the aim is to develop the optimum analysis, heat treatment, melting practice, etc., for a particular class of alloy.
2) process research in which the basic effort is devoted to translating the results of the physical metallurgy program into usable structural parts capable of competitively performing a specific function.

It is the intent to discuss both methods for, and results of, physical metallurgy and process programs aimed at improving the quality level of steel castings.

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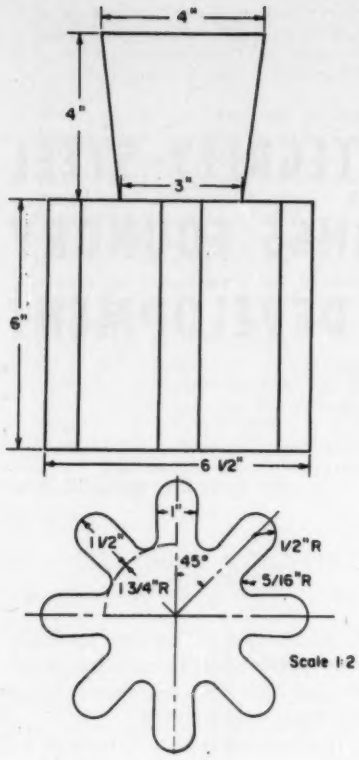



Fig. 1 — Test bar casting for one in. section size.

## STEEL CASTING RESEARCH

### *Physical Metallurgy*

One of the issues which arises almost the instant that foundry and design engineers begin to discuss their mutual problems is the question of separately cast test bars versus bars cut directly from the casting. If the distinction is made between physical metallurgy research and process research there can be absolutely no argument about the real need for a series of well designed reproducible test bars, if the variables of analysis, heat treatment, melting practice, etc., are to be effectively explored.

To use anything but such a series of reproducible test bars can lead only to erratic results and questionable conclusions relative to the capability of the alloys being investigated. No one can question the need of bars cut from castings or some similar approach when the time comes to transfer a new alloy development to functional castings. At this point, however, the problem becomes one of process research and must be treated accordingly. In the last analysis, the aim in both cases is reliability; in the first instance it is the reliability of the materials research itself which must be assured; in the second it is the reliability of the end product which must be confirmed.

In looking first at the area of physical metallurgy research on high strength cast steels, consideration must be given to the types of tools that are required and how these tools can best be applied.

First and foremost and obviously some sort of foundry facility must be available. There is no rule as to how large or how small such a facility should be any more than there is a requirement that an effective penetrant inspection can only be performed in a stainless steel emporium rather than an inexpensive beaker. It is important, however, that the equipment be able to reproduce in heat weights, casting section sizes and melting methods those situations which are to be encountered in actual production, as all of these are vital variables in insuring long range duplication of research results under operating conditions.

The foundry facility should be supported by the normally required tools of any physical metallurgy research program such as chemical, metallographic and mechanical test laboratories. Additionally required are inspection, heat treat and machine shop facilities. All of the above should be located in close proximity if an orderly and effective program is to proceed with any degree of efficiency. Finally, it is felt essential that such a facility be staffed by individuals who are free from the daily routine of getting tonnage out the door but can instead "step out of the main stream of production and ask themselves the question 'why'."[1]

Unfortunately, insufficient facilities of such a nature exist. If, however, high strength steel castings are to achieve their greatest potential it is essential that substantially more effort be devoted to the type of programs which can be best and perhaps only performed by such facilities.

### *Test Bars*

Before reviewing briefly several recent developments in the physical metallurgy of high strength steels, it seems appropriate to return for the moment to the question of test bars and present some illustrations of typical examples which have been used in different phases of the programs which will be discussed. Figures 1 and 2 show the types of bars used to obtain the data for two programs performed by the authors' organization for the Wright Air Development Division.[2,3] The basic design of these bars comes from work performed by Wayne, Bishop and Taylor[4] at the Naval Research Laboratory.

The smaller of these two bars is a particularly effective substitute for the conventional keel block, not only because of improved casting yield, but also in terms of reproducibility. Table 1 compares the results obtained from the eight legs of one test bar poured in a modified 4330 heat treated to the 180,000 psi strength level. Additional standard bars have been utilized for various process research programs and will be covered later.

It is essential to realize that it is not expected that comprehensive studies of a cast modification of 4330 or 8740 will reveal drastic differences in fundamental metallurgical behavior from that shown by the wrought grades. On the other hand, when the required degree of reliability is considered it becomes essential to find out wherever possible just what differences between cast and wrought grades may exist, how these differences must be handled in processing and how they may influence the various design factors. Until this is done, all that exists is a cast alloy, not an

engineering material. The latter is what is required.

The two W.A.D.D.[2,3] reports previously mentioned have resulted in several important findings pertinent to improved metallurgical quality in the high strength low and medium alloy steels. Figures 3 and 4 show two of the observed effects of using high purity melting stock which maintains phosphorus and sulfur levels at 0.010 per cent or below. Additional, and as yet undetermined, variables are also working to provide the observed improvement as the addition of phosphorus and sulfur to high purity charges to obtain what might be reasonably expected levels from conventional melting stock does not result in the predicted degradation of ductility.

These data also indicate that high purity plus an 1850 F (1010 C) homogenization is superior to normal purity and a 2280 F (1249 C) homogenizing treat-

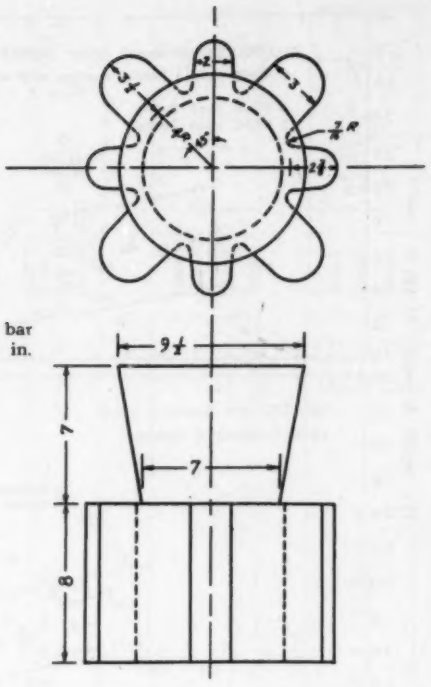



Fig. 2 — Test bar for 2 and 3 in. section sizes.

TABLE 1 — REPRODUCIBILITY OF MECHANICAL PROPERTIES FROM OCTABAR DESIGN TEST BAR

| Heat No. | Composition, % C | Mn | P | S | Si | Ni | Cr | Mo |
|---|---|---|---|---|---|---|---|---|
| 60-437 | 0.35 | 0.79 | 0.016 | 0.007 | 0.50 | 1.89 | 0.93 | 0.40 |

Heat Treatment: 1850 F-3-AC
1600 F-2-OQ
875 F-2-AC

Mechanical Properties

| Bar No. | Yield Str., psi | Tensile Str., psi | Elong., % | R.A., % |
|---|---|---|---|---|
| 60-437-4A | 183,900 | 198,000 | 9.5 | 28.7 |
| -4B | 182,700 | 200,500 | 10.5 | 29.8 |
| -4C | 182,400 | 198,000 | 10.0 | 39.2 |
| -4D | 183,900 | 199,000 | 10.0 | 29.6 |
| -4E | 185,400 | 201,000 | 10.0 | 29.5 |
| -4F | 183,600 | 199,000 | 10.0 | 29.2 |
| -4G | 186,600 | 202,000 | 10.0 | 28.8 |
| -4H | 182,700 | 199,000 | 9.0 | 25.1 |

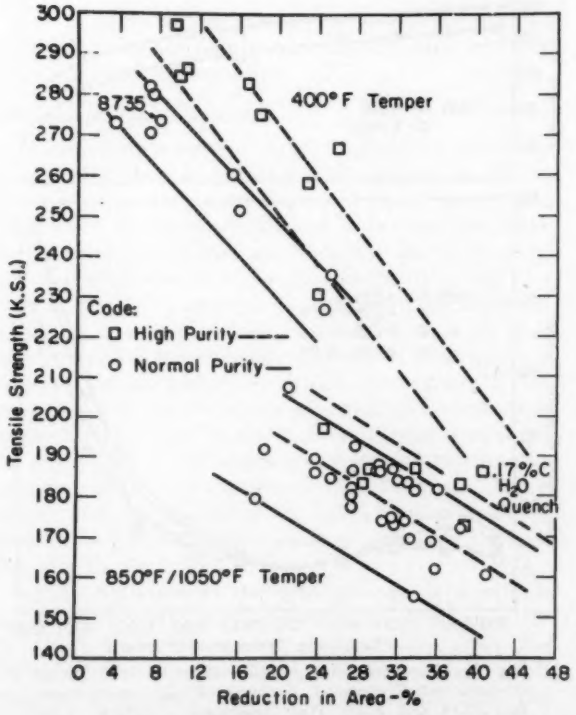



Fig. 3 — Purity effect on tensile properties of 4335. Heat treatment — 1650 F, 2 hr, oil quench; temper.

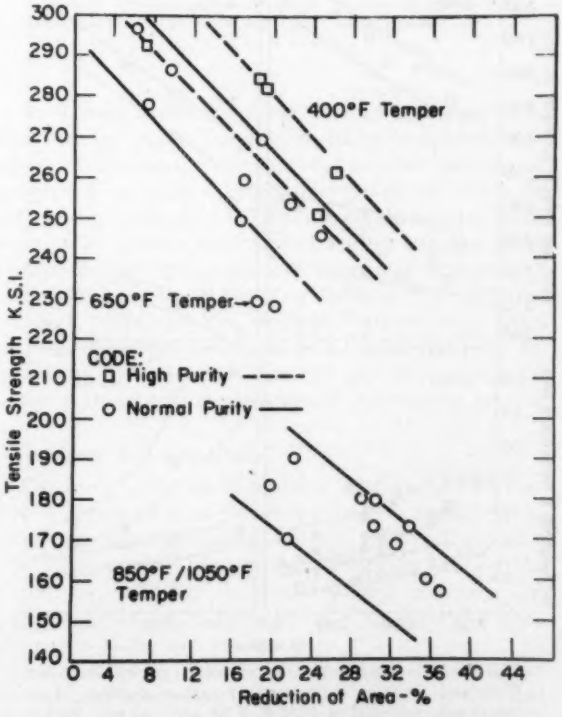



Fig. 4 — Purity effect on tensile properties of 4335. Heat treatment — 2280 F, 3 hr, air cool; 1650 F, 2 hr, oil quench; temper.

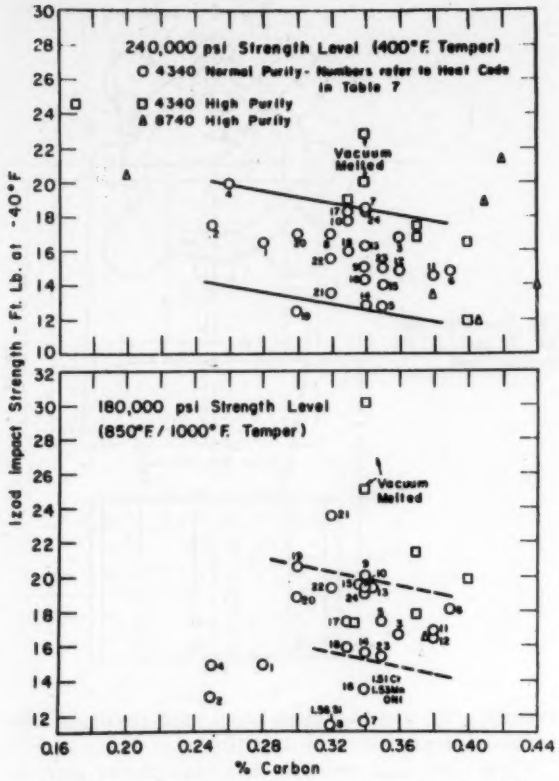



Fig. 5 — Variation of Izod impact strength at — 40 F, with carbon content.

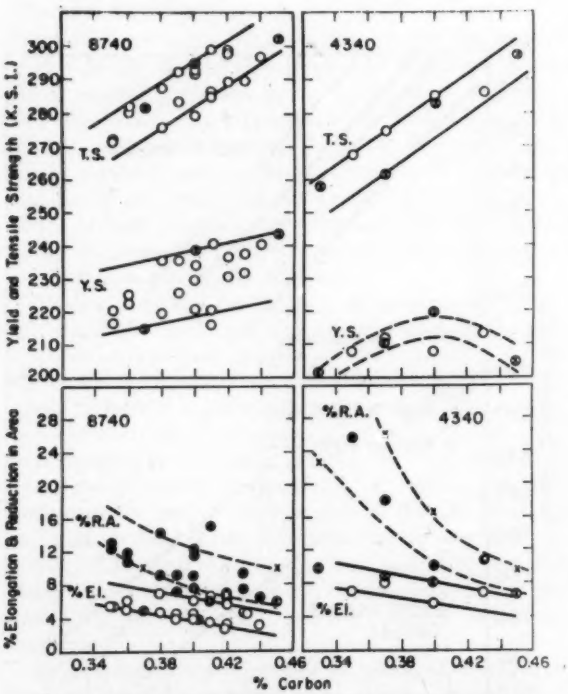



Fig. 6 — Tensile properties of high purity modified 8740 and 4340 as a function of carbon content. Heat treatment consisted of 1850 F, 3 hr, air cool and 1525 F to 1630 F, 2 hr, oil quench, followed by tempering at 400 F for 4 to 6 hr. X heats made with electrolytic iron, others made with sponge iron briquets.

ment. Knowledge of this effect is extremely valuable when one considers the technical difficulties which would be associated with heat treating large complex shapes at 2280 F (1249 C) if this were the only solution.

Figure 5 presents impact data showing again the beneficial effect of high purity, and in addition the marked effect of vacuum melting. The latter effect in particular represents a fruitful area for further work.

### *8700 Series Steel Advantages*

The advantages of using a leaner alloy, such as an 8700 series base as compared to the 4300 series, is clearly defined in Fig. 6, which compares these two alloys over a range of carbon content. The reduced retained austenite level in the 8700 series results in substantially improved yield strength in the range investigated. As a final example of the type of data required to give these cast alloys engineering significance, Fig. 7 illustrates the effect of tempering temperature on impact strength at both room temperature and at —40 F. It is apparent that for maximum impact resistance at any given strength level, carbon should be so specified as to avoid tempering in the range between about 500 to 800 F (260-427 C).

Additional data on both the low alloy and 5 per cent chromium steels have been developed on these two W.A.D.D. programs, and is available to design people through normal channels. The data presented here have been selected as examples of what is being done, and what can be done to improve the existing levels of knowledge relative to these cast alloys. It is

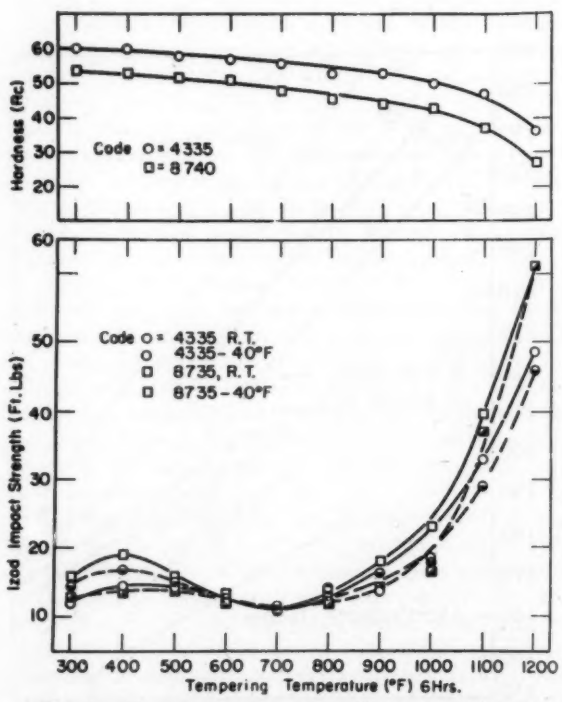



Fig. 7 — Tempering temperature effect on Izod impact strength and Rockwell C hardness of high purity modified 4335 and 8740. Heat treatment — 1850 F, 3 hr, air cool; 1575 F to 1600 F, 2 hr, oil quench. Temper as indicated.

essential that this type of investigation be continued if the cast alloys are to maintain a prominent position as engineering materials in critical applications.

### *Process Research*

The preceding has demonstrated the need for and benefits of a continuing program of physical metallurgy research in the field of high strength cast steels. It has in addition presented some tangible examples of the type of results which have been and must continue to be obtained if cast steels are to maintain status as engineering materials. This, however, is only part of the job. Unless the mechanical properties shown to be available in sound reproducible test bars can be translated to structural shapes of equivalent integrity, the whole matter of the physical metallurgy of cast materials becomes only an academic curiosity.

Process research then is that area of foundry inquiry which is intended to provide the information which will make possible such a translation from test bars to shape castings. Also included in process research are studies of those factors which influence surface finish and dimensional control, the ultimate aim being to provide an end product which can competitively perform a specific and desired function.

Specifically, then, what are the major defects which influence casting integrity and what can be done to study and control such defects? It is the belief of the authors that the four most serious deficiences encountered in the steel castings are:

1. Shrinkage of all types.
2. Hot tears.
3. Section size effects.
4. Unsatisfactory surface finish and dimensional control.

Each of these major defects is reviewed, and test methods and data pointing towards effective methods of control are demonstrated.

### *Shrinkage Type Defects*

Without attempting to either illustrate or define the many variations of shrinkage which can occur, depending upon the circumstances, it is sufficient to say that when shrinkage does occur something has interferred with directional solidification and feed metal is not available in the right place at the right time.

It is further important to note that the normally accepted criterion of x-ray or other forms of radiography are not necessarily capable of detecting the degradation of properties which can occur in the presence of a minor amount of shrinkage, particularly at the high strength levels. This is illustrated typically by Fig. 8, which correlates test bars and microradiographs taken along the length of a flat plate chilled at one end and risered at the other. The plates were perfectly sound by conventional x-rays taken through the one in. direction.

Figure 9 illustrates the arbitrary set of standards set up to rate microradiographs taken through 0.020 in. thick slices cut at various distances from the chilled end of the bar. Note the exceptionally good properties obtained adjacent to the chill, and the general degradation with increasing amounts of micro shrinkage.

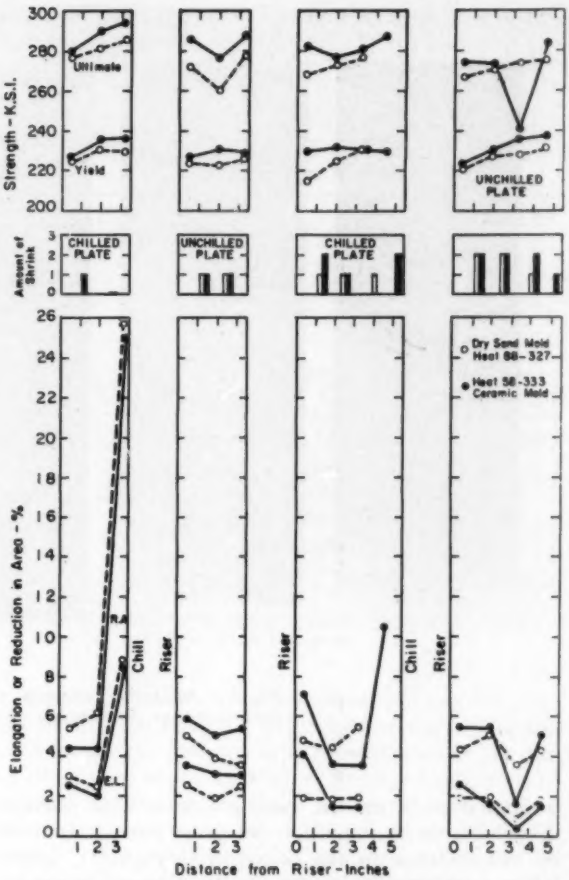



**Fig. 8 — Feeding distance effect in one in. thick cast steel plates. Test bars are 0.505 in. diameter.**

Similar studies have been performed and reported[2,3] for various plate lengths molding materials and alloys. Additional studies in this general area are in progress to provide a better understanding of the degree of taper required for completely directional solidification in various section sizes and in various alloys. Complimentary programs are investigating quantitatively the effects of molding materials of different thermal properties on feeding characteristics. Data such as the effects of taper, mold materials, and behavior of various alloys must be available if the designer is to take full advantage of the casting process.

### *Directional Solidification*

The basic point to be made is that relative to control or elimination of shrinkage there is absolutely no substitute for directional solidification. Directional solidification may be influenced by riser size and location, by padding, chills and gating methods, but it must be present if the casting is to yield the same properties as the well fed test bar.

Some of the answers which issue from these studies are not easy to swallow. The effect upon casting yield and cleaning time can sometimes be phenomenal. The end result is a cost picture which is sometimes somewhat staggering on a per pound basis if it is

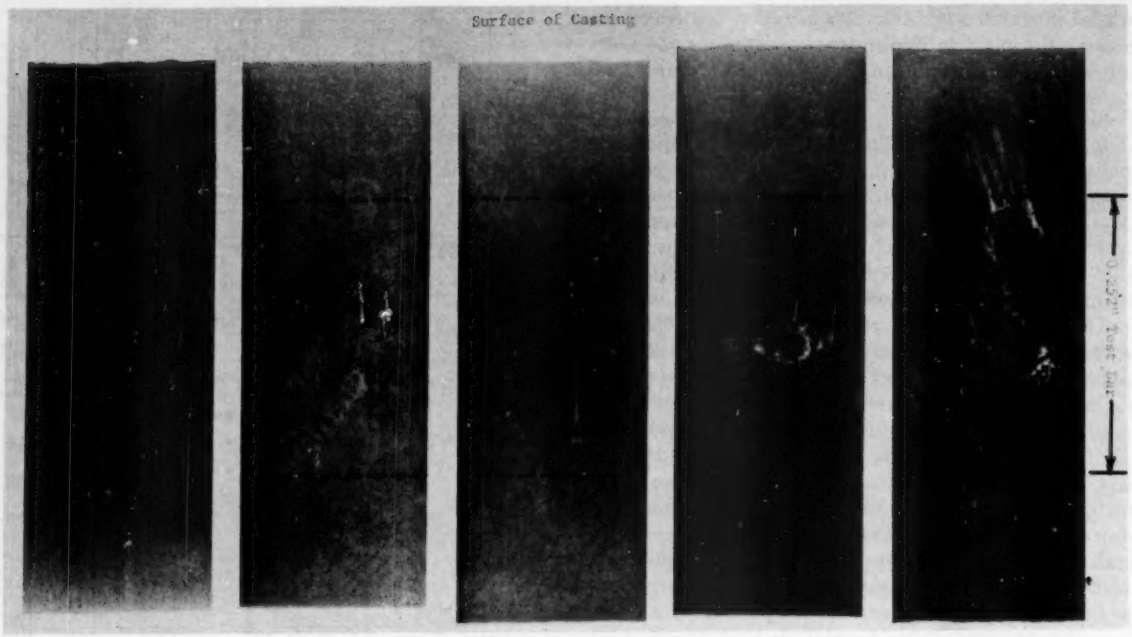



**Fig. 9 — Arbitrary reference radiograph standards for evaluating microshrinkage. Radiograph of 0.20 in. slice from ½-in. plate. 10×.**

compared with normal casting prices. The essence of the matter is that high integrity castings cannot be looked upon in the conventional manner. High integrity castings represent one solution to obtaining a given part to do a given job reliably and reproducibly. Castings are not to be arbitrarily desired any more than a forging or a weldment should be arbitrarily selected. The two criteria are:

1) Will it do the job?
2) How much does it cost compared to an alternate method?

The answers to these two questions must provide the basis for selection. No reputable foundry wants to make a casting that will not do the job or that will not make money. The big challenge to those foundries which do want to participate in this high integrity program is to find, or even more important, to help create those designs which will stand the low foundry yield, the high cleaning and inspection expense and still be the cheapest way to accomplish the desired function. This may seem somewhat remote from feeding problems, but the two are irrevocably intertwined.

It is the responsibility of the foundry industry to generate these fundamental data so necessary to proper casting design. It is also the responsibility of the designer to listen to the qualified foundryman and believe him when he says this or that can or cannot be done; or at least to believe him when he says that can only be done with difficulty and lots of money.

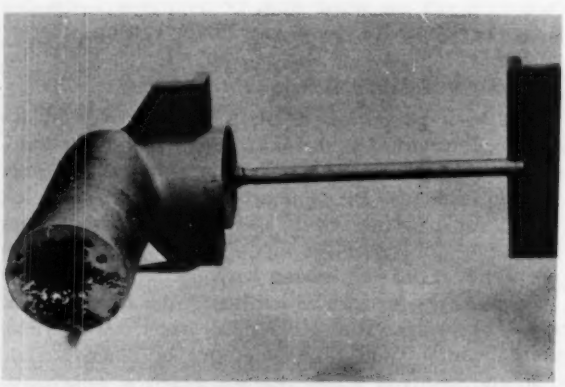


**Fig. 10 — Hot tear test specimen.**

### *Hot Tears*

The hot tear is perhaps an even more insidious and harmful defect in steel castings than shrinkage. Tears most generally occur on the surface at re-entrant angles or sharp corners where stresses are already high. The additional notching caused by the presence of a hot tear then serves to compound the problem, particularly when notch sensitive materials are involved.

Detailed studies of hot tear formation have been made and thoroughly discussed by Pellini,[4,5] Christopher[6] and others, and will not be repeated here. What will be discussed is one type of specimen for measuring hot tearing tendencies, and data showing the effect of several variables upon the hot tearing.

The test bar used for the work reported herein is shown in Fig. 10. This test bar consists of a ½-in. diameter rod with a restraining cross piece at one end. At the opposite end the casting becomes 3 in. in di-

ameter and is well risered. The fillet connecting the 3 in. section to the ½-in. section is a loose piece on the pattern, and may be varied from an ⅛-in. radius to a one in. radius. This test casting was poured under a series of different conditions involving changes in chemistry, melting practice and mold materials to determine what the effect of these variables might be upon hot tearing.

It was not anticipated that the data obtained could necessarily be utilized on a quantitative basis, as overall casting geometry itself is certainly one of the factors involved. It was felt, however, that the trends observed would be important, and that by incorporating the more favorable ones in casting design and molding procedures that benefits would be obtained.

### *Hot Tear Elimination Factors*

One of the most potent factors involved in hot tear elimination, as revealed by this particular test procedure, was the effect of using high purity melting materials so that phosphorus and sulfur were maintained at 0.010 per cent or lower. Figure 11 shows the summary of a series of tests made in investigating this particular variable. With the particular molding material and fillet radius involved, no tearing occurred with the high purity material, while some tearing occurred at all pouring temperatures involved with the normal purity material.

| | 1/2 INCH FILLET RADIUS | | | | |
|---|---|---|---|---|---|
| POUR TEMPERATURE | HI PURITY 4330 | | STANDARD 4330 | | |
| AIM 3100° F | OK | OK | OK | OK | OK |
| RANGE 3120-3060° F | OK | OK | OK | O | X |
| AIM 2975° F | OK | OK | OK | OK | OK |
| RANGE 2980-2950° F | OK | OK | O | O | O |
| AIM 2850° F | OK | OK | X | X | X |
| RANGE 2870-2815° F | OK | OK | X | X | X |

Fig. 11 — High purity effect on hot tearing of cast 4330 steel.

The severity of tearing increased as pouring temperatures decreased. Based on these results, high purity melting stock is used exclusively for all high integrity castings produced at the authors' facility.

A second factor investigated was the effect of cold versus hot molds. This study included, as an additional variable, the use of refractory grains of two distinctly different thermal characteristics. These were calcined kyanite and alumina, ($Al_2O_3 \cdot SiO_2$) and ($Al_2O_3$), respectively.

Table 2 shows the freezing time of a 3 in. diameter cylinder 12 in. long using alumina as the base refractory in one instance and calcined kyanite in the other. In addition, preheated molds were poured using each of these refractories. The time to freeze is nearly twice as long in the room temperature kyanite mold as in the room temperature alumina. Now observe the hot tearing results presented in Fig. 12.

TABLE 2 — FREEZING TIME OF 3 IN. DIAMETER CYLINDER USING VARIOUS REFRACTORY MOLD MATERIALS

| Refractory Grain | Mold Condition | Time To Freeze, Sec | Relative Times |
|---|---|---|---|
| Alumina | RT | 245 | 0.57 |
| Kyanite | RT | 435 | 1.00 |
| Alumina | 2200 F | 460 | 1.05 |
| Kyanite | 2200 F | 765 | 1.76 |

| HOT MOLDS | | | | COLD MOLDS | | | |
|---|---|---|---|---|---|---|---|
| KYANITE | | ALUMINA | | KYANITE | | ALUMINA | |
| OK | OK | OK | OK | OK | OK | OK | OK |
| OK | OK | OK | OK | OK | OK | OK | O |
| OK | | OK | OK | OK | OK | O | O |
| | | | | OK | OK | X | X |
| | | | | O | O | X | |

Fig. 12 — Mold temperature and refractory grain effect upon hot tearing of 4330 cast steel.

When the molds were preheated and poured hot, no tearing was observed with either refractory. With cold molds, the alumina type showed tears in six out of nine cases, while the kyanite produced minor tears in only two out of ten cases. All of the tests in this illustration were poured in a range between 2890 and 3000 F (1588-1649 C). In all cases the molds are of an extremely rigid nature, yet the thermal properties of the kyanite prevent or at least minimize the tearing as compared to the more chilling alumina grain. Mold restraint therefore is not necessarily the major factor influencing hot tearing.

Preheating reduces the cooling rate in both types of molds and again reduces tearing. Table 2 indicates that a preheated alumina mold simulates the rate of a cold kyanite mold which checks with the observed hot tearing tendency.

### *Conventional Mold Materials*

More conventional mold materials can also be evaluated by such techniques, as shown in Fig. 13, wherein the relative tearing tendency is measured for three commonly used types of sands.

| POURING TEMPERATURE | SILICA BASE DRY SAND | | ZIRCON BASE DRY SAND | | SILICA BASE GREEN SAND | |
|---|---|---|---|---|---|---|
| AIM: 3100° F. RANGE: 3090-3040° F. | O | O | X | X | OK | OK |
| AIM: 2975° F. RANGE: 2970-2930° F. | O | O | X | X | OK | OK |
| AIM: 2850° F. RANGE: 2860-2840° F. | O | X | X | X | OK | OK |

Fig. 13 — Hot tearing tendencies in several commonly used sand mixes.

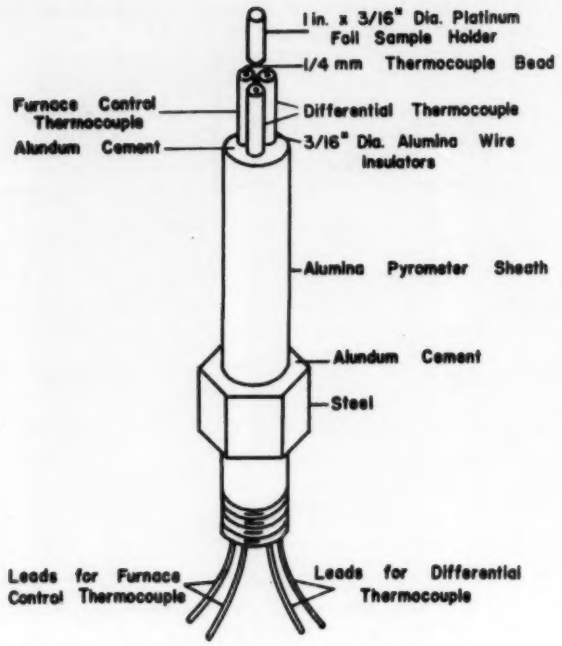



Fig. 14 — Schematic diagram of differential thermal analysis unit less the heating furnace.

Additional studies covering deoxidation effects, carbon level effects and melting practice variables have been performed. The data will not be reported herein, as it is the basic aim of this paper to discuss approaches and methods rather than to present a mass of data which may be pertinent to only the specific operation involved. The main point is that by the use of a relatively simple method a great number of the variables which influence hot tearing can be studied on a quantitative or at least semi-quantitative basis and the trends determined. The application of these trends to planning the production of a specific part can in many cases cut weeks off the time required to produce acceptable components.

### *Section Size Variables*

Most steel casting specifications call out properties based on a keel block or some similar test bar which normally has about a one in. section. A major question then arises, when the casting section exceeds this one in. by any substantial amount, as to how the casting properties compare with the test bar properties.

The answer to this question is often complicated by the fact the heavier sections are harder to feed and harder to evaluate radiographically so that soundness itself becomes a questionable variable. It it is assumed for the moment that by sectioning and by microradiography it has been determined that the section is sound, then certain generalizations can be made based on work performed under a recent Air Force contract.[3] These are:

1) Satisfactory properties can be obtained in up to 2 in. sections at the 240 ksi ultimate strength levels using low alloy steels of the 8700 and 4300 type.
2) Above a 2 in. section, properties at the 240,000 psi level are impaired, particularly with respect to ductility, and additional work is required to reach reasonable combinations of engineering properties.
3) At the 180,000 psi ultimate strength level, it is possible to maintain reasonable (6 per cent) elongation in sections up to 3 in.
4) Cast 5 per cent chromium steels of the hot work die type require a substantially increased research effort, if properties useful for structural applications requiring ductility are to be guaranteed in sections much over one in.

In summary, therefore, it must be realized that these section size problems may exist, and that careful consultation with the proposed foundry vendors is essential if sections in excess of several inches are present in proposed casting designs which require high mechanical properties throughout.

### *Surface Finish and Dimensions*

Surface finish and dimensional control are added factors which in various ways are related to the integrity of the final product. Depending upon end use, the stringency required in these two areas can vary substantially. In any event, it is the aim of the competent foundry to supply the best combination consistent with service and cost requirements.

First, relative to surface finish, some consideration must be given to methods of designation. For general use and versatility, the authors prefer a surface comparator gage as being the most practical if combined with some additional specification which defines size, location, etc., of discrete defects such as trapped gas or nonmetallic inclusions.

Surface finish deterioration can be attributed to four general sources:

1) Mechanical entrapment of slag, dirt, etc., from the ladle during the pouring operation.
2) Mold-metal reaction or mechanical erosion of the gating system during the pouring operation, which is then carried into the mold cavity.
3) Mold-metal reaction within the mold cavity itself due either to organics which are used as binders, or in the case of ceramic type molds, to inadequate firing to remove volatiles.
4) In ceramic molds which have been adequately fired, mold-metal reaction can still occur if an improper combination of binder and refractory aggregate is selected relative to the pouring temperature and section sizes involved.

Sources (1) and (2) normally produce discrete defects which can be minimized or eliminated by careful metal handling procedures, properly designed gating and by developing sufficient strength in the mold material to resist mechanical erosion. Sources (3) and (4) are more subtle, but are also amenable to certain laboratory procedures in evaluating adequate foundry procedures.

### *Thermal Analysis*

One such useful procedure is differential thermal analysis, which has been described in detail elsewhere,[7] but which will be reviewed briefly herein.

Figure 14 shows schematically the type of apparatus involved. The unknown specimen such as a potential binder material is placed in the furnace along with a known inert material, and each has inserted in it a thermocouple. These thermocouples are bucked and the specimens heated at a slow and uniform rate.

If no reaction or phase change is occurring in the unknown, the net result is a zero emf. If a reaction occurs in the unknown, it must be either exothermic or endothermic in nature, and a resultant emf either positive or negative, respectively, occurs. This procedure then allows the analyst to follow the effects of heating of refractory combinations and determine, for example, how high a firing temperature is required to achieve complete stability.

Figure 15 shows a specific example. This is the differential thermal analysis curve for a calcium aluminate type binder used both for molds, cores and as a furnace refractory. The thermogram of a sample previously dried at 212 F (100 C) indicates that in all probability the dehydration of two compounds is required to completely cure the mold or core. The first compound to dehydrate is the major component ($CaO \cdot Al_2O_3 \cdot Aq$) which begins to lose water about 440 F (227 C), and which is completely decomposed at about 900 F (482 C). The evolution of water from a second constituent which is probably a di-aluminate starts at about 900 F (482 C) and ends at 1100 F (593 C). There is no evidence of further reactions.

From this evidence it is possible to conclude that a baking temperature exceeding 1300 F (704 C) should be adequate to prevent defects due to mold gases. Field use has confirmed the recommendations which can be derived from this differential thermal analysis data of calcium aluminate as well as other binder systems. Thermogravimetric studies, that is, measurement of weight losses as a function of increasing temperature, provide similarly useful data for evaluating new binder and refractory combinations. In this test the specimen to be evaluated is placed on a balance which is so constructed that the scale pan is in the heating furnace.

The temperature of the furnace can be raised at a controlled rate, and the weight change with temperature plotted directly. Figure 16 shows a typical result wherein a sample of western bentonite used in some wash studies was subjected to such a thermogravimetric analysis. Some weight loss, probably absorbed water, occurs gradually up to about 1200 F (649 C) at which the rate increases sharply. This increase in rate is caused by the loss of the more tightly bound chemical water. Complete dehydration is reached at about 1500 F (816 C). A total weight loss of about 6 per cent occurs.

### *Thermogravimetric and Thermal Analysis*

The combination of the thermogravimetric and differential thermal analysis results provides an extremely useful means of studying refractory-binder systems, as the former provides weight change information only, while the latter will detect phase changes even where no change in weight is involved. As indicated earlier, the differential thermal studies also reveal whether such changes are exothermic or endothermic.

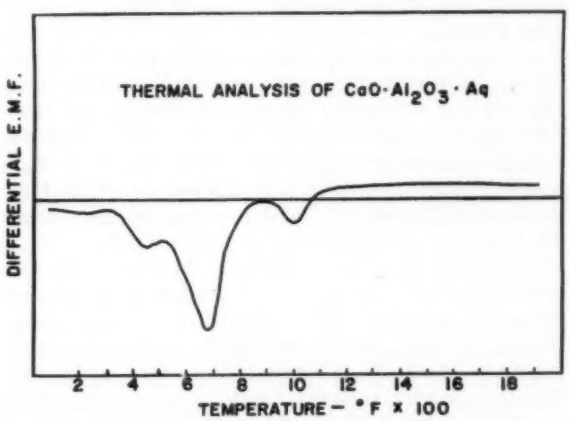



Fig. 15 — Differential thermal analysis curve for curing $CaO \cdot Al_2O_3$ binder.

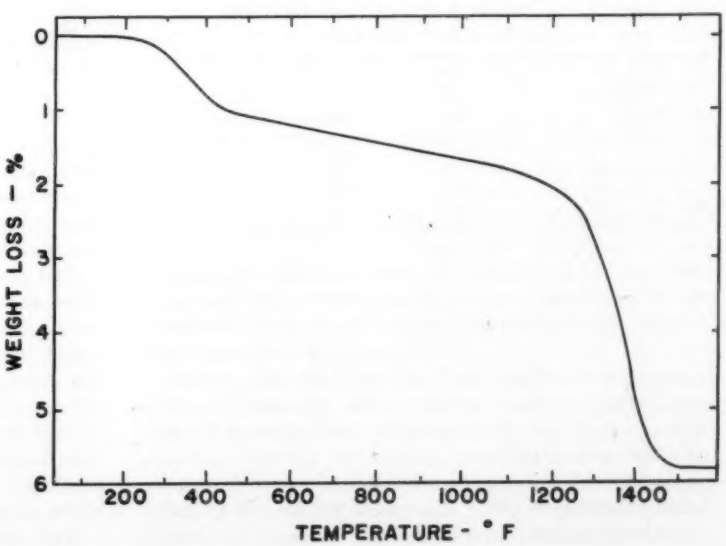



Fig. 16 — Temperature-weight loss curve for Western bentonite obtained by thermogravimetric analysis.

The aim again of such tests, in addition to the more conventional types of measurement, is to provide an increasing degree of knowledge about those materials which are put together to make a mold, and as a result to learn to prevent or control the reactions which may be detrimental to surface quality.

Relative to casting dimensions, there appears at the moment to be no substitute for experience, good records and the willingness to consider prototype patterns where the situation demands the ultimate in tolerances.

In summary, therefore, this initial section has attempted to define some of the various types of research programs essential to a continued high degree of progress in the field of high strength steel castings. A distinction has been made between physical metallurgy and process research both of which provide ample opportunity for forward steps in the state of the art.

## CASE HISTORIES

Having discussed various phases of a development program for steel castings, the authors will next examine how such a program may be applied to the manufacture of technically advanced, high integrity, steel castings.

One of the first major decisions involved is the selection of the proper alloy. Relative advantages of the two major types of alloys have been noted. The 4300 type is favored at the 180,000 tensile level. At tensile strength in excess of 200,000 the 8700 type may be more advantageous particularly if a high yield-to-tensile ratio is necessary and if the casting has generally light sections (less than one in.).

Experience has shown that the best molding material is a slurry type ceramic if the maximum quality surface is required. The inherent insulating characteristic of this ceramic minimizes the hot tearing tendency of the high strength low alloy steels. Preheating of the ceramic mold is also a standard practice. Table 3 shows the effect on scrap rate of preheated versus nonpreheated molds for a small

TABLE 3 — MOLD PREHEATING EFFECT ON SCRAP RATE

| | Alloy 4330 steel castings (approx. 5 lb each) | | | |
|---|---|---|---|---|
| Part | No. Cast in Cold Molds | Scrap in Cold Molds, % | No. Cast in Hot Molds | Scrap in Hot Molds, % |
| A | 84 | 61 | 20 | 5 |
| B | 32 | 37 | 8 | 0 |
| C | 98 | 62 | 38 | 21 |
| D | 87 | 56 | 24 | 4 |

4330 steel alloy casting. The most common practice followed in the authors' company's experimental foundry involves simply preheating by gas burners down the sprues and thence through the runner and gating system. This tends to establish the proper temperature gradient in the mold, especially if the rigging system has been designed to make maximum use of hot metal risers.

Considerable importance is assigned to preliminary design planning with the customer. This is the foundryman's opportunity to inform the customer of those features which will most certainly assure the production of a successful casting. At this point the concept of proper taper to gain a high degree of soundness should be discussed. Also other such considerations as draft angles, minimum wall sections, fillets and any provisions that will simplify machining operations should be made.

Large high integrity steel castings require considerable time and effort in the matter of pattern planning. The large mahogany patterns which are involved must be constructed on exceptionally sturdy boards with boxed type construction. Gating and risering must all be planned in advance and placed on the pattern. With slurry ceramic molding it is not possible to cut gates and risers by hand. Patternmakers shrinkage must be carefully determined. Sometimes it is necessary to build a prototype pattern almost exactly like the finished product in order to predict accurately how much the casting will shrink.

### *Inspection Standards*

Prior to establishing the cost of such castings, a definite agreement must be reached with the customer on standards of inspection. It must be clearly understood just how much is expected in the way of radiographic soundness. As this paper is written, the steel casting standard most often used seems to be A.S.T.M. B71-52, although this leaves much to be desired. Methods and standards for either magnetic particle inspection or penetrant type inspection must also be agreed upon.

Several specific examples are given in the remainder of this paper. They are interesting and unusual, inasmuch as they combine the highest considerations of soundness, dimensional accuracy and general complexity for any high strength, high integrity steel castings of their size ever made. These examples are all part of the Transporter-Erector-Launcher (T.E.L.) for a missile.

The prototype T.E.L., which has been used in all of the research and development test firings of the surface-to-surface missile, is designed so that it can be easily transported by helicopter or cargo transport, as well as on the XM-474 tracked vehicle. The T.E.L. gives the missile a "shoot and scoot" capability that makes it possible for the missile to be transported to a site, erected and fired in a matter of minutes. The critical weight, center of gravity limitations and high strength quality parts and requirements imposed on the T.E.L. design showed that many potential advantages could be obtained by selecting the casting process as a method of fabricating complex structural components of the launcher.

The authors' company was contracted to produce these parts as high purity modified 4330 high strength steel castings. At the date of this writing approximately six of each have been made and shipped to the customer. They have certain features in common. All are cast of high purity 4330 alloy. All are heat treated to a tensile strength level of 180,000 psi with a minimum required yield strength of 160,000 psi and a minimum elongation of 8 per cent (these properties to be attained in test bars attached to the castings).

The dimensional tolerance requirements are ap-

proximately the same. The castings are roughly 40 in. in diameter with a tolerance on that diameter of $\pm 1/16$-in. All of these parts are cast in slurry ceramic molds, in large heavy duty steel flasks purchased especially for this program.

All five parts are made to the same general radiographic requirements based upon A.S.T.M. specification E71-52. Table 4 lists the minimum standards for both Class I and Class II areas, a designation created by the customer and appearing on each of the five castings in question.

TABLE 4 — RADIOGRAPHIC STANDARDS OF ACCEPTANCE FOR LAUNCHER PAD, AZIMUTH RING, PIVOT RING AND MISSILE RINGS

| Plate No. Ref. ASTM E71-52 | Class I Areas | Class II Areas |
|---|---|---|
| | **Gas and Blow Holes** | |
| A1 | Acceptable | Acceptable |
| A2 | Borderline | Acceptable |
| A3 | Unacceptable | Borderline |
| | **Sand Spots and Inclusions** | |
| B1 | Borderline | Acceptable |
| B2 | Unacceptable | Borderline |
| B3 | Unacceptable | Unacceptable |
| | **Internal Shrinkage** | |
| C1 | Acceptable | Acceptable |
| C2 | Borderline | Acceptable |
| C3 | Unacceptable | Unacceptable |

## LAUNCH PAD CASTINGS (FIG. 17)

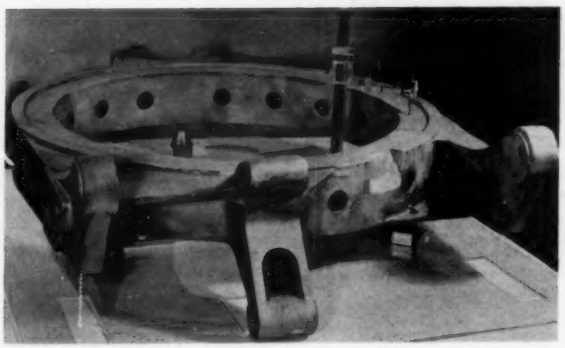


Fig. 17 — Launcher pad casting.

### *Design*

Two basic design considerations were employed without which it would not have been possible to cast this part. The first is the principle of taper for maximum soundness. Figure 18 shows how the taper was designed into the internal wall surfaces of this part. Although this cross-section is typical, there are many other areas in the casting where the same principle was used, but they are simply too numerous to show in their entirety. The second principle used in designing this part was simply careful planning for easy machined removal of riser padding.

It can be seen in Fig. 18 that three of the four exterior surfaces of the rectangular cross-section had ample riser padding which had to be removed. Since this could be most easily accomplished in the final stages by a turning operation, it was desired to keep those three surfaces free of any intercepting surfaces which might interrupt a machining cut. In general the amount of taper applied was 3 degrees. The riser padding had a more generous taper than that, being a minimum of 5 degrees.

After the basic design was decided upon and drawings issued two models were made, a half scale one for use by the foundry, and a full size one for customer's use. The customer's model was helpful to him for planning purposes and in deciding on any minor design changes. The foundry's model, on the other hand, was a distinct aid in visualizing details of the necessary gating system. One test bar was placed on each of the two 4 ½-in. diameter pivot arm bosses visible in Fig. 17. The size of the test coupon was approximately 1 x 1 cross-section and 3 in. long.

### *Pattern Construction*

Figures 19 a and b show the mounted pattern. Easily visible in both views is the tapered padding arrangement necessary to achieve soundness in this part. Two separate gating systems are employed, one connected to the 13 inside diameter blind risers, and one connected to the three main arms which extend from the outside diameter. There was no time in this program for the construction of prototype pattern equipment. Therefore, it was necessary to take an "educated guess" as to the proper patternmakers shrinkage. An average shrinkage value was chosen based on measurements of thirty-five 36 in. diameter missile ring castings previously made in the experimental foundry.

### *Foundry Processing*

Considerable time is required to assure proper positioning of the cores that form the internal passages. It is necessary to gage the cores for both vertical and radial position with the use of dial gages and calipers. Each core is then securely wired in position through the drag. Figure 20 shows a drag section with cores gaged and fastened in place.

Initially, this casting was poured with two ladles simultaneously using the inside diameter runner system. When it was observed that the metal had reached the top level of the main ring of the casting one of the ladles was moved to the outside runner system, in order to introduce hot metal to the heavy bosses on the ends of the three main arms. The other ladle kept pouring into the inside diameter system while this change was being affected. The practice was altered after a few castings were poured because it was found somewhat unwieldy to make the ladle movement during pouring, and because somewhat more consistent x-ray results were obtained by the altered pouring practice.

Now one ladle is used to pour the inner gating system, and pouring from the second ladle into the outer system is started when the level of liquid metal has reached the top portion of the main ring of the casting. Total pouring time is about 25 sec. Pouring temperature is approximately 2950 to 3000 F (1621-1649 C). Upon the completion of pouring with the

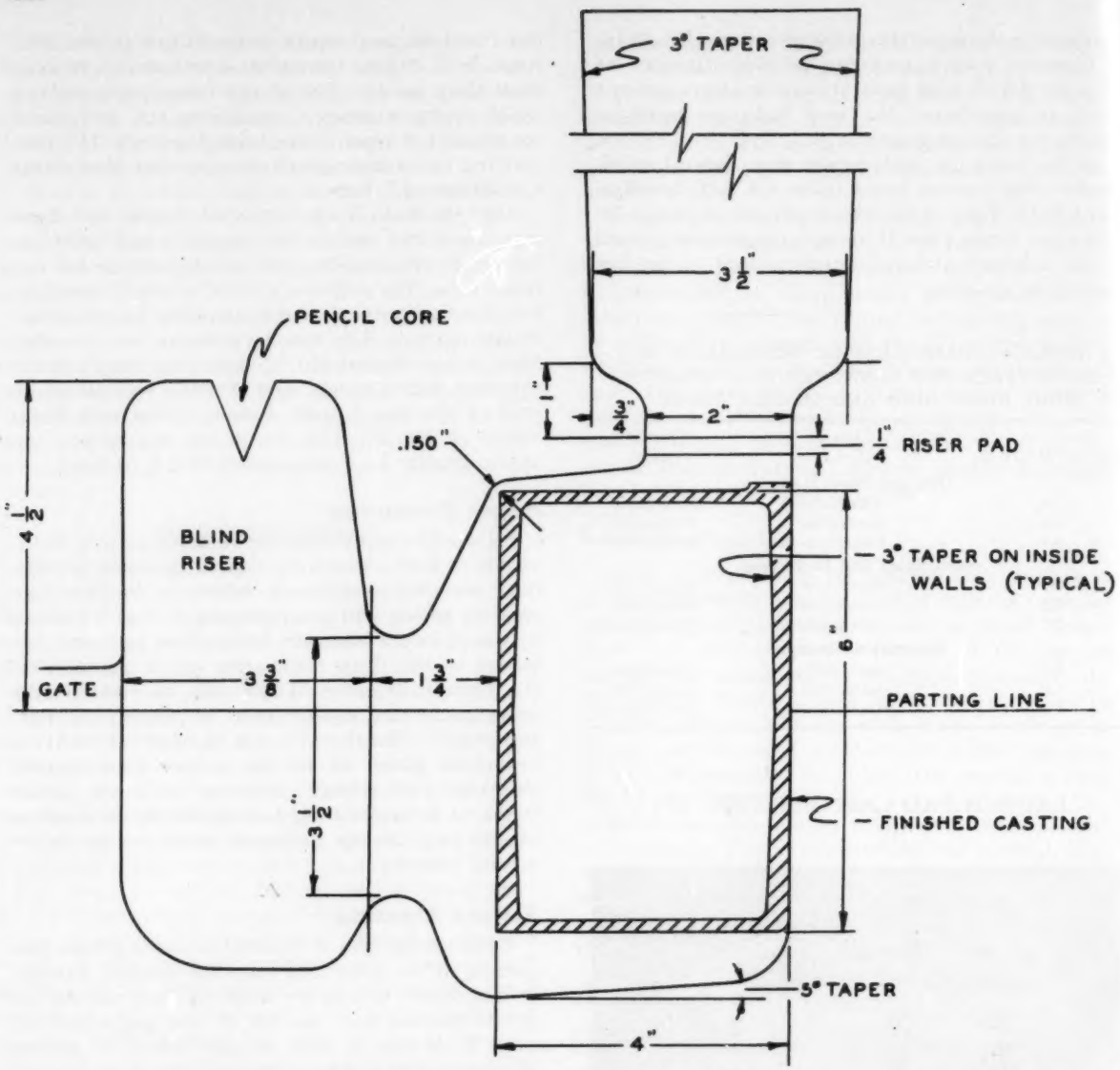



Fig. 18 — Launcher pad casting typical cross-section showing risers and padding.

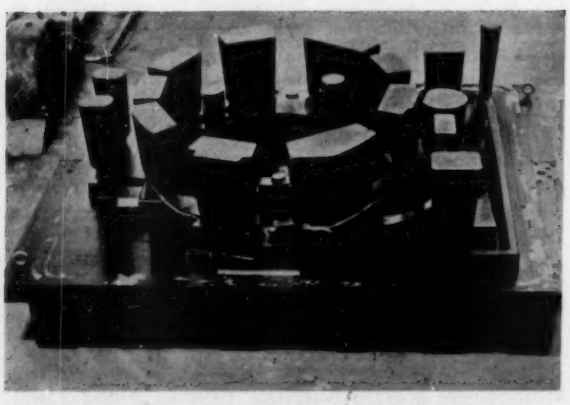


Fig. 19a — Launcher pad pattern equipment (cope).

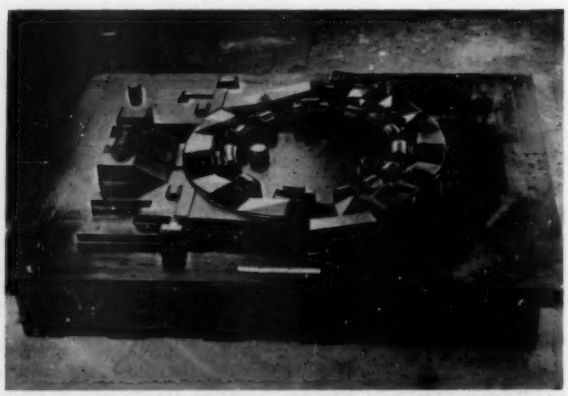


Fig. 19b — Launcher pad pattern equipment (drag).

metal approximately 1½-in. from the top of the mold, each riser is covered with an exothermic compound of the thermite type to help insure proper temperature gradients in the risers.

Removal of the inside diameter risers is facilitated by means of oxyacetylene flame cutting equipment. In general it is necessary to leave about ¼-in. adjacent to the flame cut surface to be ground more closely to size. This stock allows for any cracks and irregularities resulting from the flame cutting operation.

Prior to machining the casting is annealed for 2 hr at 1300 F (704 C) and furnace cooled. The casting is then rough machined leaving approximately ⅛-in. on each of the three main surfaces of the cross-section to be removed by subsequent machining operations.

### *Inspection*

A sonic type gage is used to measure the wall thicknesses in the basic cross-section. This procedure coupled with the careful setting of cores and recording of their position are the key to proper dimensional control of this casting. Wall thicknesses have to be held to a tolerance of ± 0.032 in. Part of this is achieved by proper initial core positioning, but it is also necessary to conduct the final machining only after it has been absolutely determined where the core passages are located.

After the casting has been returned from final machining, it is radiographed using Iridium 192 and Cobalt 60 sources. A total of 87 exposures are needed to give an essentially complete coverage of this complex part. The casting is then checked using both magnetic particle and dye penetrant methods. Finally, it is subjected to dimensional layout using the template supplied by the customer.

### *Weld Repair*

Permission was obtained from the customer to perform minor weld repairs. Major weld repairs are performed only with customer's specific approval. An argon shielded tungsten arc is used. The weld metal is a ⅛-in. diameter high purity 4330 rod, which was cast as a billet in the foundry and extruded for us to the desired size. After welding all areas are re-x-rayed and penetrant inspected.

### *Heat Treatment*

Heat treatment is—1850 F (1010 C) —3 hr in cracked ammonia atmosphere and air cooled. 1600 F (871 C)—2 hr in cracked ammonia atmosphere and oil quenched; 875 F (468 C)—2 hr air cooled. The heat treating vendor is responsible for necessary fixture arrangements to prevent warpage or to correct warped or out-of-round conditions incurred in the casting process.

### *Results*

X-ray results have generally confirmed the conclusion that 3 degrees of taper is sufficient to feed for a distance of approximately 3 in. within the requirements of A.S.T.M. specification E71-52, Grade C1. In a few areas which are complicated by the geometry of adjacent bosses, etc., a Grade C2 shrinkage may be observed. In some cases these have been eliminated by changes in gating and risering. In others it has been necessary to add more taper and remove it by extra grinding operations.

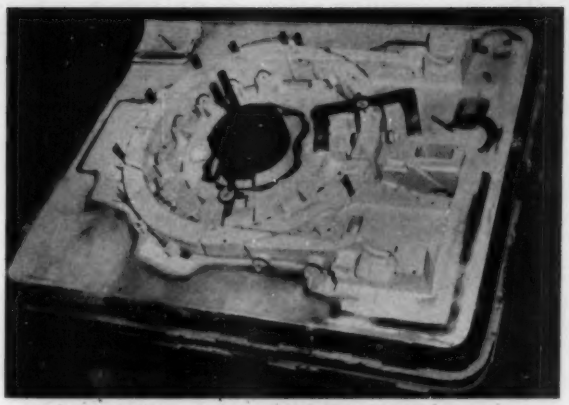


**Fig. 20 — Drag of mold for launch pad casting showing core setup.**

Magnetic and penetrant type inspection has confirmed the conclusion that the combination of sound metal cast in a ceramic mold produces freedom from hot tears. There have been an average of one to two welds necessary for this type defect in certain marginal areas on each casting.

Dimensioning of castings made to date has shown that the chosen shrink rule was quite reliable on the main diameter measurements of this casting. In fact these generally fall between the ±1/16-in. tolerance of the blueprint. A major discrepancy was immediately apparent, however, in the shrink rule applied to the dimensions governing the ends of the three main arms. Here it appeared that a different shrink rule was required. It was, therefore, necessary after the first part was made to return the pattern to the pattern shop and adjust the arm positions to compensate for this.

Test bar results on both separately cast and attached coupons for the five castings made and tested to date are shown in Table 5. It is readily apparent that all of them met the required properties with a considerable margin of safety.

**TABLE 5 — MECHANICAL PROPERTIES — LAUNCH PAD CASTINGS**

| Serial No. | Yield, psi 0.2% Offset | | Ultimate Tensile, psi | | Elongation, % | |
|---|---|---|---|---|---|---|
| | Separately Cast | Attached | Separately Cast | Attached | Separately Cast | Attached |
| 1 | 177,600 | 174,500 | 190,000 | 190,200 | 11.0 | 16 |
| | 171,600 | — | 188,500 | — | 11.0 | — |
| 2 | 186,600 | 179,000 | 196,000 | 195,000 | 9.5 | 11 |
| | 184,200 | 179,000 | 196,000 | 195,000 | 9.5 | 13 |
| 3 | 179,100 | 178,700 | 196,000 | 193,000 | 11.0 | 13 |
| | | 176,700 | | 192,200 | | 12 |
| 4 | 174,300 | 174,500 | 189,000 | 190,500 | 9.5 | 14 |
| | 176,400 | 172,800 | 189,000 | 189,800 | 9.5 | 12 |
| 5 | 181,920 | 187,600 | 194,600 | 198,300 | 10.0 | 14 |
| | 180,480 | 181,200 | 193,200 | 195,500 | 11.0 | 13 |

NOTE: Separately cast test bars are machined to 2 in. gage length and 0.505 in. diameter. Attached test bars are machined to one in. gage length and 0.252 in. diameter.

The gross pouring weight of the launch pad is 2600 lb. It is shipped to the customer weighing 435 lb for a casting yield of 16.7 per cent.

## PIVOT RING CASTING (FIG. 21)

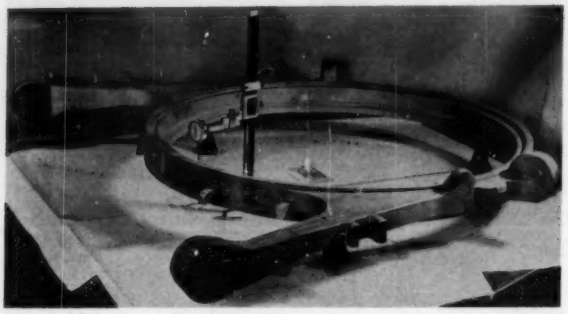


Fig. 21 — Pivot ring casting.

### *Design*

Because of the relatively simple geometry of the pivot ring, no major design problems were encountered from the foundry point of view. No modifications were proposed to the original design. Test bar locations were selected by the foundry with customer's agreement. One is located on a face at the ends of each of the two pivot arms, and one each on the drag surface of the arms half way between the end and the ring section (Fig. 22). Each is adequately fed by adjacent risers on the casting itself.

### *Pattern Construction*

Figures 22a and b show the pattern for the pivot ring casting. All but two of the 18 risers are hot metal risers with a gate at the base of each one. As in the case of the launch pad casting, the pouring is designed for two-up practice.

### *Foundry Processing*

No serious foundry problems were presented by this part. It was necessary to modify only two risers in order to achieve casting soundness easily greater than that required by the customer's standards. This modification amounted to the addition of padding between the two risers on each of the pivot arms. Prior to that a small amount of shrinkage existed in the area which was deemed borderline, and in two cases required weld repair.

### *Inspection*

Immediately after cleaning the casting is radiographed using Cobalt 60. A total of 32 films are required to give adequate coverage to this casting. Magnetic particle and dye penetrant inspection methods have been used to assure freedom from surface imperfections. Dimensional inspection is accomplished with the use of templates similar to those described previously.

### *Results*

Radiographic quality generally meets, without difficulty, all requirements in both Class I and Class II areas. A borderline shrinkage condition equal to about C2 in the A.S.T.M. specification did exist occasionally in the pivot arms between the two risers until tapered padding was added in this area to correct the condition. Penetrant and magnetic particle inspection have shown no problem with hot tears or cracks. Text bar results for both separately cast and attached coupons are shown in Table 6. Eighteen hundred lb of metal are required to pour the pivot ring casting and the shipping weight is 392 lb. The yield, therefore, is 21.8 per cent.

## AZIMUTH RING CASTING (FIG. 23)

### *Design*

Test bars were located at 90 degree angles on the drag surface of the ring portion of the casting. Aside from that no other design suggestions were made to the customer.

### *Pattern Construction*

Figures 24a and b show this pattern equipment. The inside diameter wall of the finished casting has a 30 degree taper, as can be seen on the casting in Fig. 23. It was decided from the start, however, that the inside diameter wall would be cast straight (ie., at 90 degrees to the plane of the casting). This arrangement would aid in feeding from the cope risers and would simplify pattern construction. The critical

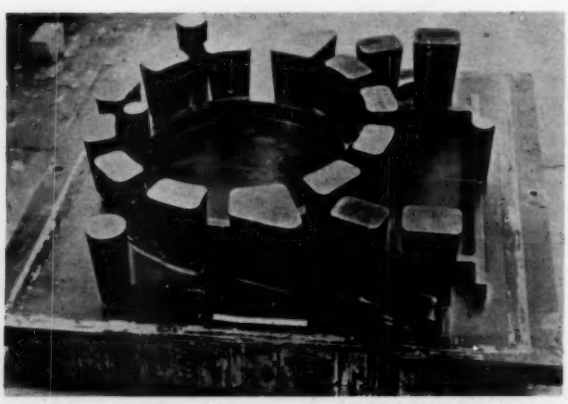


Fig. 22a — Pivot ring pattern equipment (cope).

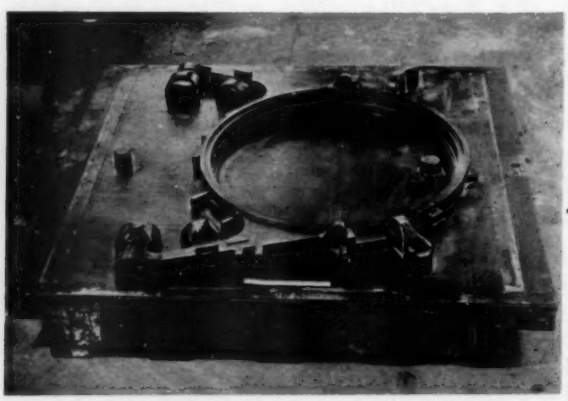


Fig. 22b — Pivot ring pattern equipment (drag).

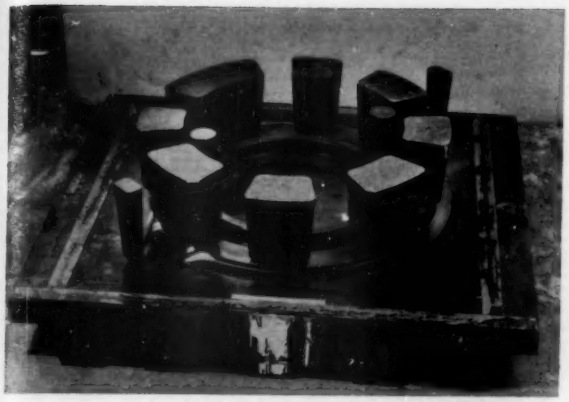


Fig. 24b — Azimuth ring pattern equipment (drag).

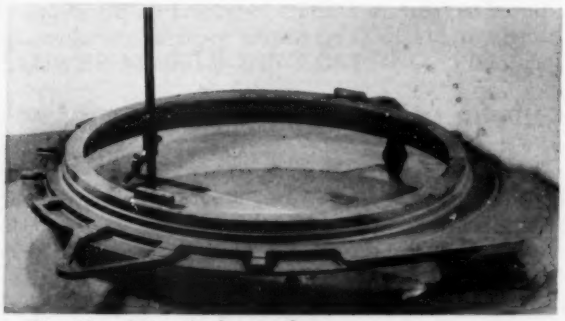


Fig. 23 — Azimuth ring casting.

area of this casting is the flat visor-like part which protrudes from one side. It seemed, therefore necessary to add taper in this area towards the nearest riser to insure soundness. This taper can also be seen in the cope of the pattern (Fig. 24a).

The pattern is gated on the inside diameter from a circular runner which is again designed for two sprues and a two-up pouring practice. Each of the eight gates enters the inside diameter immediately under a riser. Two other risers exist, one each on the thin lugs which protrude from points approximately 90 degrees around the circumference on either side from the visor.

### *Foundry Processing*

The casting is poured from two ladles simultaneously in about 12 sec. No unusual problems in foundry handling were encountered in the initial production of this part.

### *Inspection*

Radiography is performed with x-ray and Cobalt 60. X-ray is required for the 1/4-in. thick outer flange

Fig. 24a — Azimuth ring pattern equipment (cope).

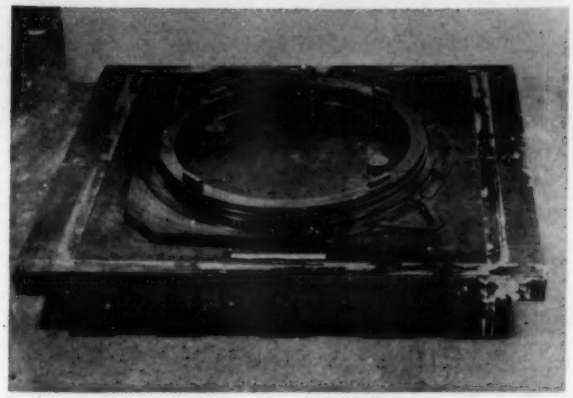


TABLE 6 — MECHANICAL PROPERTIES — PIVOT RING CASTINGS

| Serial No. | Yield, psi 0.2% Offset | | Ultimate Tensile, psi | | Elongation, % | |
|---|---|---|---|---|---|---|
| | Separately Cast | Attached | Separately Cast | Attached | Separately Cast | Attached |
| 1 | | 178,000 | | 192,600 | | 11 |
| | | 178,200 | | 193,000 | | 12 |
| 2 | | 174,800 | | 188,200 | | 16 |
| | | 173,800 | | 186,500 | | 17 |
| 3 | | 177,000 | | 190,000 | | 13 |
| | | 176,200 | | 188,400 | | 10.5 |
| 4 | 181,500 | 178,000 | 196,000 | 195,200 | 11 | 13 |
| | 183,600 | | 197,000 | | 11 | |
| 5 | 183,300 | | 200,000 | | 10 | |
| | 184,800 | | 200,000 | | 10 | |
| 6 | | 174,500 | | 187,500 | | 9 |
| | | 174,300 | | 188,400 | | 11 |
| 7 | | 180,300 | | 199,200 | | 12 |
| | | 186,200 | | 200,000 | | 10 |

NOTE: Separately cast test bars are machined to 2 in. gage length and 0.505 in. diameter. Attached test bars are machined to 1 in. gage length and 0.252 in. diameter.

of the casting. Cobalt 60 takes care of the rest of the part. A total of 36 exposures are made to give complete coverage of the azimuth ring.

### *Results*

Although the first casting was usable and was ultimately sent to the customer, radiographic inspection revealed several areas of borderline quality when referred to the A.S.T.M. Standards for shrinkage and for gas and blowholes. The shrinkage trouble occurred in the V-shaped lifting lugs which protrude from the 1/4-in. thick outer flange. This condition was corrected by the addition of taper towards the nearest riser on

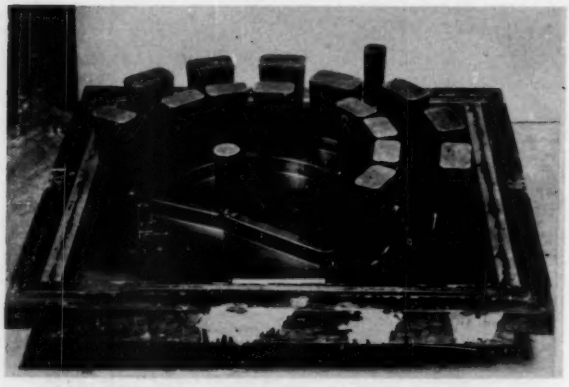


Fig. 26b — Upper missile ring pattern equipment (drag).

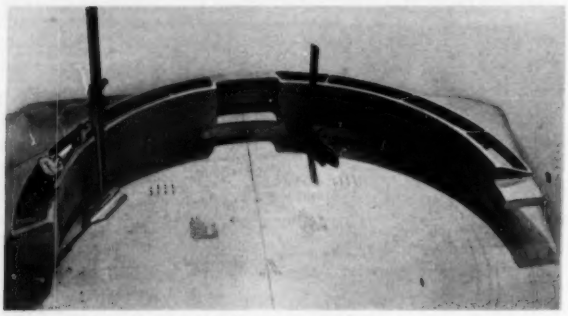


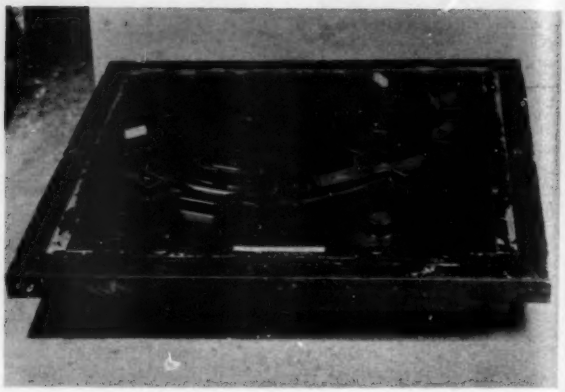
Fig. 25 — Upper missile ring casting.

the main body of the casting. Entrapped air was remedied in the outer ¼-in. thick flange when taper was added to it.

Approximately 5 degrees of taper were used, and as in the case of the lifting lugs the thickness was made to increase towards the inside diameter. Magnetic particle and dye penetrant inspection both revealed areas of trouble associated with the V-shaped lifting lugs, but this condition was also improved by the tapered stock which remedied the shrinkage. Some sponge type shrink also occurred in the ¼-in. outer flange, and it was completely corrected by the addition of taper.

Considerable dimensional difficulty was experienced initially with the part. Contraction in heat treatment in addition to expected pattern shrinkage caused the casting to be somewhat smaller than anticipated. A slight tendency for the part to go out-of-round then caused certain areas of the outside diameter to be deficient in machining stock. The pattern, therefore, had to be altered after the first three castings were made, to add stock to the outside diameter surfaces. Some trouble with warpage from a flat plane was experienced, but castings are generally made to meet the required flatness.

Fig. 26a — Upper missile ring pattern equipment (cope).



Results of tensile tests from attached coupons as well as separately cast coupons appear in Table 7. One thousand four hundred fifty lb of metal are required to pour an azimuth ring, and its net shipping weight is 318 lb. The casting yield is, therefore, 22 per cent.

## UPPER MISSILE RING CASTING (FIG. 25)

### *Design*

The design of this casting was not altered by the foundry from the customer's original blueprint. Test bars were attached to three of the risers rather than directly to the casting. This decision was dictated by the basically thin sections in all parts of the casting.

### *Pattern Construction*

Figures 26a and b show this pattern equipment. The casting is gated from each of two runners into hot metal risers on either side of the casting. As in the case of the previously described parts, it is designed for two-up pouring.

### *Processing, Inspection, Results*

The casting is poured simultaneously from two ladles in approximately 10 sec.

Complete radiography of the upper missile ring casting requires the use of x-ray, Iridium 192 and Cobalt 60, and a total of 51 exposures are needed to give full coverage of the part.

No difficulty was encountered meeting customer's requirements for Class 1 and Class 2 areas on this part, and with the exception of a rare crack generated by heat from cut-off operations, no problems have been discovered in either magnetic particle or dye penetrant inspection.

Table 8 shows the results of tensile tests from separately cast test bars and from the test bars attached to the risers of the casting. The gross pouring weight of the upper missile ring is 1,150 lb, and the

TABLE 7 — MECHANICAL PROPERTIES — AZIMUTH RING CASTINGS

| Serial No. | Yield, psi 0.2% Offset | | Ultimate Tensile, psi | | Elongation, % | |
|---|---|---|---|---|---|---|
| | Separately Cast | Attached | Separately Cast | Attached | Separately Cast | Attached |
| 1 | | 170,400 | | 187,300 | | 19 |
| | | 169,300 | | 186,200 | | 18 |
| 2 | 181,500 | 180,000 | 192,000 | 194,000 | 10 | 13 |
| | 182,100 | 182,500 | 191,500 | 197,000 | 10 | 12 |
| 3 | 175,800 | 171,500 | 191,000 | 189,000 | 11 | 14 |
| | 175,800 | 171,700 | 191,500 | 189,000 | 11 | 11 |
| 4 | | 183,300 | | 203,200 | | 11 |
| | | 182,700 | | 202,400 | | 14 |
| 5 | 182,400 | 180,800 | 197,000 | 195,500 | 10.5 | 14 |
| | 182,700 | 183,400 | 197,500 | 196,700 | 9 | 11 |
| 6 | | 180,400 | | 196,000 | | 11 |
| | | 181,800 | | 197,200 | | 14 |
| 7 | 182,700 | 178,560 | 196,000 | 192,800 | 10 | 13 |
| | 180,900 | 178,560 | 195,000 | 195,200 | 10 | 13 |

NOTE: Separately cast test bars are machined to 2 in. gage length and 0.505 in. diameter. Attached test bars are machined to 1 in. gage length and 0.252 in. diameter.

net shipping weight 114 lb for a casting yield of 10 per cent.

## LOWER MISSILE RING CASTING (FIG. 27)

The lower missile ring resembles the upper missile ring in almost every respect, both in design and in manner of processing. The main difference is that the lower ring has two additional heavy areas not found in the upper ring. These areas are fed with top risers, as can be seen in Figs. 28a and b, showing the pattern equipment. Inspection of this part requires a total of 46 exposures using x-ray, Iridium 192 and Cobalt 60. In all other respects the results of effort made toward casting lower missile rings has been the same as for the upper missile rings.

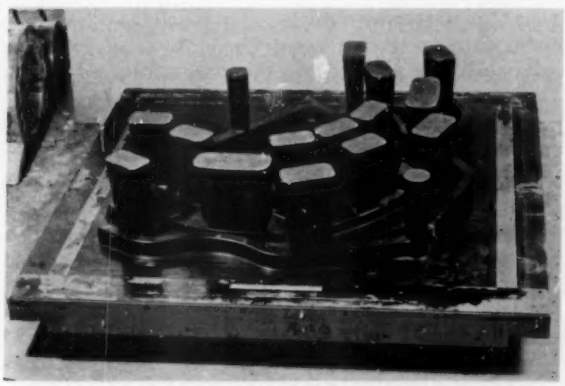


Fig. 28b — Lower missile ring pattern equipment (drag).

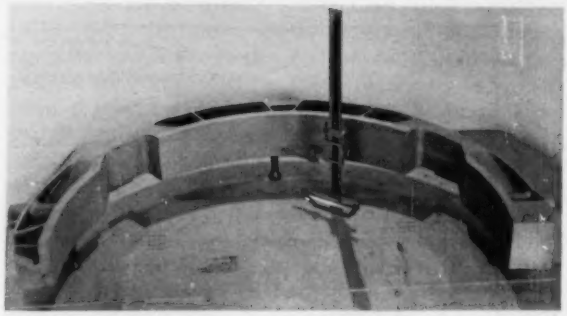


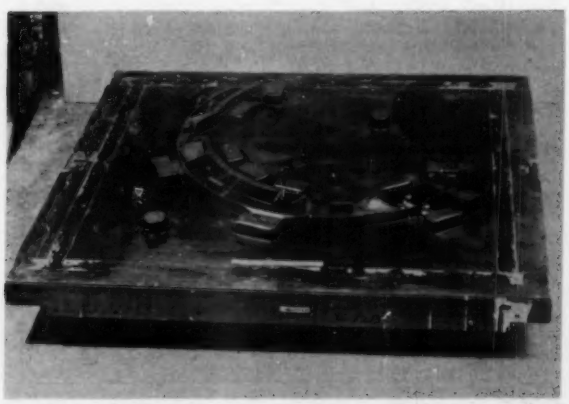
Fig. 27 — Lower missile ring casting.

TABLE 8 — MECHANICAL PROPERTIES — UPPER MISSILE RING CASTINGS

| Serial No. | Yield, psi 0.2% Offset | | Ultimate Tensile, psi | | Elongation, % | |
|---|---|---|---|---|---|---|
| | Separately Cast | Attached | Separately Cast | Attached | Separately Cast | Attached |
| 1 | 181,800 | 175,200 | 198,000 | 190,600 | 10 | 9 |
| | 180,600 | 173,200 | 196,000 | 188,200 | 8 | 11 |
| 2 | 183,000 | 178,000 | 192,000 | 193,000 | 11 | 7 |
| | 181,200 | 178,000 | 192,000 | 190,000 | 11 | 5 |
| 3 | 177,900 | 182,500 | 189,500 | 193,400 | 12 | 13 |
| | 177,600 | 188,200 | 189,500 | 194,200 | 12 | 11 |
| 4 | 180,000 | 176,000 | 189,500 | 191,300 | 12 | 14 |
| | 179,400 | 178,800 | 190,000 | 192,200 | 10 | 15 |
| 5 | 180,600 | 180,200 | 195,000 | 194,200 | 9.5 | 12 |
| | 183,000 | 182,000 | 195,000 | 195,000 | 9.5 | 11 |
| 6 | 180,000 | 179,700 | 195,500 | 194,300 | 11 | 11 |
| | 178,800 | 178,700 | 195,000 | 193,400 | 10 | 9 |

NOTE: Separately cast test bars are machined to 2 in. gage length and 0.505 in. diameter. Attached test bars are machined to 1 in. gage length and 0.252 in. diameter.

Fig. 28a — Lower missile ring pattern equipment (cope).



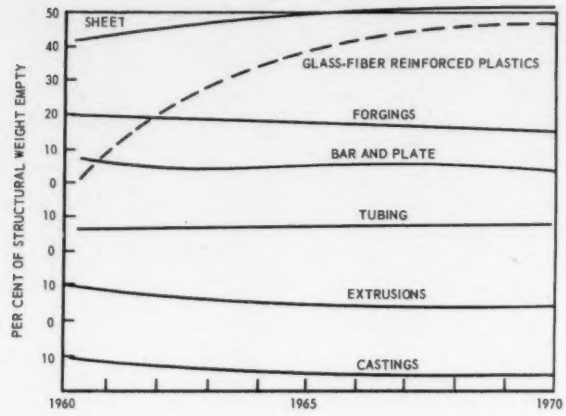



**Fig. 29 — Estimate of usage of various material forms in air vehicle structures from 1960 to 1970.**

Mechanical properties for attached and separately cast coupons appear in Table 9.

## CONCLUSIONS

An extensive development program has resulted in both a process and a technique for the production of high integrity, high strength steel castings, larger in size than has heretofore been found in the casting industry. Certain distinct advantages have accrued as a result of this program:

1. High surface quality.
2. High degree of soundness.
3. Reduction of cracks and hot tears.
4. A reliability factor much greater than might be expected with normal sand castings. This particular advantage is borne out by the fact that each of the five exhibits described above was manufactured successfully on the first attempt and sold to the customer as a usable item.

**TABLE 9 — MECHANICAL PROPERTIES — LOWER MISSILE RING CASTINGS**

| Serial No. | Yield, psi 0.2% Offset | | Ultimate Tensile, psi | | Elongation, % | |
|---|---|---|---|---|---|---|
| | Separately Cast | Attached | Separately Cast | Attached | Separately Cast | Attached |
| 1 | | 174,500 | | 193,800 | | 14 |
| | | 174,900 | | 193,800 | | 13 |
| 2 | 179,400 | 179,500 | 191,000 | 190,000 | 10.5 | 13.5 |
| | 179,900 | 182,000 | 190,500 | 192,000 | 10.5 | 11 |
| 3 | 179,100 | 178,000 | 189,500 | 192,500 | 11 | 11 |
| | 182,100 | 175,400 | 185,500 | 192,700 | 11 | 12 |
| 4 | 174,600 | 180,000 | 189,000 | 190,800 | 10 | 11 |
| | 174,000 | 180,800 | 188,500 | 192,500 | 11.5 | 10 |
| 5 | 180,120 | 181,200 | 195,500 | 194,800 | 8.5 | 12 |
| | 180,000 | 182,000 | 195,500 | 195,200 | 10 | 13 |
| 6 | 182,700 | 182,200 | 197,000 | 195,400 | 10 | 12 |
| | 180,000 | 182,300 | 195,000 | 196,700 | 11 | 11 |

NOTE: Separately cast test bars are machined to 2 in. gage length and 0.505 in. diameter. Attached test bars are machined to 1 in. gage length and 0.252 in. diameter.

## SUGGESTIONS FOR FURTHER WORK

The *1960 Aerospace Forecast of Technical Requirements* published by the Aerospace Industries Assoc. presents two sets of curves which define in part the areas where additional work is required. These curves are reproduced herein as Figs. 29 and 30. Figure 29 shows the anticipated application of various material forms as a percentage of the empty structural weight of aerospace vehicles. It is possible to interpret these data in several ways. Certainly it suggests the advisability of getting into the glass-reinforced plastic business.

Of more importance to the subject at hand, it means that those foundries interested in the class of work represented by this chart are going to have to do a continually better job if a fair share of the existing market is to be captured or maintained.

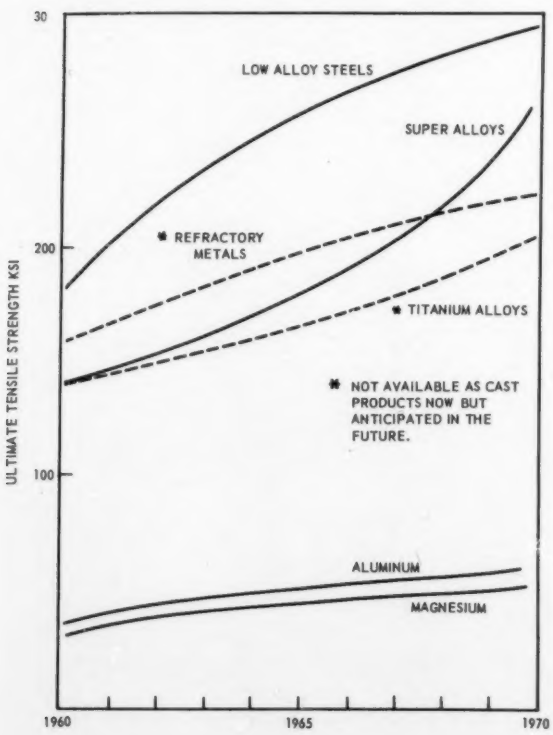



**Fig. 30 — Anticipated increase in properties of cast alloy in period 1960-1970 as forecast by Aerospace Industries Assoc. Actually low alloy steels in 260-300 psi range are currently available.**

The second chart (Fig. 30) is included because it is felt that it requires comment, and because it does not represent the most correct story relative to low alloy cast steels. Actually there are available at the current time cast low alloy steels with mechanical properties in the 260,000-300,000 psi range, so that the target date of 1970 forecast for such alloys is certainly on the pessimistic side.

Strength alone is not sufficient, and one definite area of further fruitful work exists in determining the best methods of translating these high mechanical properties to all parts of a complicated casting. This means increased knowledge of feeding distance, use of padding, the effect of dissolved gases on feeding and similar factors which control metal soundness.

A second major area for added work relates to section size effects. This is complicated by the difficulties of feeding and radiographing heavy sections, and improved nondestructive testing techniques may be required before this problem is completely solved. Solved it must be, however, if the casting process is to be used to its fullest potential.

### *Dimensional Control*

A third major area for improvement of casting quality is in dimensional control. As a starting point, however, one must understand where the situation stands today. Investment type castings which are generally considered to be among the front runners where precision is concerned normally are obtainable with tolerances of ±0.005 in./in. If the experiences outlined earlier of this presentation are compared with investment castings on an in./in. basis it can be seen that approximately $\pm\frac{3}{32}$ obtained corresponds relatively well with investment castings on an in./in. basis. This does not mean that those who produce large castings should be satisfied with the state of the art, but it does mean that the state of the art is not as black as it is sometimes painted.

Dimensional control improvements will come about in several ways; improved understanding of the mechanics of solidification and cooling, increased use of machine tools as part of the cleaning operation and improved fixturing during heat treating operations.

Surface finish of steel castings is a fourth area where studies of refractories and binders will bring about some improvements. Machine finishing will also contribute to surface improvement in the same way that it will aid dimensional control. One factor to be considered in deciding how much effort to place on surface finish research is the fact that a major portion of the surface area of adequately fed steel castings represent riser pad or gating locations which by their nature are hand or machine finished anyhow, and improved refractory systems will yield no advantage at such locations.

A fifth type of work to be done involves the writing of specifications for steel castings which represent the current state of the art. This includes not only mechanical property specifications but also various inspection specifications. At the present time no government specifications exist for steel castings much beyond the 175,000 psi level, and no radiographic references are available for sections under one in. thick. Work is underway with several government agencies to correct both of these conditions so that greater advantage can be taken of what has already been proved feasible. This does not mean that the search for improvements should not go on. It does mean, however, that the best use is not being made of what is now known.

Finally, there is a requirement for a more detailed study of the effects of such tools as vacuum degassing, ultrasonics, exotic ladle injections upon the behavior of cast steels relative to both solidification behavior and subsequent mechanical properties.

It is the authors' firm belief that the last ten years have been exceptionally fruitful ones in terms of technical gains for high integrity steel castings. It is intended to try to make the next ten years even more fruitful.

## ACKNOWLEDGMENTS

The tactical prototypes of the transporter-erector-launcher were designed and manufactured by the Tapco Group of Thompson-Ramo-Wooldridge, Inc., for the Martin Co., prime contractor for the Pershing weapons system. The Pershing program is under the technical supervision of the Army Ballistic Missile Agency and within the responsibility of the U.S. Army Ordnance Missile Command, Redstone Arsenal, Huntsville, Ala.

The American Brake Shoe Co. gratefully acknowledges the help and cooperation of the Tapco Group, and in particular of Mr. Joseph Chase, Materials Applications and Metallurgy and Mr. Robert Wilson, Tapco R & D Engineering.

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# PRINCIPLES OF DIE CASTING

## *Some fundamental thermal and mechanical aspects*

by Alan B. Draper


### ABSTRACT

**The principles of fluid mechanics and thermodynamics have been applied to die casting in a number of scattered articles. This report condenses many of these publications and compares their conclusions.**

**The design of dies for high pressure casting depends largely upon experience and art. Several concepts of die filling are presented and experimental evidence that Frommer's concept is essentially correct is given. This theory of die filling should aid a designer significantly in developing the proper location of gates and vents.**

**Barton's concept of the change of kinetic energy into heat energy with a subsequent thermal rise within the casting is reviewed. The present author uses the equations of thermodynamics and fluid mechanics to demonstrate that this should happen. Sakui presents experimental evidence that at the end of the injection stroke there is sufficient temperature rise to cause the die casting to remelt.**


### INTRODUCTION

The principles of fluid mechanics and thermodynamics have been applied to die casting in a number of scattered articles. Most of them were written by men with great practical experience in the field, a few were based on laboratory experimentation. The present paper condenses these publications and compares the findings of the previous investigators.

### METAL FLOW IN DIE CAVITIES CONCEPT

The first theory of forced metal flow in dies was presented in 1932 by Leopold Frommer.[1] His conclusions were drawn from practical experience in the field of zinc die casting. Frommer postulated that molten metal follows the laws of fluid mechanics, and established several ratios relating the gating area to the cross-sectional area of the die cavity, i.e., G/C. However, ratios greater than ¼ are rarely used in practical die casting.

According to Frommer, metal injected into a die of rectangular cross-section, traverses the cavity in a jet with a cross-section identical in shape to that of the gate (Fig. 1a). The jet strikes the opposite wall where it gathers into a pulsating pool. As the pool fills, it becomes more turbulent; portions of it, called forerunners, flow back along the walls of the cavity ahead of the advancing pool (Fig. 1b). Eventually the pool overtakes the forerunners because they slow down from friction and heat loss to the cavity walls; considerable turbulence (Fig. 1c) and splashing occurs as the die back fills to capacity (Fig. 1d).

With a small injection velocity, and a G/C ratio less than ⅓, the forepart of the molten pool is rather uniformly turbulent while the remainder is relatively stable; otherwise, the entire pool is exceedingly turbulent. Adequate venting is a third requisite for minimizing turbulence and consequent entrainment of air in the casting.

The length of the die cavity filled before the pool overtakes the forerunner (Fig. 1d) is a function of the velocity and temperature of the injected metal, of the cavity geometry and of the die temperature.

#### *Brandt's Experiments*

In 1937, W. R. Brandt[2] challenged Frommer's concept of metal flow. He injected aluminum into a cavity of rectangular cross-section through a gate which extended across the full width of a shorter side. To indicate metal flow within the cavity, he

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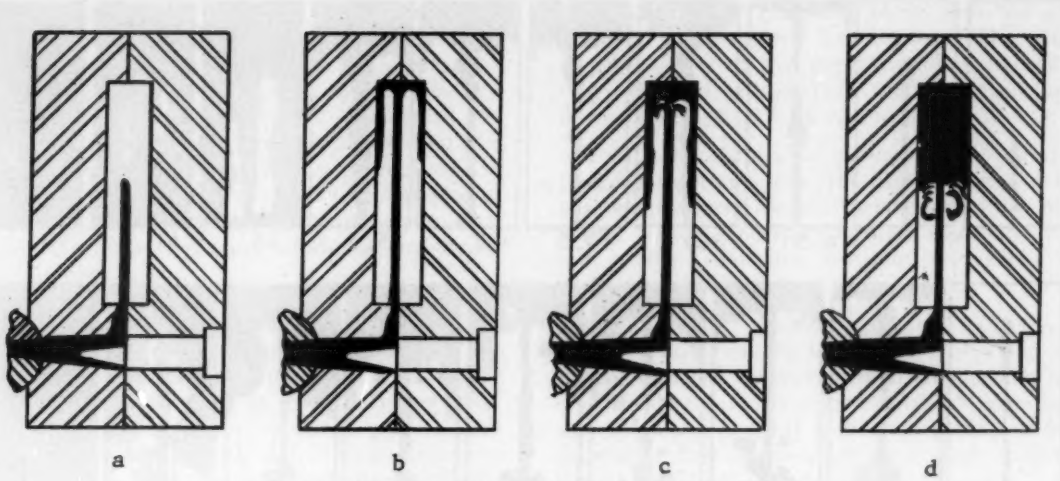



Fig. 1 — Frommer's concept of metal flow within a die.[1]

arranged a series of electrical contacts to project within the die cavity and connected them to instruments and recording equipment. From his data, he concluded that the metal stream filled the cross-section of the die immediately after passing the gate and flowed forward to fill the cavity completely (Fig. 2).

Examination shows that Brandt's data are not consistent and give little support to his conclusions; indeed, one series of tests confirms Frommer's theory rather than refutes it. Further, Brandt failed to specify the aluminum alloy he used, its temperature at injection or the temperature of his dies. The low injection pressures (367 to 515 psi) were certainly far below those used in present day commercial practice.

A combination of low pressure and temperature may account for Brandt's inconclusive evidence, or, as Koester and Goehring[3] challenge, splattering rather than flowing could have closed his circuits prematurely, thus rendering the data random and inconclusive. Finally, Brandt did not claim to have repeated his experiments.

Brandt's research was valuable because it focused the attention of Koester and Goehring[3] on Frommer's work, and inspired them to design an experiment to resolve the opposing views of Frommer and Brandt. Their ingenious equipment included two rectangular plates of heat resisting glass which formed windows on opposite surfaces of a steel die which had a geometry and size similar to Brandt's die. On one window they centered a spotlight, and on the other a camera which could expose up to 3500 frames/sec. They injected Wood's metal.

Koester and Goehring concluded that the flow of metal into a die is basically that presented by Frommer (Fig. 3), but they minimize the effect of chilling, cohesion, friction and turbulence. Frommer maintained that the pool opposite the gate detains the majority of the metal, finally passes the forerunners and then fills the die. Koester and Goehring state that the forerunners themselves form a shallow pool at the gate end with a certain increase in turbulence. However, they concede that their conclusions may have been affected by the glass die walls, and that the behavior of Wood's metal is not necessarily that

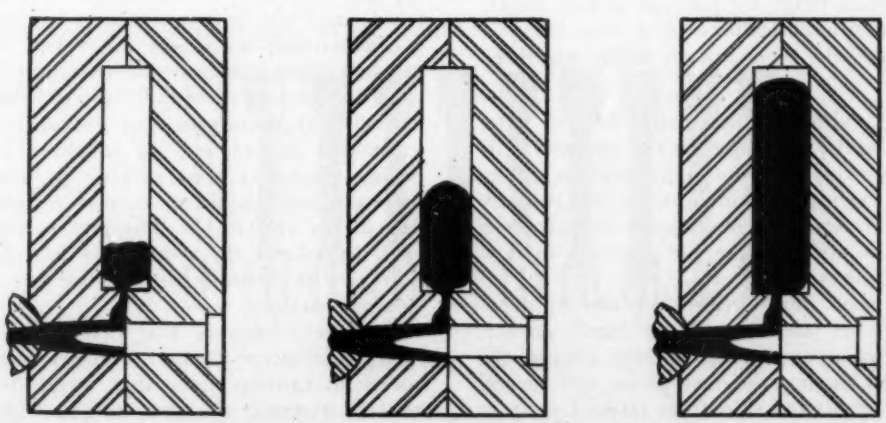


Fig. 2 — Brandt's concept of metal flow within a die.[2]

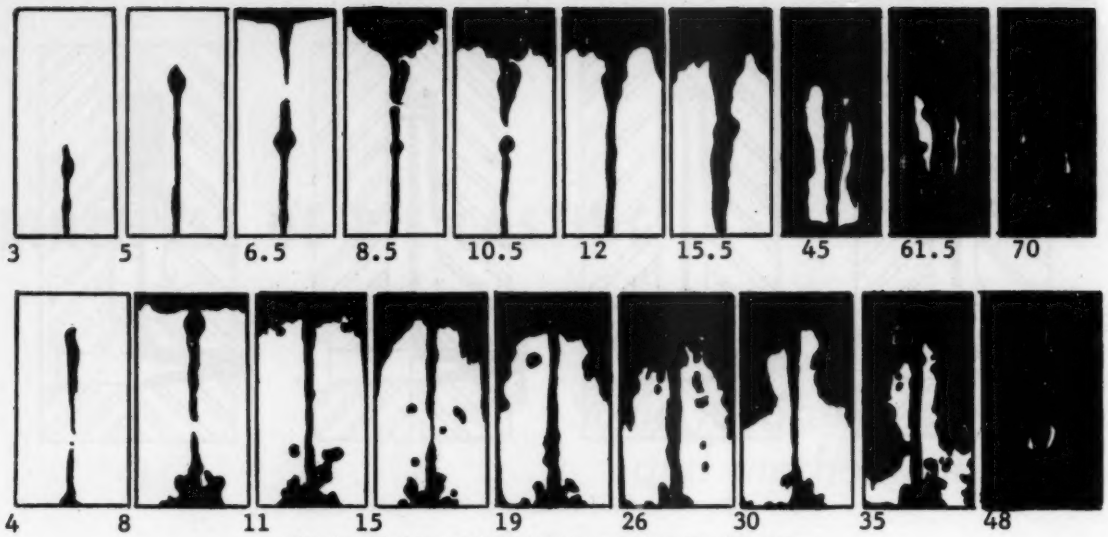



**Fig. 3 — Koester and Goehring's experimental demonstration of metal flow within a die cavity.[3] Time in 0.01 sec.**

of commercial die casting alloys. Koester and Goehring's experimental proof of Frommer's concept of die filling is a major contribution to the formulation of a theory of metal flow in die casting.

### *Barton's Theory*

In several articles from 1944 to 1952, H. K. Barton[4,5,6] discussed the effect of injection pressure in greater detail than Frommer. Having studied the works of Frommer, Brandt and Koester and Goehring, he accepted Frommer's hypothesis, and stated that it was not his purpose to give further evidence that the injection of metal into dies follows the laws of fluid mechanics. Rather, he concluded that only by Frommer's theory could he explain formations which he personally had observed in many thousands of faulty, imperfect or first run castings.

Presuming Frommer's hypothesis correct, he considered the effect of temperature gradients, and velocity variations within the molten metal and in the layers adjacent to the metal-die interface throughout the casting cycle. Barton states that the way the metal travels over the cavity surface largely determines the surface finish, the presence of flow marks, cold laps and other imperfections.

Barton[7] divided the process of die filling into three phases:

*Phase I:* A jet, with a cross-section like the gate, traverses the cavity, impinges on the opposite wall and thence spreads over the major portion of the die surfaces. During the formation of this initially projected skin, small accumulations of turbulent metal occur at points where the direction of flow changes abruptly.

*Phase II:* The shell thus formed then fills by deposition of further metal upon its inner surface. The turbulent accumulations enlarge around the nuclei formed during the first phase and merge. The initially projected skin from phase I is at or near the solidus during phase II.

*Phase III:* With the cavity completely filled, the full injection pressure is uniformly exerted up on the casting through its still fluid core.

Barton[7] states that the characteristics of injection in the first phase govern the surface finish, in the second phase the solidity, and in the third phase, the strength of the casting.

### *Other Experiments*

Kayama and Ichida[8] reconfirmed Frommer's theory in 1953. Using a rectangular die with replaceable gate blocks to vary the gate geometry, they injected aluminum with 12 per cent silicon at 1310 to 1450 F (710 to 793 C) and 570 to 580 psi into dies at 175 to 250 F (79 to 121 C). Their studies showed that a high velocity stream of metal crossed the die cavity, spread on the opposite wall and flowed back along the sides to the gate where it formed vortices between the inflowing jet and the forerunners.

Although confirming Frommer's basic theory, these investigations disputed his optimum G/C ratio when applied to thin walled castings. But Kayama and Ichida used no vents or ejector pins, no casting had a commercially acceptable finish and even the best one had 6 per cent porosity by weight. Whence it is clear that venting profoundly affects filling.

Lieby[9] apparently accepts Frommer's theory, but states that in practice, the idealized free jet often changes character after striking an oblique wall or digresses and divides at corners or edges. So it is difficult to analyze the flow process properly. However, he admits the concept of free (unobstructed) flow aids in planning the parting line, runner location and venting.

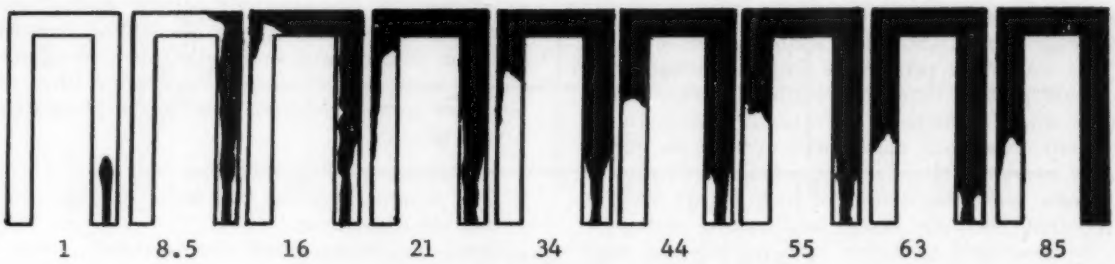
Doehler[10] observes that practical die casters are inclined to accept Frommer's theory because most unsuccessful castings show that metal freezes first on the side opposite the gate, and that shrinkage most often appears close to the gate.

Sakui and co-workers added a new dimension to the study of metal flow within die cavities by using radioactive tracers as additives to the die lubricant.[11] They blended radioactive phosphorous 35, dissolved in dilute hydrochloric acid, with a commercial die lubricant. The latter was applied to the test die in the conventional manner at selected locations in the die cavity. Examination of autoradiographs of sections from the castings made in that die permitted them to deduce the extent and location of turbulence within the die cavity.

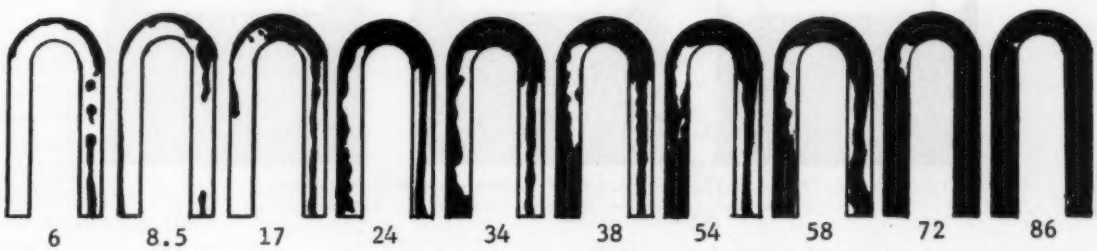
The sections showed that the area of the casting just under the surface contained almost no radioactive material (Fig. 5), but that the interior of the casting was thoroughly contaminated. This would seem to indicate that the initially projected forerunners did not entrain die lubricant, whereas the turbulently back-filling metal did. This is dramatically shown by the fact that lubricant placed in projections near the gate was swept into the casting body by the turbulent flow during filling (Fig. 6). A gas analysis of the porous sections showed that they contained hydrocarbons which were not present in the molten metal prior to injection.

Tensile test bars gave similar results. As the injection temperature rose, the higher degree of superheat greatly increased the turbulence within the die cavity, causing widespread entrainment and dispersion of the radioactive lubricant throughout much of the casting (Fig. 7).

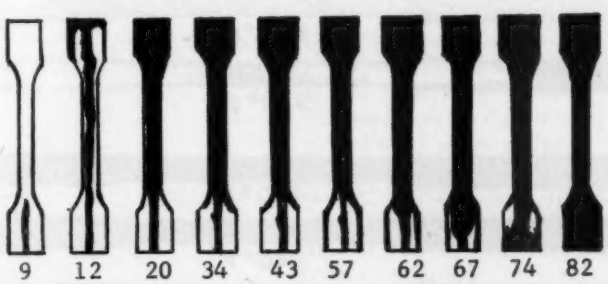
Barton[12] has previously pointed out that the filling velocity and turbulence within a die cavity during filling is proportioned to the specific weight of the injected metal (Fig. 8). Sakui's investigation showed that zinc filled a die cavity most smoothly, aluminum had more turbulence, and magnesium flowed with high turbulence.[13]




a. Filling a U Shaped Die with Right Angle Bends and a Rectangular Cross-Section




b. Filling a U Shaped Die with Rounded Bends and a Rectangular Cross-Section




c. Filling a Tensile Test Bar Die

Fig. 4 — Metal flow in dies of more complex geometry.[3]

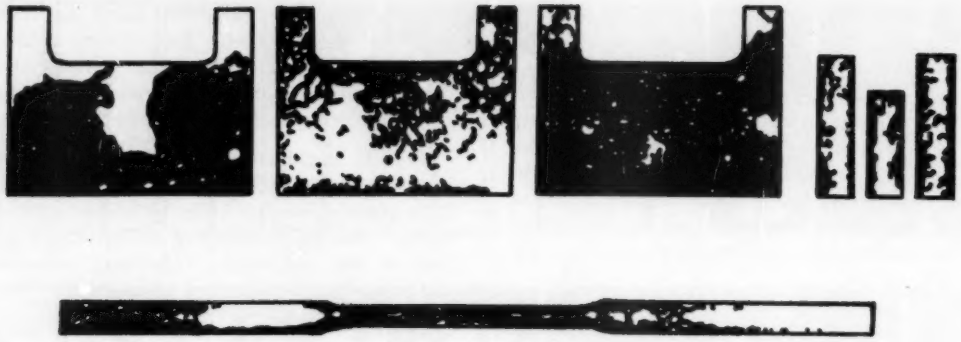


Fig. 5 — Autoradiographs of casting sections from dies lubricated with materials containing radioactive tracers.

## GATING SYSTEM DESIGN

The design of a gating system for a die cavity begins with defining the major criteria of the product. The designer must know, for example, realistic values for the minimum permissible impact strength, the allowable porosity and the required surface finish. Where minor centerline or other shrinkage is of little importance, it is uneconomic to strive for higher quality standards. If, on the other hand, radiographic soundness, maximum density or high impact strength is required, then die design and casting technique can be developed to assure such qualities in most castable parts.[14]

Undoubtedly, for a given casting, there are many practical gating designs, some of them closely approaching the optimum for certain casting conditions. The art of die casting is yielding to science slowly as fundamental knowledge and understanding of the multitude of variables increases. Most variables are now controllable during the processing of a new product.

1. The designer specifies the alloy.
2. The designer specifies the initial gating, venting and cooling systems.
3. Both the designer and the operator control the injection variables and the thermal balance.

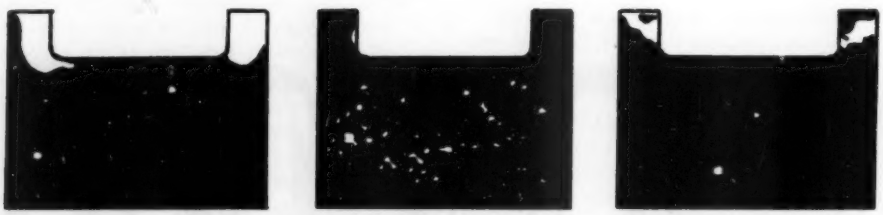


Fig. 6 — The distribution of radioactive lubricant initially placed in recesses on the gate side of the die cavity.[11]

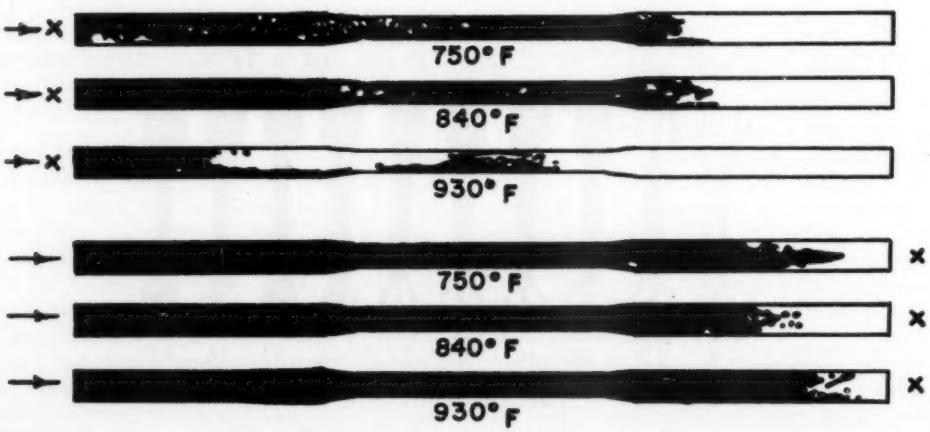



Fig. 7 — Effect of injection temperature on the distribution of radiographic lubricant within a die casting.[11]

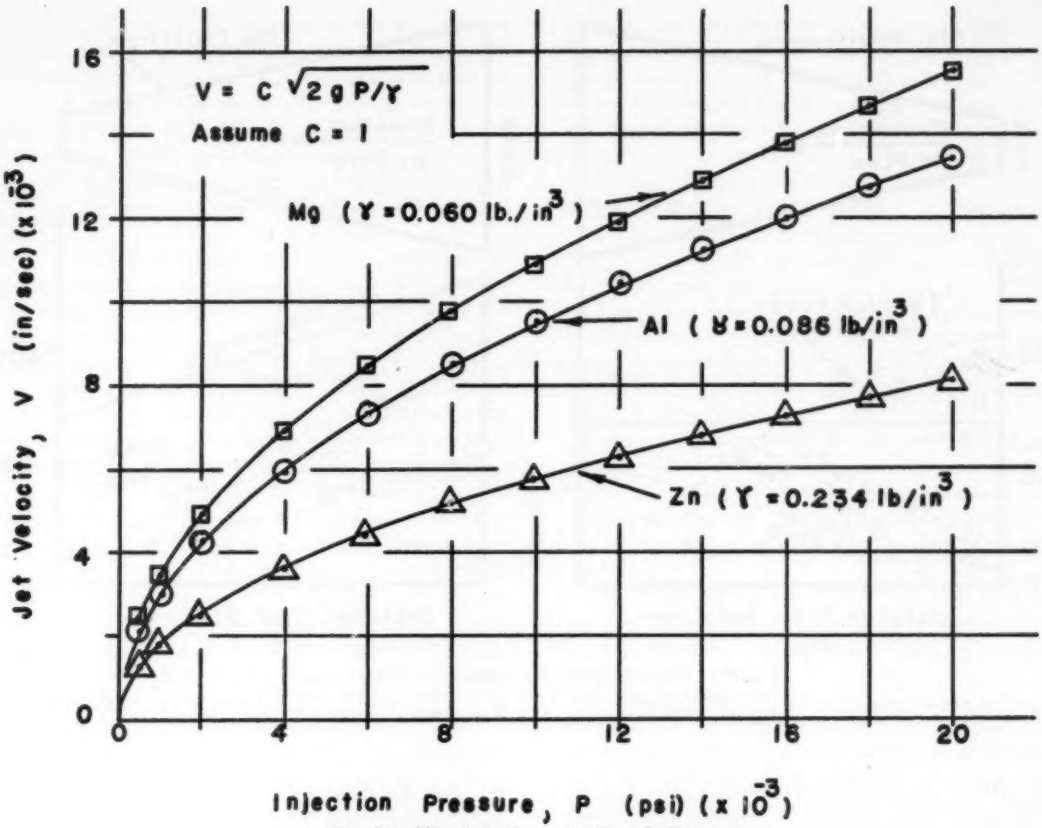



Fig. 8 — The injection velocity of die cast metals as a function of injection pressure.

Barton,[4] Doehler,[10] Sachs[15] and others list these variables:

1. Alloy Properties
   - Fluidity.
   - Specific heat.
   - Latent heat of fusion.
   - Solidification range.
   - Conductivity.
   - Specific weight.
2. Geometry of the Dies
   - Geometry of the die cavities.
   - Geometry and location of the gate(s).
   - Geometry and location of overflow wells.
   - Geometry and location of vents.
   - Geometry and location of runners.
   - Geometry and location of cores.
3. Injection Variables
   - Filling time.
   - Pressure variation during injection.
   - Final pressure after injection.
   - Height of metal in gooseneck (zinc).
   - Jet velocity during filling.
   - Accumulator and hydraulic system characteristics.
   - Plunger and piston areas.
4. Thermal Conditions of Injection
   - Temperature of alloy at injection.
   - Temperature gradients in the die.
   - Heat conductivity and radiation.
   - Die cooling and heating.
   - Relation of die mass to casting mass.
   - Average metal flow distance in the casting cycle time.
   - Energy converted to heat through friction, turbulence, and impact.

### *Gate Location*

Barton and Barton[16] advocate a gate along an edge from which it can be clearly sheared, or where its trimming will not impair the finish. But gate location also affects the character of metal flow; hence porosity and surface finish. The authors show four possible gate locations for a thin, flat, rectangular plate, and recommend a runner passing the length of the plate, and a corner gate which feeds metal diagonally across the die cavity.

Richter[17] urges gating at the smallest cross-section of the die cavity, so that the surface area of the casting increases with increasing distance from the gate (Fig. 9). This technique creates a better heat balance within the die because hotter metal impinges on surfaces of lesser area, while metal which travels farther and so cools, strikes broader areas. Thus, the heat input to the die near the gate and far from the gate is more uniform. A comparison of the metal temperature, die temperature, metal flow

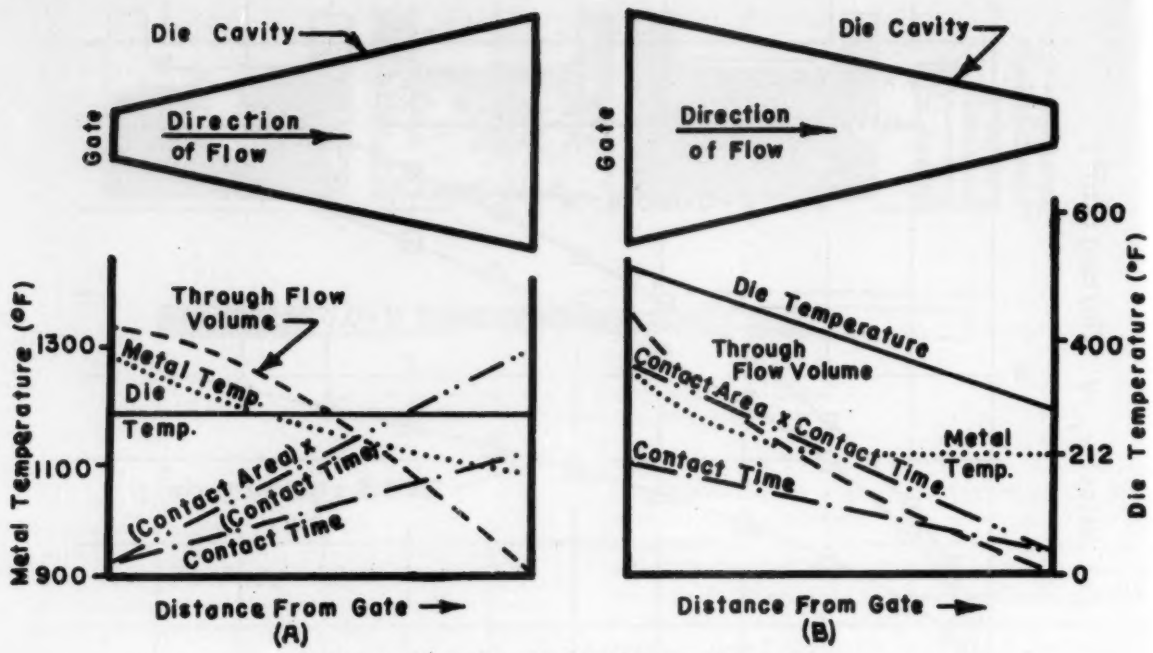



Fig. 9 — Thermal gradients within die blocks gated into the thick and thin ends of a wedge shaped casting: a no thermal gradient B, high thermal gradient.[13]

and time of flow can be shown schematically (Fig. 9).

Lieby[9] agrees with Richter,[17] but for a different reason. He recommends single gates where possible, and filling thicker sections through thinner sections to avoid trapping air in thin sections. He states that even for intricate parts, the best parting line and venting can be determined from the geometry of the casting using hydrodynamic principles. The position and arrangement of the runner depends largely on the geometry of the product. Warning against severe changes in cross-section, he illustrates with three castings, one of which is ring shaped with a tangential gate.

### *Geometry of the Gating System*

Frommer[1] delineated several criteria for the gate of a rectangular cavity without cores:

1. For a gate area to cavity area ratio of $1/4 < G/C < 1/3$ much of the die is filled before the initial pulsations subside. For $G/C > 1/3$ pulsations continue while the entire cavity fills. However, even if $G/C < 1/3$ high injection velocities engender severe turbulence.
2. Turbulence increases with both the injection velocity and the gate to cavity area ratio (G/C).
3. Regardless of injection velocity if $G/C < 1/3$ the flow of metal steadies as the cavity fills.

Brandt[2] contradicts Frommer's analysis of the effect of G/C. In his work, he found no variations in turbulence as G/C varied from less to more than 1/3. But he does grant that minimum filling time did occur using gates which were as wide and 20 to 30 per cent as thick as the cavity. Thus, both seem to agree. Brandt further warns that if G/C is too large (a) the die cannot be vented sufficiently to prevent blisters (porosity) near the gate and (b) the gate cannot be so easily removed.

Kayama and Ichida[8] say that gate thickness does not influence metal flow nearly as much as the gate width. They state that the optimum specific weight or soundness of the casting was obtained with a G/C of 1/9 to 1/3, limited by a gate width to cavity width of 1/3 to 5/6 and a gate thickness to cavity thickness of 1/6 to 1/2. The data were for an aluminum casting about 1/8-in. thick. Thus, they conclude that Frommer's gate size values are not applicable to a thin walled section. It should be noted that no venting was applied to the die used in their research.

Yet their findings do confirm Frommer's statement to the extent that better filling occurs at $G/C < 1/3$. They disagree with the limitation of $G/C < 1/4$ for thin walled castings. None of the castings produced by these investigators would be acceptable in commercial practice because of the poor surface finish. Barton and Barton[16] state that for thin walled castings such as a bowl, the thickness of the gate should be nearly equal to that of the casting wall. In addition, overflow wells of a volume nearly equal to the casting should be provided. However, the latter should not be located opposite the gate.

### *Alignment Chart*

Proutt and Huling[18] have developed an alignment chart which relates the length of gates $L$, casting volume $V$ and the gate thickness $T$. The chart is based on the empirical equation

$$L = \frac{0.03347\ V^{0.746}}{T}$$

All solutions of the equation would be on a plane intersecting the three coordinate axes at various distances from the origin. The authors give no text to explain the chart or the derivation of the empirical formula, but they do give several examples from which one may define gate width, thickness and gate length. No provision is made for the geometry of the castings or for the many other variables listed earlier.

One die casting company proportions the gate by using the concept of filling time. They know from past tests that at least 0.03 sec is needed to fill a die because of limitations in the hydraulic injection system. Also, they have established that, except in extremely large castings (over 30 lb/shot), 0.10 sec is too long to achieve sound castings. Once the filling rate for various gate sizes is determined by experimentation, the concept of filling time can be a valuable tool for designing gating systems.

Barton[5] stresses the need for adequate venting, and a gating system designed to obtain proper heat balance by correct positioning of gates, vents, blind runners and overflow wells with respect to the geometry of the die cavity. He recommends that a single sufficiently large gate is preferable to multiple gating because the latter causes extra turbulence. Blind runners and overflow wells permit greater heat flow to cooler areas, thus making single gating practical.

Barton cites an example which Brandt maintained could not be made successfully by single or even double gating. Barton's[5] solution is a correctly located single gate of appropriate size combined with a blind runner and venting both inside the cavity through a loosely fitted core or ejector pin and outside where the backflowing metal meets the incoming stream. Barton claims there is no practical limit to the length of metal flow in a cavity provided that the mold has correct even heat throughout and is properly vented.

Brandt gives limits of metal flow through gates of various sizes, he specifically claims that based on actual practice, with an injection pressure of 515 psi, no more than 160 in. of aluminum can be forced through the smallest cross-section (0.5mm or 0.02 in.)$^2$ of a gate opening. This applies only to castings with a small surface area to volume ratio. Brandt also states that there is no theoretical calculation of the maximum distance over which a metal can flow through a die cavity before solidification.

He feels that such a calculation is of great importance for correct gate location. Barton's view seems the more logical, because when today's alloys, machines and temperature controls are considered, there is no practical limitation to metal flow within a properly designed die cavity.

### *Vents*

Barton[5] states that porosity is most likely to occur where vents are too small or so poorly located that they are blocked before the cavity fills. To prevent this, some vents should be placed as near the gate as possible to allow air to escape from that part of the cavity which fills last.

Koester and Goehring[3] show that venting has a most important bearing on injection speed and casting quality. They show that the speed of the entering jet is seldom equal to that of a free jet. When the vents were sealed, the jet speed which was 18-20 m/sec (59 to 66 ft/sec) was reduced to 8-10 m/sec (26-33 ft/sec), i.e., a reduction of 50 per cent. Increasing injection pressure will not increase the injection speed unless adequate venting is provided. Their data show that the per cent decrease in jet speed is 20 per cent at 20 kg/cm$^2$ (283 psi) and 50 per cent at 60 kg/cm$^2$ (859 psi).

Brandt[2] states that vents should not exceed 0.1 mm (0.004 in.) in thickness and should be positioned so as to conform to the metal flow in the die cavity, usually at the most distant point opposite the gate and near projections and recesses. Certainly vents should not exceed 0.008 in. thickness if fins are to be avoided, but commercial practice uses vents up to 0.016 in. deep. Of course, since Brandt's concept of die filling has been shown to be erroneous, his recommendations based on this concept are subject to question.

However, Werley[14] does support Brandt's suggestion by recommending a vent and overflow well nearly opposite the gate and then adds additional vents where needed.

Recently, die casting into previously evacuated die cavities has been extolled as a means for drastically reducing venting problems and dross inclusions. While it is undoubtedly true that this technique does improve casting density, proper die design would have been equally effective in improving casting quality in many instances. In addition, vacuum die casting may slow up the total cycle time. In certain difficult to vent cavities (as deep fins) or thin sections, vacuum die casting has proven effective. No quantitative research data exist in the literature, but will become available in the future.

Ejection pins, slides, insert clearances and the parting line all add to the total cavity venting area. Some designers use a vent area of at least 50 per cent of the ingate area, yet logically the air in the gating system as well as that in the cavity must also be vented. At the moment, no good logical venting area design criteria is known.

### *Overflow Wells*

Barton[19] considers that an overflow well at the far end of the die cavity serves merely as a heat source. Moreover, it functions only moderately well because the metal therein is cooled by its passage over the die walls. A better arrangement would be to gate through the far end of the cavity and use a blind runner or waster bar if needed to achieve an even heat distribution throughout the die block.

For thin walled castings, Barton and Barton[16] recommended that overflow wells be placed where they are not directly opposite the feeding gate to permit the overflow well to vent the die cavity. In such cases, they recommend an overflow well with nearly the same volume as the casting. This is a rather different viewpoint from that of Werley[14] who utilizes the overflow to remove the first flow of cooler metal. He strongly recommends the proper use of overflow wells

to produce sound castings. He states that they perform three important functions, to:

1. Serve as receivers for the air within the die cavity prior to venting (vents should lead from the overflow well to the atmosphere).
2. Serve as reservoirs of heat for the cooler portions of the die.
3. Draw off skins and the initial flow of cold metal, thereby improving surface finish.

Experimental dies were developed at the New Jersey Zinc Company to produce standard tensile specimens (¼-in. diameter) and impact specimens (¼-in. square). The test dies furnished convincing proof of Werley's principles. The data show that conventional gating (0.030 in. deep) into the test bar and an overflow well at the far end produced castings with considerable porosity, especially at the end opposite the gate. The original design was modified.

Four vents were added along the impact bar's lateral dimension and a 5/16-in. square waster bar, with a large overflow well was added between the original two cavities. The gates and overflow wells were increased in size in a series of graduated steps until radiographically perfect specimens were obtained. The final result was a gate thickness for the impact bar of 0.083 in., the overflow runner was increased to 0.088 in., and the overflow size was changed from 0.07 to 0.6 cu in. The ¼-in. tensile bar had a gate 0.075 in. deep, an overflow well gate 0.030 in. deep and the well volume increased from 0.06 to 0.15 cu in. When the overflow well could not be further expanded because there was no more room in the die block, it was found that some porosity still existed in the grip portion of the tensile bar.

### *Test Specimen Improvement*

These die modifications resulted in generally improved test specimen uniformity. Any five bars taken at random were within ± 10 per cent of the true average in 95 out of 100 cases. Such test bars could reliably reveal the variations in mechanical properties of experimental alloys.

Additional work was carried out in the field where customers were dissatisfied with casting quality, or where there were too many rejects. Three particular die design modifications, using larger overflow wells and deeper ingates, resulted in 500 per cent increases in impact values and the elimination of porosity. Trimming problems were more difficult, but not after suitable tooling was built. One surprising result was that the larger gates and overflow wells reduced the number of rejects to such a low level that the remelt load decreased.

An interesting side light to the work done at one company is that a number of years ago they designed a die which had provisions for 96,000 variations in gating geometry. After working with it for a considerable period, the project was dropped because it was felt that suitable modifications in gating size, overflow well geometry and venting could make any die produce good castings in the field. Thus the fundamental relationships among the important variables were not discovered or recorded.

### *Runner Design*

Barton and Barton[16] note that there are a number of possibilities for runner designs and locations, which can be varied to meet the needs of conservation or dissipation of heat to maintain thermal equilibrium and appropriate die temperature gradients. Shapes of runner cross-section vary from round to trapezoidal or elliptical; from wide and flat to narrow and thick. If the heat transfer rate of the runner system is high (high surface area to volume ratio) the metal is cooled somewhat by the time it reaches the gate.

This is an advantage for a casting of thick cross-section, but is poor practice for a thin walled casting because the metal will lose more heat traveling within the cavity, thus causing incomplete fusion and deep striations. One solution would be raising the injection temperature or pressure, but this increases the total quantity of heat to be dissipated. A better solution is found in altering the heat transfer between the runner and cavity by using a runner more nearly circular in cross-section, thus giving a low surface area to volume ratio and conserving heat.

In practice, most die casting runners are trapezoidal; about two times as wide as they are deep. Radiography of gating systems has shown that as they approach a round or square cross-section, internal porosity is more likely to appear.

### *Sprue Pin Design*

Barton and Barton[16] state that in edge gated cavities typical of cold chamber machines, the sprue is merely a small appendage to the main runner and has little influence on metal flow or the thermal equilibrium within the die blocks. In center gated dies for zinc die casting, the sprue may have considerable influence on the thermal balance within the die block. Heat transfer from the sprue depends upon the contact area between the sprue pin and the cast metal.

A slender sprue pin presents a small heat absorption mass and consequently, because of friction, sprue pin tip temperatures may occur which are equal to or greater than the temperature of the injected metal. This impedes rather than aids solidification at the gate. Therefore, it is better to use a well splayed sprue of large diameter in conjunction with a blunt conical sprue pin. Thus an adequate cross-section can be maintained in a relatively thin walled sprue.

This design allows the sprue to be water cooled and a better die block temperature gradient can be maintained, because a large proportion of the heat from the injected metal is absorbed by the sprue pin and a lesser quantity passes to the adjacent die cavity.

## THERMAL EQUILIBRIUM IN DIE CASTING

In any casting process which employs reusable molds, the establishment of a thermal equilibrium is an important part of successful operation. Thus, in a die casting operation, the dies gradually absorb heat until the heat input is balanced by heat losses caused by radiation to the surroundings and heat transfer by conduction to the machine frame. Barton[12] states that the rate at which the temperature rises and the value finally attained are a function of the operating cycle if other conditions are equal.

Assuming no external heating or cooling, the heat input to the die per unit time is equal to the heat content of the injected metal plus the heat given up by the solid alloy as it cools to the ejection temperature and plus the latent heat of the alloy. Stated in mathematical terms this is:

$$H = M(n)\,[\,S_a\,(t_i - t_m) + S_b\,(t_m - t_e)\,] + L$$

where:

- $H$ = the heat input to the dies (Btu/lb or kc/kg).
- $M$ = the weight of the shot (lb or kg).
- $n$ = the number of shots per unit time.
- $S_a$ = the specific heat of the liquid alloy (Btu/lb − F or kc/kg − C).
- $t_i$ = the injection temperature of the molten metal (F or C).
- $t_m$ = melting point (F or C).
- $S_b$ = specific heat of the alloy when solid (Btu/lb − F or kc/kg − C).
- $t_e$ = temperature of ejection (F or C).
- $L$ = latent heat of alloy (Btu/lb − F or kc/kg − C).

Thus the heat input to the machine can be calculated with a fair degree of accuracy, but the heat losses are more difficult to assess because the heat conducted to the machine and radiated to the atmosphere cannot be evaluated. Therefore, a heat input-output analysis based on the above formula is impractical. Barton then attempts to show that the injection pressure profoundly affects the heat balance within the die. He proposes that the injection energy which is converted to heat at the instant the die cavity is completely filled is significant, and that it can be calculated from basic energy equations and simple thermodynamic relationships.

### *Modification of Barton's Thermodynamic Analysis of Metal Injection*

Brandt[2] calculated the injection velocity during die casting from the filling time of his circuit closures. He compared those values with values calculated by Bernoulli's theorem which states that the sum of the pressure, kinetic and potential energies at any point in a fluid is a constant. In mathematical terms, this is equivalent to:

$$WP_1/\gamma + WV_1^2/2g + Wh_1 = WP_2/\gamma + WV_2^2/2g + Wh_2 \qquad (1)$$

Dividing by $W$ we obtain the usual form of the equation:

$$P_1/\gamma + V_1^2/2g + h_1 = P_2/\gamma + V_2^2/2g + h_2 \qquad (2)$$

where:

- $P_1$ is the pressure exerted on the liquid at station 1 (lb/ft² or kg/m²).
- $P_2$ is the pressure exerted on the liquid at station 2 (lb/ft² or kg/m²).
- $V_1$ is the velocity of the liquid at station 1 (ft/sec or m/sec).
- $V_2$ is the velocity of the liquid at station 2 (ft/sec or m/sec).
- $h_1$ is the elevation head at station 1 (ft or m).
- $h_2$ is the elevation head at station 2 (ft or m).
- $\gamma$ is the specific weight of the liquid (lb/ft³ or kg/m³).
- $g$ is the gravitational constant (32.2 ft/sec² or 9.8 m/sec²).

Certain terms in Bernoulli's equation when applied to the injection of metal into a die cavity may be assumed to be negligible in magnitude when compared to the others. These are tabulated:

| | |
|---|---|
| $h_1 = h_2 = 0$ | From the geometry of the system and the magnitude of the injection pressure. |
| $V_1^2/2g = 0$ | The shot chamber is large in cross-sectional area compared to the ingate. |
| $P_2/\gamma = 0$ | The cavity is vented during this injection process so the liquid merely is filling against atmospheric pressure. |

Applying these values to the Bernoulli's equation:

$$P_1/\gamma + 0 + 0 = 0 + V_2^2/2g + 0 \qquad (3)$$

Thus $V_2^2 = (P_1/\gamma)\,(2g)$

$$\text{or } V_2 = \sqrt{2g\,P/\gamma} \qquad (4)$$

Brandt[2] found that his actual velocities were only about 5 per cent of the theoretical velocity, as calculated by equation(2). This led him to develop his concept of die filling. It is probable that inadequate venting and consequent excessive back pressure and turbulence were responsible for his unrealistic data.

Koester and Goehring[3] also investigated the filling time for dies. They found that their velocities were actually about 50 per cent of the values calculated from Bernoulli's equation, and that excessive back pressure greatly reduced the injection velocity.

Barton[12] attempted to explain the beneficial effects of a high injection pressure by a mathematical analysis of the temperature rise in the casting as a result of the transformation of the kinetic energy of the flowing metal into thermal energy and strain energy at the instant of impact. Unfortunately, Barton was inconsistent and confused in his attempts to apply mathematics to the problem but his physical concept is sound. His major error seems to have been his failure to use the correct thermodynamic equation to relate the quantity of heat involved to the specific heat, the weight of the shot and the change in temperature within the metal.

### *Analysis Assumptions*

The succeeding analysis assumes that the entire filling process based on data from Koester and Goehring[3] takes place in such a short time (0.030 to 0.080 sec) that most of the heat involved is absorbed by the metal because there is too little time for conduction or radiation to dissipate a significant amount of heat. Second, it assumes that the strain energy absorbed by the dies is small compared to the thermal energy absorbed by the injected metal. Therefore, most of the kinetic energy of the flowing metal will be changed into heat by:

1. Friction in the gate.
2. Viscous shear forces adjacent to the die wall (according to Brandt's boundary layer theory).

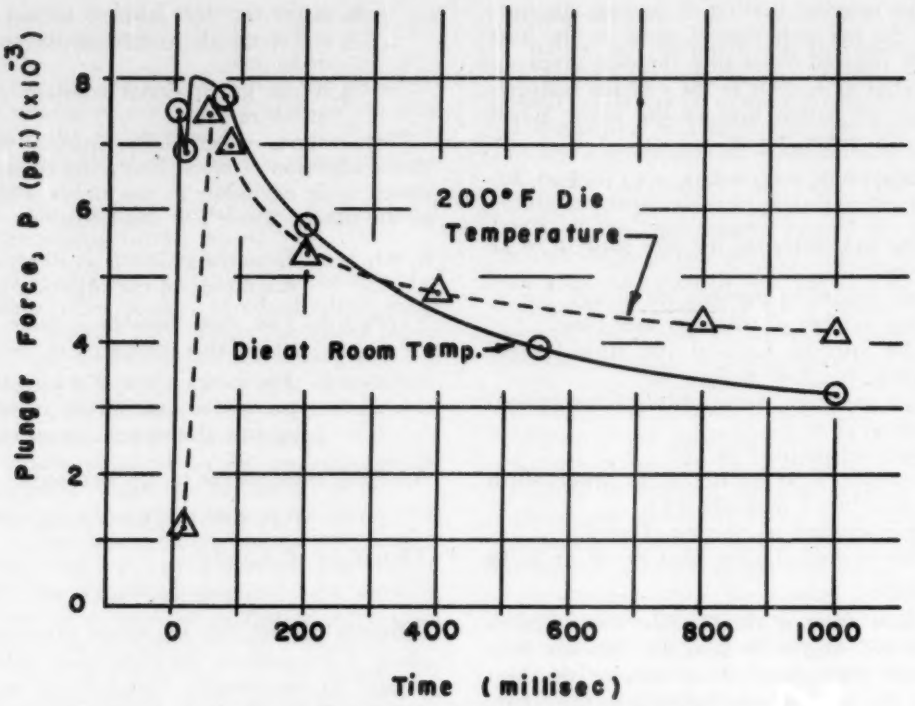



Fig. 10 — Temperature rise versus time curve for a zinc test specimen gated two ways: through one end, solid line, die at room temperature; into the side of one end, dashed line, die temperature 200 F.[13]

3. Turbulence during filling.
4. Impact at the instant when the die is completely filled.

To determine quantitatively how much temperature rise is involved, a mathematical expression must be derived to relate the rise in temperature to the kinetic energy or physical properties of the flowing metal. Such an expression is derived:
The kinetic energy *K.E.* of the flowing metal is expressed by

$$KE = 1/2\,MV^2 \qquad (5)$$

where:

M = the mass of the molten metal = w/g.
KE = kinetic energy (ft-lb or kg-m).
w = the weight of the metal flowing (lb or kg).
g = the acceleration of gravity (ft/sec² or m/sec²).
V = the velocity of the metal flow (ft/sec or m/sec).

From Bernoulli's theorem as shown earlier:

$$V = \sqrt{2gP/\gamma} \qquad (6)$$

By substitution

$$KE = (1/2)\,(w/g)\,(2gP/\gamma) \qquad (7)$$

Engineering handbooks give the mechanical equivalent of heat in any system of units desired.

One kilocalorie = 427 kg-m or one Btu = 9336 in.-lb therefore:

$$Q = KE/427 \qquad (9)$$

Where Q is the heat content of the injected metal (kc)

From thermodynamics

$$Q = w\,c\,\Delta t \qquad (10)$$

where:

w = the weight of the injected metal (kg).
c = the specific heat of the injected metal (kc/kg − C).
t = the change in temperature of the injected metal (C).

Thus by substitution:

$$Q = wP/427 = w\,c\,\Delta t \qquad (11)$$

Therefore:

$$\Delta t = P/427\,\gamma\,c \qquad (12)$$

where $\Delta t$ is the maximum temperature rise possible in a die casting during the injection process.

Although this expression is different from any proposed by Barton or the one given by Doehler[10] but attributed to Barton, the present author believes it is the correct expression for the maximum energy which could be converted to heat during the filling of a die cavity. This equation makes it apparent that the temperature rise caused by energy conversion in filling is directly proportional to the injection pressure.

### *Temperature Rise Within Die Cavity During Injection Cycle Examples*

The magnitude of the rise of temperature in a die cavity caused by the transformation of mechanical

energy to heat energy can be demonstrated by an example. An aluminum die casting is injected at a pressure 2,000,000 kg/m$^2$. Aluminum, as given by Barton, has a specific weight of 2,660 kg/m$^3$ and a specific heat of 0.214 kc/kg − C in the cgs system.

Then from equation (12):
$\Delta t = 2{,}000{,}000/427\ (2660)\ (0.214)$.
$\Delta t = 8.2$ C.

A similar problem is solved using typical English units, but a somewhat higher injection pressure, i.e., 10,000 psi, a specific weight of molten aluminum alloy of 0.086 lb/in.$^3$, a specific heat of 0.214 Btu/lb − F and a thermal equivalent of heat of 12 × 778 in. − lb/Btu.

Equation (12) will now be:
$\Delta t = P/9336\,\gamma c$.
$\Delta t = 10{,}000/9336\ (0.086)\ (0.214)$.
$\Delta t = 58$ F.

Professor Sakui developed special instrumentation for a 400 ton, horizontal platen, cold chamber die casting machine. With carefully calibrated thermocouples and an electromagnetic oscillograph, he plotted temperature versus filling time curves for molten zinc-aluminum-copper alloy injected into cavities of various configurations.[13] He also used strain gages fastened to the injection plunger to record the injection forces.

His data demonstrate conclusively that mechanical energy does transform to thermal energy each time the inertial status of the molten metal varies (Fig. 10). Such changes in inertia were shown by pressure versus filling time curves (Fig. 11) to occur when:

1. The flowing metal is forced to move in another direction (as in a complex die).
2. The quantity of metal in motion varies (as when the cavities of a multi-station die fill successively).
3. The motion ceases at the end of the plunger stroke.

These data show that the energy transformation at the completion of injection releases sufficient thermal energy to remelt the entire casting as first predicted by Barton[12] and demonstrated mathematically herein.

This research also shows the great effect gate location has on the manner in which a cavity fills, the velocity of the metal flow over the die cavity surfaces and the amount of energy transformation which will occur. A poorly gated die will not result in a sufficient rise in temperature at the end of injection to cause remelting. Contrast the temperature versus filling time curves for two tensile test bars, one gated through the end and one gated across one enlarged end (Fig. 10).

The research presented by Barton and Sakui[11,13] has shown that the injection of molten metal into die casting dies does follow the laws of fluid mechanics and thermodynamics. This indicates that the design of die casting dies can be founded on engineering

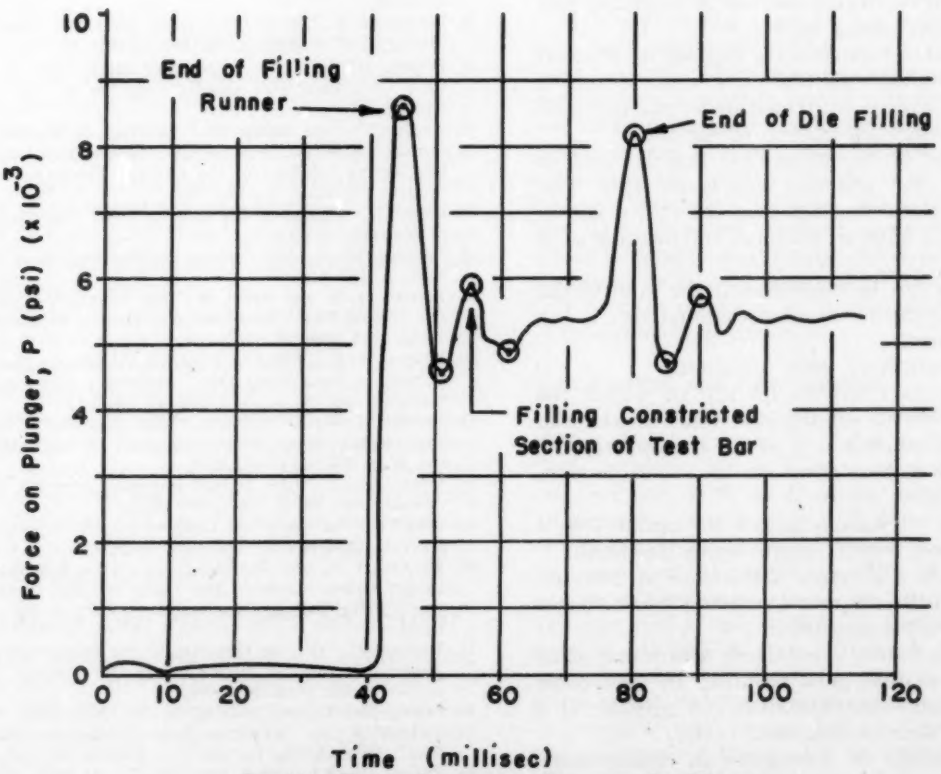



Fig. 11 — Plunger force versus time curve for filling a test specimen cavity gated into the side of one end.[13]

science tempered with judgment and experience rather than on art alone.

## SUMMARY

A number of investigators have applied the principles of fluid mechanics and thermodynamics to the study of die casting. Their contributions to the basic knowledge of the field have been significant. This paper has summarized their work in the areas of fluid flow die casting, die gating and venting and the thermal aspects of die filling. Several principles of die casting have evolved from this work including:

1. Molten metal leaves the gate in a jet with a cross-section shaped like the gate. The jet traverses the mold, impinges upon the opposite wall and follows the contour of the die wall.
2. Because the jet strikes the opposite wall, overflow wells and vents should be located so that they will not be prematurely sealed off by the pool of molten metal as it fills the die cavity. This may mean the use of several vents and especially the adding of venting near the gate.
3. Only one gate should be used wherever practical.
4. If several gates are required, they should be so designed and located that the merging jets will reinforce each other rather than meet each other head on with a consequent increase in turbulence.
5. A die casting should be gated through a thin section into a heavy section wherever possible. This will give better control of the die heat and will improve the filling of the thin section.
6. A gate should be located along an edge from which it can be cleanly sheared in trimming and thus maintain casting finish.
7. Gates should be located so the injected jet of metal does not impinge directly on a core. If this is impossible, a gate with increased cross sectional area will reduce the erosion of the core.
8. The concept of die filling time may be used to design the cross sectional area of the gate. The gate area should be such that the cavity fills in from 0.030 to 0.100 sec for small and medium sized castings.
9. Turbulence can be minimized if the ratio of the gate area to the cavity cross sectional area is less than one-third.
10. Overflow wells have three functions:
    a. To serve as reservoirs for the air within the cavity prior to venting (the vents should lead from the overflow well to the atmosphere).
    b. To serve as heat reservoirs for cooler portions of the die.
    c. To draw off skimmings and the initial flow of cold metal, thereby improving surface finish.
11. Vents should not exceed a thickness of 0.004 in. according to Brandt, but commercial practice uses gates from 0.008 to 0.016 in.
12. Vacuum die casting is at present most useful when deep thin sections must be filled. In most other cases, correctly vented cavities can produce at a faster rate than an evacuated cavity.
13. Runners should be trapezoidal in cross-sectional shape to reduce trapping of air within the molten stream and permit good ejection.
14. The energy of the moving metal is changed to heat during the injection cycle and at the moment of impact. The temperature rise can be calculated from the relationship $\Delta t = P/427\gamma c$ in metric units or $\Delta t = P/9336\gamma c$ in English units. This is most significant when casting aluminum. If the injection pressure is 10,000 psi, the temperature rise is about 58 F.

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# MISSILE CASTINGS DESIGN PROBLEMS

by R. G. Bassett


## ABSTRACT

**Casting problems are analyzed from the viewpoint of the missile designer. By discussion of design situations and by means of illustrations of several actual missile parts the foundryman is shown how his castings are integrated into more complex structures. Examples emphasize the interdependence of parts rather than the details of making castings. Reliability of the missile transcends all other considerations; therefore the discussion concludes with a note on this subject.**


## INTRODUCTION

A complex and somewhat vague procedure determines the materials and processes that go into the making of a missile. This paper will try to analyze the procedure as it applies to the specific problem of design of steel castings for missiles. It will take the viewpoint of the missile design engineer as opposed to that of the foundryman; but it will also attempt to resolve differences in approach in a manner profitable to both.

The decision to use a steel casting, or any other material, begins with a study of design parameters. By a rough process of elimination based on experience and reference to handbooks, the designer separates obviously unsuited materials from those that show some merit. Promising survivors of the preliminary screening are now ready for more detailed evaluation. Assume for this evaluation that a design study has reached the stage where material selection has been narrowed to a low cost, high strength, easily obtainable steel casting.

## MATERIALS AND PROCESSES SELECTION

The first hurdles to be cleared are specifications. The philosophy of missile specifications is entirely different than that of some consumer goods specifications. A designer of household appliances may use specifications to protect himself from excessive consumer dissatisfaction. He will be no more restrictive than competition demands. He is not sure exactly how the product will be used, maintained or serviced; therefore, guarantees are vague. An automobile buyer does not expect to get with his purchase a set of shakedown test documents, functional test reports, statements of guaranteed specific fuel consumption, acceleration performance, a mathematical statement of reliability or much of any other technical data—nor does he necessarily need it.

Missile producers generally are bound by contract to a high degree of control of all parts, including steel castings. Government specifications must be used wherever possible, but if a material or process not covered by such specifications will do the job better, it may be employed provided that approved safeguards are applied. In the case of one of the author's company's missiles, substantial supporting test data must be obtained, approved and maintained to qualify non-government standards and specifications. It can be appreciated that the designer will be most conservative in departing from the use of Government specifications.

### *Specifications and Requirements*

Some major requirements and the specifications that govern them are given in Table 1. These specifications represent but a fraction of those actually in use, but they serve to illustrate some points. First, all are out of date from 5 to 20 years. Too few modern materials are available. Properties listed are inadequate and unrealistic in light of today's technology. For example, the stated tensile strength for 410 type steel is only 95,000 psi. Carbon in the same steel is allowed to range from 5 to 15 points, more than can be tolerated if close control of properties at the 200,000 strength level is desired.

A second point is that available specifications do not adequately define quality in terms of nondestructive tests. X-ray standards specified in MIL-C-6021 are A.S.T.M. : E71 films developed two decades ago

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TABLE 1 — REQUIREMENTS AND SPECIFICATIONS FOR CASTINGS

| Requirements | Specification | Title |
|---|---|---|
| Materials | QQ-S-681 | Federal specification for steel; castings. |
| | AMS 5353 | Aeronautical material specification, steel castings, precision investment, corrosion resistant, 17 Cr-4Ni-4Cu. |
| | AMS 5350 | Aeronautical material specification, steel castings, precision investment, corrosion resistant, 12.5 Cr (S.A.E. 60410). |
| Classification and Inspection | MIL-C-6021 | Military specification, castings, classification and inspection. |
| Radiographic Inspection | MIL-I-6865 | Military specification, inspection, radiographic. |
| Penetrant Inspection | MIL-I-6866 | Military specification, inspection, penetrant method. |
| Magnetic Inspection | MIL-I-6868 | Military specification, inspection, process, magnetic particle. |

for large thick-walled marine and machinery castings. The tightest controls provided by this specification for sand spots and inclusions will pass flaws 0.35 in. in diameter. Can missile designers be expected to accept such quality in parts having 0.1 in. walls?

Situations not covered by specification must be handled by means of notes on the engineering drawing. Familiar to all are the notes for casting structural classification, surface finish dimensional tolerance and other obvious information. Less familiar, but equally important, drawing requirements and the reasons for them merit close attention. A few examples will serve to illustrate their importance:

1. A drawing specifies type 410 alloy steel with a range of tensile strength of 180,000 to 210,000 psi and with 6 per cent minimum elongation. Heat treatment shall be the responsibility of the foundry. The question often asked is why limit the upper strength? The reason is that machining will be difficult; broaching and drilling may be impossible.
2. Then, one might ask, why require the foundry to provide the heat treatment? The answer is that the alloy is so sensitive to variations in chemistry that heat treatment must be tailored to each heat lot. The foundry is in the best position to know the metallurgical history and its implications.
3. A closer control on chemistry, particularly the carbon and the chromium, would simplify heat treatment. Could not the designer specify chemistry on his drawing, or could not the manufacturer of missiles issue his specifications delineating his needs? Attempts in these directions meet with the objection that too many conflicting requirements are being circulated. Also, sources of supply become limited.

These examples demonstrate that a little detail, such as whether or not to limit the upper strength level, can have a number of implications. The responsible designer is obliged to critically analyze the effect of his choice of specifications and drawing notes on every aspect of the casting. He will seek the assistance of his stress analyst, the materials and processes engineer, tooling and production experts and foundry representatives, as required. His final effort will be the best compromise available considering the factors involved.

### *Designer Viewpoint Acceptance*

In the experience of the writer, the one outstanding contributing factor to the unenthusiastic approach to the use of steel castings has been the failure of some foundries to accept the viewpoint of the designer. Perhaps it is the fault of the designer that his purposes are not always clear. In either case, the designer favorably rates foundries more or less in direct proportion to their performance in areas which can be listed as:

1. The foundryman in the missile business obtains, reads and assures himself that he understands every word of each specification and note appearing on the drawing or purchase order.
2. He does not assume that what he thinks is a better, cheaper or more sensible way of doing a job will satisfy the particular needs of his client.
3. He asks questions and offers suggestions before bidding a job, and he exercises restraint in making or seeking to make changes after an order has been secured.

These remarks would appear to be self evident, but failure to heed them recently resulted in the cancellation of five orders for steel castings. The decision to machine the parts from bars not only cost the foundry industry current opportunities to demonstrate its capabilities, but it most likely has prejudiced some designers concerning future use of castings.

## GEOMETRY REQUIREMENTS

The words that appear on drawings have been discussed; now the drawn lines that determine the shapes of parts shall be considered. Two conditions determine shape—first, the relation of the casting to surrounding parts, the second, the inherent requirements of weight, loads, castability and anything else that might restrict the form.

In the first instance, the designer is concerned with fitting parts together and with tolerances, warpage, draft or other characteristics that affect the fit. Examples of the types of parts that may be joined to castings are given in Table 2, which indicates some of the variables that can influence fit. Specifications, design practice and the preferences of the designer

TABLE 2 — TYPES OF TOLERANCES FOR SOME TYPICAL PARTS

| Part | Types of Tolerances |
|---|---|
| Extrusions | Linear, angular, twist, straightness. |
| Joggles | Linear, angular. |
| Sheet metal | Linear, flatness. |
| Machined parts | Close tolerances of all types. |
| Formed shapes | Linear, angular, twist, flatness, straightness. |
| Formed shapes | Linear, angular, twist, flatness, straightness. |
| Fasteners | Linear, angular. |

determine the amount of tolerance permitted. Understanding of casting design goes beyond knowledge of good foundry techniques; it includes the whole scope of fabrication and assembly of structures and mechanisms. A review of several common situations in some degree demonstrates how interdependent these factors can become.

Figure 1 illustrates typical aerodynamic smoothness requirements. Note that wave, bow, step and thickness tolerances are involved in uniting the parts. In such instances as these, either close-tolerance sheet may be obtained by selection at the mill, or the thickness of the skins may be controlled by grinding or chemical milling. Precision castings compatible with such rigid standards can be produced only when all conditions are ideal. If one suspects any possibility that a precision cast surface will have to be improved by machining, a cheaper sand casting with liberal cleanup tolerances should be specified.

Another problem results when blind fasteners are used. Washers cannot be installed to compensate for thread interference. Figure 2 shows a cross-section representative of a missile part. The nut is an N.A.S. 697 type with 0.062 in. minimum counterbore. Next to the nut is a casting with a wall thickness at the minimum side of a 0.11 to 0.13 in. range. A chemically milled skin joined to the casting is at the low end of its tolerance of 0.112 to 0.127 in. The ¼-28 thread N.A.S. 517-4-4 screw has a 0.235 grip, the maximum allowed. The head of the screw is recessed 0.005 in., also the maximum permitted. Clearly, any additional opening of tolerances would create drastic interference of the threads and damage to the nut plate, which must remain operable for repeated use. Similar reasoning will show that thick castings and sheet can lead to insufficient thread engagement and a weakened structure.

### *Foundry Missile Considerations*

Every missile casting should be treated by the foundry as a potentially special situation. Most of them are. Three degrees of draft on castings when added to two degrees angular tolerance allowed on extruded and formed angles make an excessive total for most applications. Bulges in cast surfaces mated to bows in sheet material parts makes sealing of the joint difficult. A minimum gap is required. Channel sections that are excessively opened or closed cannot be fitted to formed channels of high strength steel. Joggles have a depth tolerance of ±0.010 in.

Castings have to be compatible. Head gap on a ¼-in. bolt can be 0.004 in. maximum, and it may not exceed 40 per cent of the circumference. Cast surfaces under such heads must be smooth and flat. No more than two threads of a fastener can be in bearing, nor can more than 25 per cent of the gage of the fastened material be in bearing against the threads. No thread may be in bearing in walls less than 0.094 in. Casting walls have to be controlled accordingly. The foundry can take no chance on deviating from the drawing. Every effort should be made to resolve all questions before accepting a job.

It is not the purpose here to analyze all the dimensioning systems known. However, the system used at the author's company will be outlined with the objective of showing how ambiguities can be minimized. Castings have been dimensioned from the center line, the parting line, the big boss and from a host of equally elusive markers. No such index provides a needle-sharp point to which every inspector and machinist alike can put their gages and tools. Figure 3 represents an inspection fixture based on the tool point concept of dimensioning.

The salient feature of the system is that six reference points, never more, are used to establish three data planes in space. The number one plane is determined by three tool points, which are shown as buttons approximately parallel to a granite surface plate. When a tool point is offset from the plane, as the point at the extreme right, step dimensions define the amount of displacement. Three other points establish the two remaining planes, as shown. All three data planes are mutually perpendicular. Each plane must lie on at least one tool point.

### *Casting Inspection*

The casting shown below the fixture is marked with an (X) at each place where a tool point will pick up the part. The fixture does not know and it does not care whether the X's fall on bumps or hollows of the casting; it will pick up the surface whatever its shape, providing it is large enough. Inspection is accomplished by establishing the plane from a tool point or a reference button, as in Fig. 4, and checking distances to other key points, as in Fig. 5. A hollow spot in this casting at the step dimension would cause the part to fall under gage and be rejected.

Likewise, a thin end section or a downward bow would result in rejection. A bump at the tool point location in effect makes the casting appear too thick. Since the same tool points locate the part for machining, this system assures that proper machining stock will be available. No system of dimensioning that depends on juggling the part, trial and error fits to establish centerlines or use of cast surfaces themselves for data planes give any confidence that successive measurements will be in agreement.

## ASSEMBLY PROBLEMS

One excellent way to review design problems is to seek them out in the shops. Figure 6 shows two high strength steel castings used in the ramjet strut of the previously mentioned missile. Starting with the forward connector casting (Fig. 7—right side) one can see that aerodynamic smoothness will be determined by overall width of the part, the accuracy of the part, the accuracy of the angle of its walls and its straightness and flatness. Stiff steel channels with short joggles must fit the channel sections of the casting.

Thin walls and blind fasteners are combined. Bow in the web would fit poorly to the clips for longitudinal stiffening members. Reduction in wall thickness cannot be permitted for strength reasons; and thickness must not be increased beyond the bounds of weight.

Similar problems can be seen in the aft connector casting (Fig. 7—left side). Fitting the spikes of this

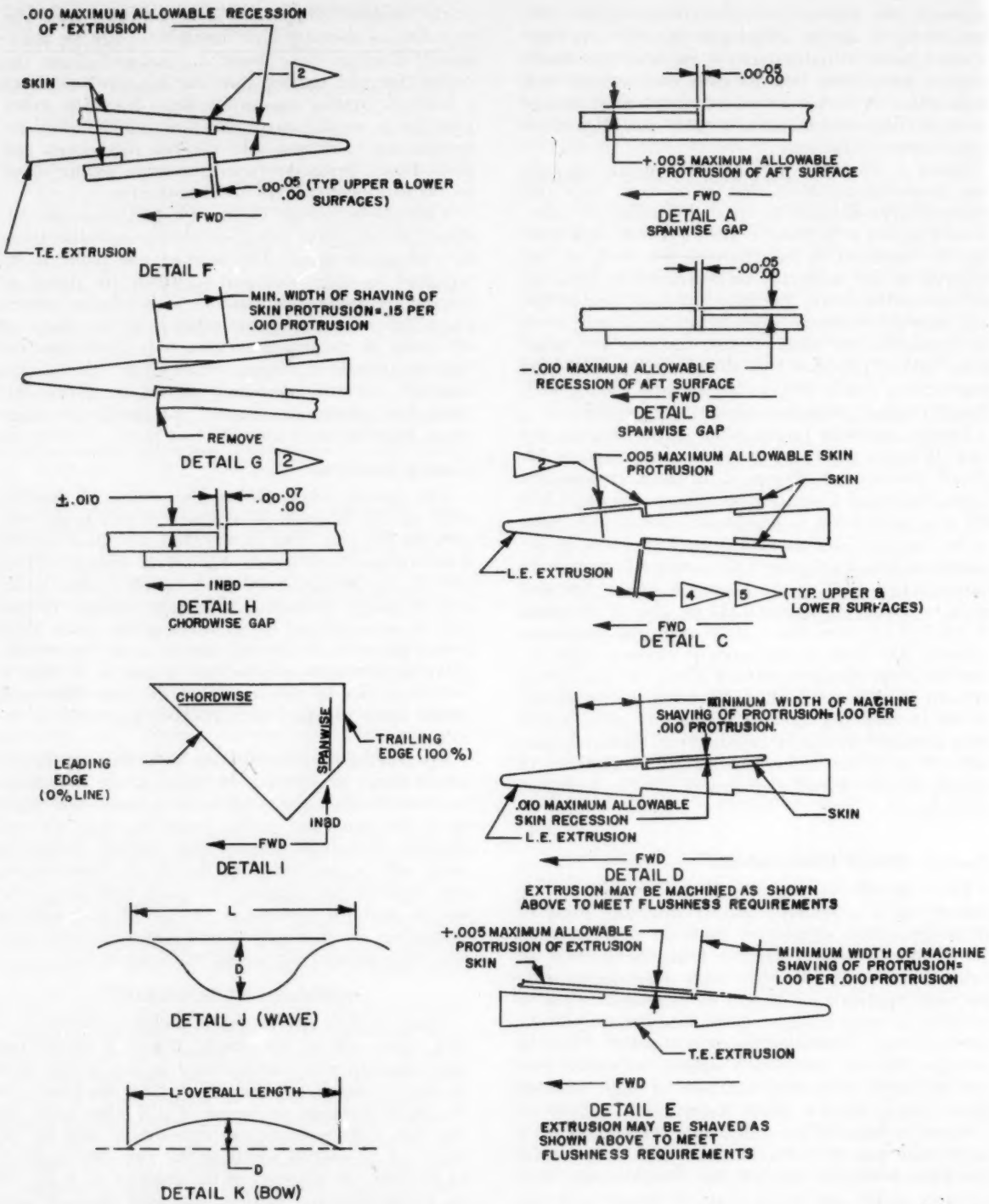
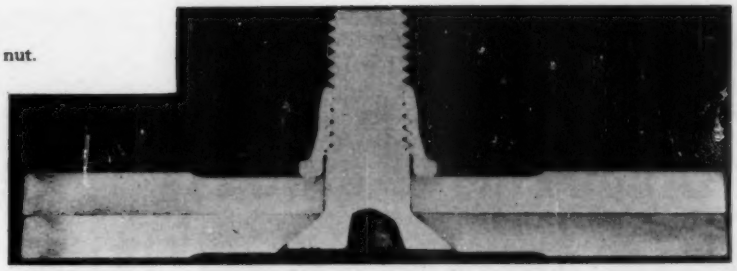



Fig. 1 — Aerodynamic smoothness and skin fastener flushness requirements.

part to the tapered channel requires close work. Bolting the casting to the extension of the bulkhead necessitates a reasonably flat back on the channel. The cast steps which receive the tabs of the sheet metal member must be closely held. One can easily recognize the pattern of rivets that hold down the nut plates. These fall close to edges of the casting, which must be accurately held to prevent breakout. Metal soundness in these areas is also critical. Up or down twist in the spikes would displace the

Fig. 2 — Thread interference in a blind plate nut.



*Data for Fig. 1*

### AERODYNAMIC SMOOTHNESS REQUIREMENTS

1. Spanwise joints in wing and tail surfaces are to be flush within $^{+0.005}_{-0.010}$ joint gap to be $0.00\ ^{0.05}_{0.00}$, see details *A* and *B*.
2. Chordwise skin joints to be flush within ±0.010 joint gap to be $0.00\ ^{0.07}_{0.00}$, see detail *H*.
3. Joints between leading edge extrusion and leading edge of skin and/or trailing edge extrusion and trailing edge of skin to be flush within $^{+0.005}_{-0.010}$, see details *C* through *F*, inclusive.
4. Leading edge, chordwise and spanwise contour tolerance:
   - A. Leading edge bow: The bow shall not exceed a maximum true depth of; d bow = (0.004) (L). See detail *K*.
   - B. Leading edge wave: The true depth of leading edge waves shall not exceed a maximum of; d wave = (0.004) (L). See detail *J*.
   - C. Chordwise contour: The chordwise contour may bow and/or wave by the following maximum limits. See details *J* and *K*.

| L | d |
|---|---|
| 1 | 0.020 |
| 25 | 0.080 |
| 100 | 0.120 |

   - D. Spanwise contour along per-cent line: The spanwise contour may bow and/or wave by the following maximum limits. See details *J* and *K*.

| L | d |
|---|---|
| 1 | 0.060 |
| 25 | 0.240 |
| 100 | 0.358 |

### SKIN FASTENER FLUSHNESS REQUIREMENTS

A. Rivet per BAC 5004 except flushness tolerances shall be in compliance with the following:

1. Hi-shear rivets and huck lockbolts installed flush $^{+0.002}_{-0.005}$.
2. All aluminum rivets (including blind rivets) shall be shaved flush ±0.002.
3. Jo-bolts in wing and aileron installed flush $^{+0.002}_{-0.010}$ except; jo-bolts through upper and lower skin and 55-1612 wing spar chord from wing STA 21.88 inboard shall be installed flush $^{+0.002}_{-0.005}$.

   Jo-bolts in fin, rudder, elevator stub and elevator installed flush $^{+0.002}_{-0.005}$.

B. Bolt and nut installations per BAC 5009 except flushness tolerances shall be in compliance with the following:

1. Bolts and screws installed flush $^{+0.005}_{-0.005}$

centerline of the whole strut, with pronounced effect on flight characteristics.

## CONCLUSION

The final goal of the designer is to achieve reliability. Complexity is one of the several factors influencing reliability. From the curves in Fig. 8 one may determine the relation of system reliability to complexity. It can be shown that a system made up of nine identical parts in series, which are 95 per cent reliable, can be expected to demonstrate a system reliability of 63 per cent. If a system is made up of 21 of the same 95 per cent reliable parts, the

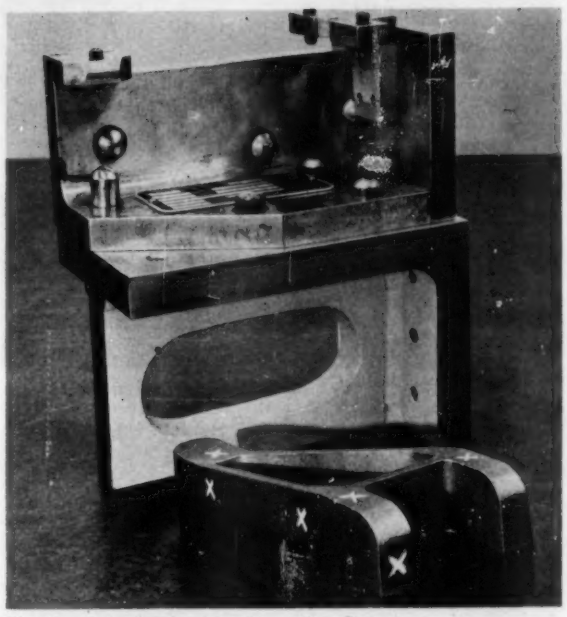


Fig. 3 — A tool point type inspection fixture.

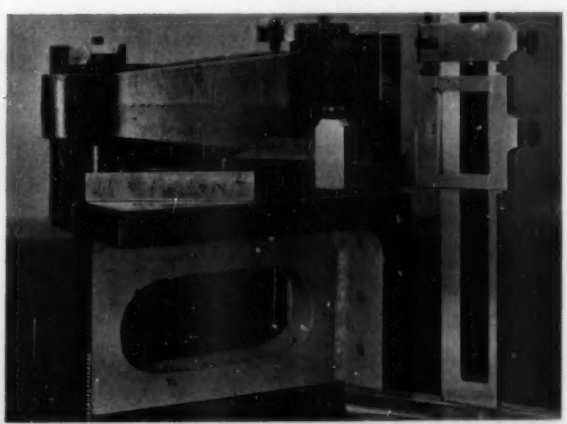


Fig. 4 — Casting setup on inspection fixture with gage on one of the data planes.

system can be expected to be 34 per cent reliable. Conversely, to maintain the original nine-part system reliability of 63 per cent, a 21 part system must have a part reliability of 98 per cent per part instead of the original 95 per cent per part, to maintain system reliability at the same level. Design simplification is the key to reliability.

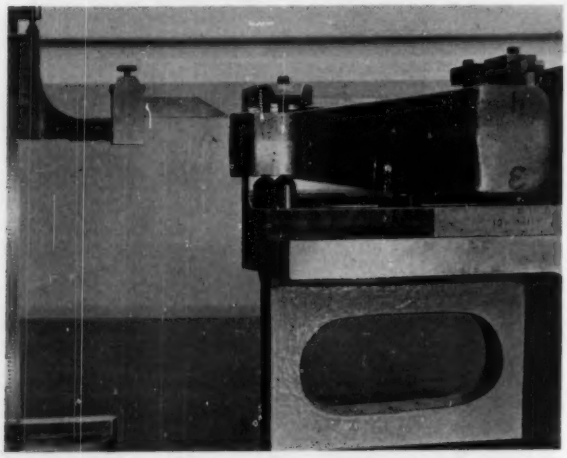


Fig. 5 — Checking of the casting from one of the data planes.

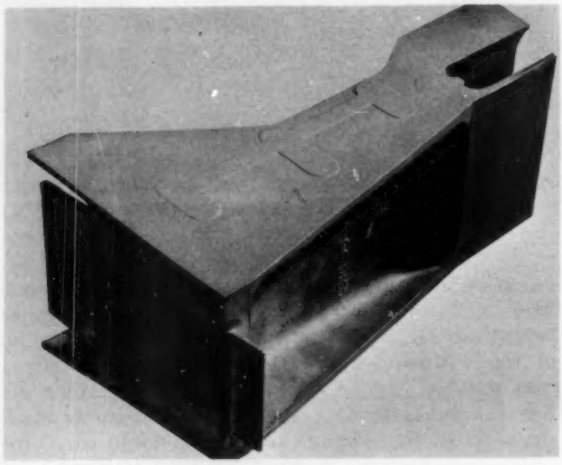


Fig. 6a — Typical missile casting.

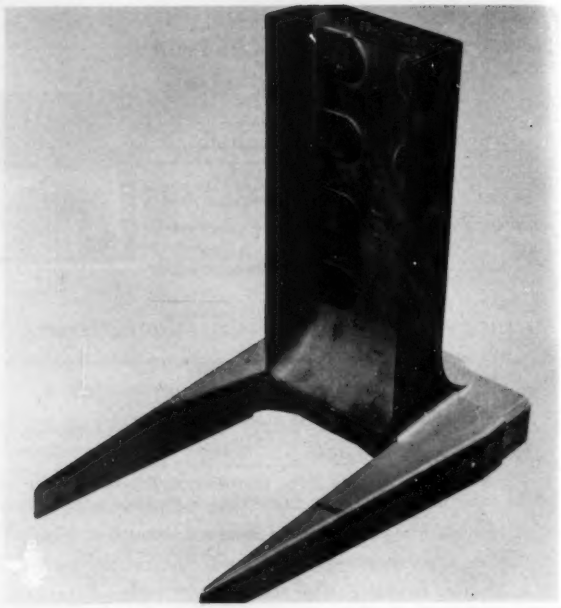


Fig. 6b — Typical missile casting.

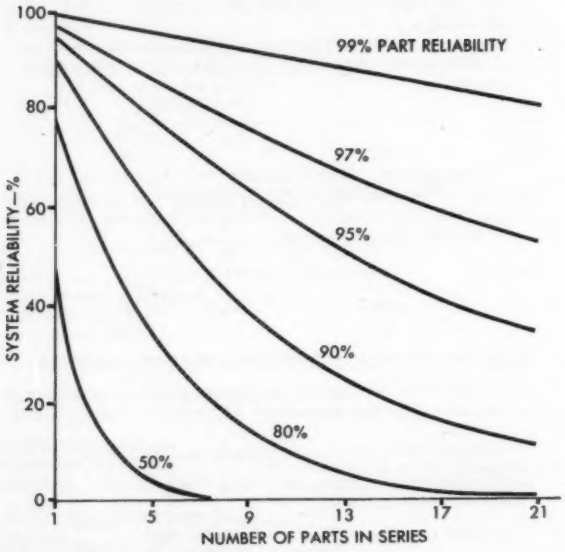



Fig. 8 — System reliability vs. complexity.

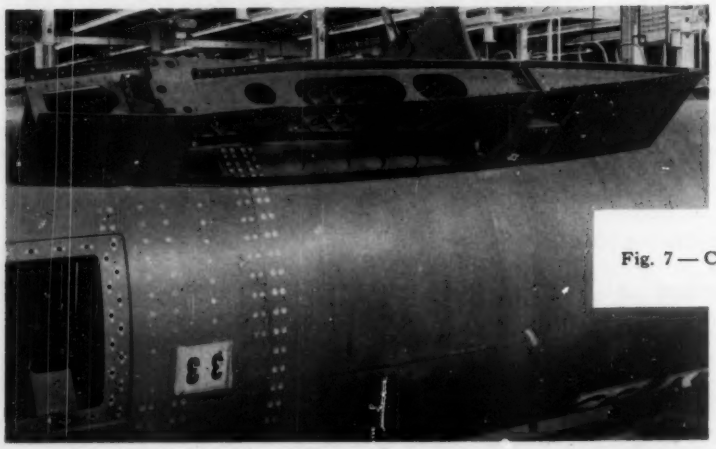


Fig. 7 — Castings installed in a ramjet strut assembly.

# DIRECT READING SPECTROGRAPH ANALYSIS OF CASTING METAL

by E. G. Mang


## ABSTRACT

**Many advantages have been realized by the author's company from the use of the direct reading spectrograph. Alloy savings, reduced scrap and better end product are some of these results. Details of the program, as used at this company, are presented, giving necessary requirements the spectrograph had to fulfill, description of the unit and calibration and analysis of the equipment as used in their program.**


## INTRODUCTION

Within the last 15 years or so, spectrographic methods of analysis have been used to supplement and, in some cases, replace the traditional wet chemical methods of analysis of steel. The author's company acquired their first spectrograph in 1951, and a second improved model in 1958. These have been used to great advantage in cutting analysis costs and in quality control.

Spectrographic methods using photographic technique have been used for many years as routine, in many steel laboratories, for more rigid controls in the manufacturing processes. In the last ten years it became apparent that still more rapid and economical methods of analysis were desirable. Customer specifications have become more rigid and are continuing to become even more so. In photographic spectography most of the time of an analysis is taken in processing the photographic plate and in measuring line densities. In addition, the accuracy of results is affected by plate characteristics, errors in processing, and density measurements.

The first instrument acquired was one using the photographic process, and was suitable for the analysis of low alloy steel and residual elements. It soon became apparent that equipment was needed to extend the spectrographic technique to the analysis of stainless steel. Also needed was a way to increase the speed of analysis so that it could be used in production control, i.e., to get an analysis of the molten steel before tapping so that adjustments could be made if necessary.

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## REQUIREMENTS

The production of both low alloy and stainless steel at the author's company is carried out in electric furnaces. The metal is melted under a definite analysis control. It is evident that an ideal method for controlling the end product accurately would be to run a complete analysis of the molten bath as soon as the melt down period is completed. This would supply the metallurgist or melter with the information necessary to make a complete and accurate adjustment to obtain the desired final analysis. To attempt such a preliminary analysis by either wet methods or the photographic spectrographic process would require an exorbitant amount of holding time in the furnace and make it impractical and prohibitive.

It was with such a preliminary analysis program in mind that the company started looking for a direct reading spectrograph. The necessary criteria for such an instrument were, accuracy, speed and being able to analyze for the desired elements, and the respective ranges of these, in both low alloys and stainless steel.

## INSTRUMENT AND AUXILIARIES

The development of improved excitation units, multiplier photographic tubes with a high amplification factor and improved grating spectrographs, have made it possible to design a unit for accurate, rapid analysis. In principle the photographic plate is replaced by a number of multiplier photographic tubes, placed behind the focal plane of the spectrometer. On the focal plane are corresponding secondary exit slits through which the selected element beam is isolated and directed to the cathode of the photographic tube.

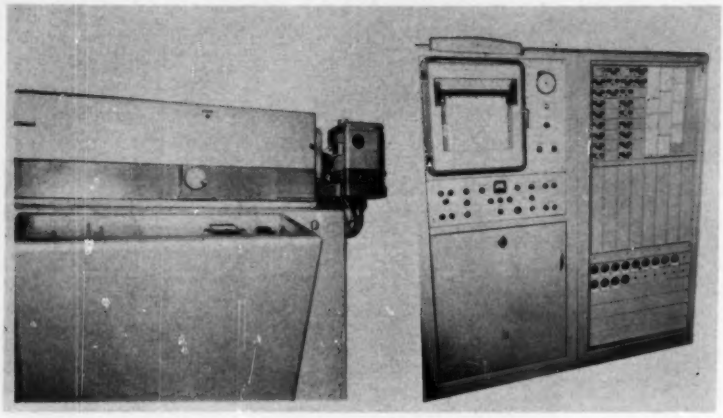


Fig. 1 — *Left* — power source (bottom) and spectrometer (top). *Right* — recorder.

The outputs of the photographic tube are integrated for about 20 sec, the photographic currents being allowed to charge condensers in the recording unit and subsequent measurement of the charge on each condenser.

The spectrograph is housed in a separate room in the laboratory. This room has a separate heating and air conditioning unit. The temperature is maintained at 74 F ± 1 F with a relative humidity of 40 ± 5 per cent. This rigid control of temperature and humidity is absolutely necessary to minimize the amount of adjustment and calibration needed during operation. It also reduces contaminating fumes and dust and thereby lessens the necessity for maintenance work on the equipment.

The spectrograph consists of the source unit, the spectrometer and the recording console (Fig. 1).

### *Source Unit*

This unit provides high precision excitation conditions varying from spark-like to arc-like discharges, as required by the analytical program. At the present time, only the spark discharge is being used in the company's analysis. Later other elements such as boron could be added to the program. Boron does require an arc-like source. The power to the unit is controlled by a constant voltage transformer which gives the necessary stability to the source. The cabinet which houses the source also provides the base for the spectrometer, thus making a compact unit requiring a minimum of space.

The spectrometer cabinet is constructed of a box-type aluminum weldment, horizontally mounted on the source unit. It contains the 1.5 m spectrometer, the diffraction grating, photographic multiplier tubes, mirrors and slit assembly. The diffraction grating used in this instrument is ruled 24,400 lines/in., and has a dispersion of 6.95 Å per mm in the first order and 3.5 Å per mm in the second order. The primary or entrance slit is 50 microns wide and the secondary (exit) slits vary in width up to 150 microns. The secondary slits are semifixed on a slit frame, and can be adjusted so that the maximum amount of light from the selected spectral lines will be focused by the mirrors on the phototube cathodes.

The excitation stand is mounted on the spectrometer case and remains perfectly aligned at all times. The electrode supports are water cooled so that there will be no adverse heating effects even when the instrument is used for continued long periods.

The console contains the integrators, the measuring circuits for the photo currents, the high and low voltage power supply, the attenuators controlling the photographic tube sensitivity, the electrometer amplifier and a recorder. Included also in this unit are the timing and switching mechanisms and the zero and sensitivity controls for each channel. The integrators each have two matched condensers with a high dielectric constant. For analysis only one of these is used, but both are required for calibration.

## CALIBRATION AND ANALYSIS

All spectrographic analyses are based on standard samples. Metal for the standard samples is melted in a small induction furnace. The compositions of the standards cover the necessary percentage range for all the elements. These are then analyzed carefully in the chemical laboratory. The values obtained are used in calibrating the instrument. At first samples from regular production heats which had varying percentage ranges were used as standards, but it was found more satisfactory to actually make up a small heat as described previously. In that way metal is obtained which has analysis much closer to the desired ranges on all elements.

In actual calibration two standards having concentrations near to the upper and lower limits of the ranges for a particular alloy are selected. Using specified excitation conditions an integration is carried out to charge one of the two integrator condensers for the low value, and a second integration to charge the other condenser with the high value. By using these charges which represent a known element concentration, each channel is adjusted by using the zero and sensitivity controls to give a definite low and high reading on the recorder.

These low and high readings correspond to two points on working curves or graphs which are prepared for conversion of recorder readings to percent-

age concentrations on unknown samples. When all the channels have been adjusted to give the desired readings the unknown sample is sparked and the recorder readings noted (usually the average of two or even three readings is taken). The readings are then converted to percentage composition from the working curves.

## APPLICATION

Each direct reader is designed to meet the needs of a particular analytical program. The company started with one group of channels. These included manganese, silicon, chromium, nickel, molybdenum, copper, vanadium and cobalt. The instrument was used for low alloy and for stainless types CF8 (304) and CF8M (316) at first. As other stainless alloys were added to the program, it was found that to recalibrate the instrument for each one required an excessive amount of time.

To overcome this, a second group of channels was added so that now it is possible to have two calibrations on at the same time and switch from one to the other. The *A* group is used for stainless steels, and the *B* group for low alloys. There is still some conflict at times with samples of different alloy type coming in for analysis at the same time. To overcome this a third group may be added to cut down on the holding times in the furnaces.

In actual practice a sample is brought to the laboratory after melt down for a preliminary analysis. Adjustments to the melt are made as necessary and the heat is poured. If major adjustments are required, a second preliminary analysis is often made before actually pouring the castings. A final sample is taken from the ladle and analyzed to give the final heat analysis. At the present time, between 250 and 350 analyses are being made each day. This number will be increased as more groups and more element channels are added to these groups. The large number of different alloys that are analyzed necessarily cuts down the actual number of separate element analyses because of the time required for recalibration for each alloy.

## COMPARATIVE RESULTS

Reproducibility and accuracy of analysis depends a great deal on samples and sampling procedure. The standards that are used for calibration should be of the same type and metallurgical background as the samples that are being analyzed. The sampling procedure should be as consistent as possible and of such a nature to give a representative, homogeneous sample. In the literature of spectrographic analysis there is some difference of opinion as to the best type of sample to give the highest accuracy and reproducibility.

Although the electrode support stand is water cooled, the size of the sample should be such that there is a minimum effect from heating while the sample is being sparked. Also there should be sufficient area for several excitations without repolishing. Another requisite for a good sample is the grain structure. A fine dendritic grain gives more reproducible results. After some trials with samples cast in sand molds and glass tubes, a sample was developed which gives good results and does not require an excessive preparation time.

The mold for this sample is made of a solid piece of copper about 5⅛-in. in diameter and 3¾-in. high. A recess is machined in one end to pour the sample of approximately the following dimensions – 2⅞-in. diameter at the bottom tapered to 3¼-in. at the top and 1 1/16-in. deep. In the bottom a rise is machined tapering from ½-in. to 1½-in. and ½-in. high (Fig. 2). This gives a sample of sufficient size for about six excitations without heating effect.

The copper causes fast chilling thus giving small dendritic grain structure. The depression in the center concentrates the shrink area away from the surface that will be sparked. Figure 3 shows the grain structure and grain orientation of a sample. Note the fine grain prependicular to the surface to be sparked. The small grain here insures that the area sparked will cover several grains. By this, accuracy and reproducibility are assured. Also note the location of the shrink area.

This sample is set up in a lathe and approximately 0.035 in. machined from the face which will be sparked. This face is finished on a disc sander using 120 grit $Al_2O_3$ paper. This gives us a sample which is absolutely flat with a good finish. Sufficient space is available to spark five or six times without excessive heating effects. The time required to prepare this sample is not more than 3 or 4 min. It has been found that by using this type of sample for standards, as well as for the regular sample, excellent reproducibility is obtained.

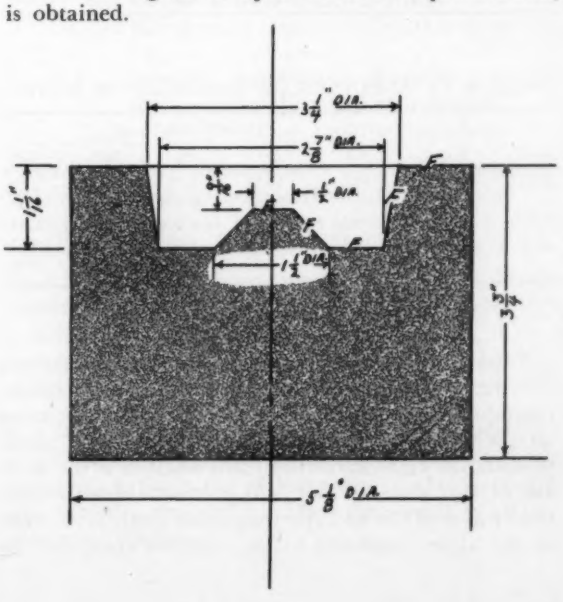



Fig. 2 — Copper mold used for spectrographic analysis.

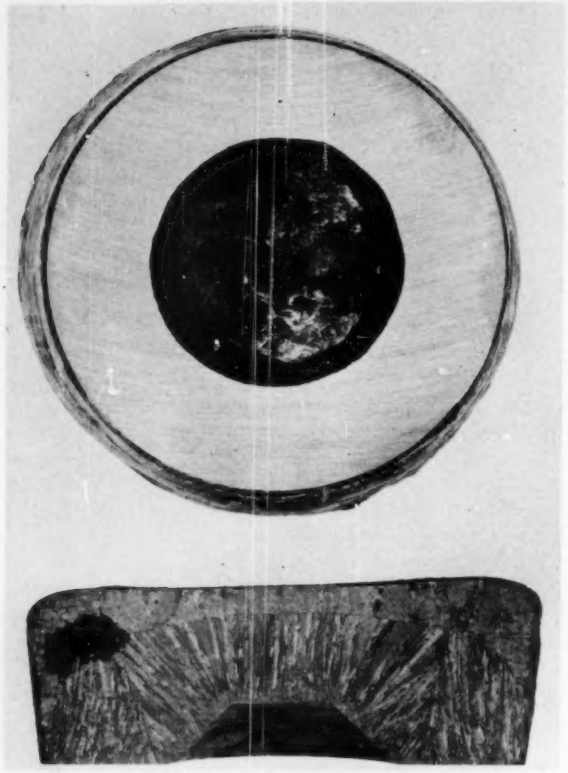


Fig. 3 — Spectrographic sample. Sample in bottom illustration has been etched to show grain structure.

TABLE 1 — CONSECUTIVE ANALYSES

| Heat | Type | Analysis, % Cr | Ni | Si | Mn | Mo |
|---|---|---|---|---|---|---|
| 6626 | CF8M | 19.60 | 10.10 | 1.55 | 0.85 | 2.55 |
| | | 19.60 | 10.20 | 1.50 | 0.83 | 2.53 |
| | | 19.65 | 10.20 | 1.55 | 0.85 | 2.55 |
| | | 19.55 | 10.25 | 1.50 | 0.83 | 2.53 |
| | | 19.60 | 10.15 | 1.55 | 0.83 | 2.55 |
| | | 19.55 | 10.20 | 1.55 | 0.83 | 2.55 |

TABLE 2 — ANALYSES ON CONSECUTIVE DAYS

| Heat | Type | Analysis, % Cr | Ni | Si | Mn | Mo |
|---|---|---|---|---|---|---|
| 6626 | CF8M | 19.60 | 10.20 | 1.55 | 0.85 | 2.55 |
| | | 19.75 | 10.30 | 1.55 | 0.83 | 2.60 |
| | | 19.55 | 10.15 | 1.50 | 0.83 | 2.53 |
| | | 19.75 | 10.10 | 1.55 | 0.85 | 2.55 |
| | | 19.60 | 10.25 | 1.50 | 0.85 | 2.55 |

Table 1 gives the results of six consecutive analyses proving the sample to be homogeneous and the results reproducible. Table 2 gives the results of the same sample analyzed on consecutive days for a week, again proving the reproducibility (each analysis is the average of two shots). Table 3 is a comparison between chemical analysis and spectrographic analysis of some of the alloys produced at the author's company. By taking the average of two or three shots the results compare well with wet chemical methods.

TABLE 3 — METHOD OF ANALYSIS COMPARISON

| Heat | Type | Analysis, % Cr | Ni | Si | Mn | Mo | Method* |
|---|---|---|---|---|---|---|---|
| 6626 | CF8M | 19.60 | 10.20 | 1.55 | 0.85 | 2.55 | S |
| | | 19.70 | 10.33 | 1.59 | 0.82 | 2.62 | C |
| 8282 | CF8 | 19.38 | 9.35 | 1.28 | 0.85 | | S |
| | | 19.42 | 9.18 | 1.34 | 0.84 | | C |
| 4129 | HH | 24.60 | 13.20 | 1.25 | 1.16 | | S |
| | | 24.48 | 13.38 | 1.27 | 1.25 | | C |
| 6745 | CA15 | 12.40 | 0.88 | 0.54 | 0.79 | | S |
| | | 12.32 | 0.84 | 0.49 | 0.82 | | C |

*S — Spectrographic analysis.
C — Chemical analysis.

## PRODUCTION CONTROL

The direct reader has been in use for about 2½ years, and it has been a valuable tool in quality control.

Only two of the most common alloys CF8 (304) and CF8M (316) were analyzed at first, but now it has been extended to cover most of the iron-base alloys produced. A preliminary analysis is made of the melt, and adjustments are made either at the furnace or in the ladle, to bring the heat to the desired chemical analysis. Then a sample is analyzed from the ladle to give the final heat analysis.

The time required to bring a sample from the furnace to the laboratory, prepare and spark it and phone the results back to the melter is about 10 min. This is not excessive, and has paid off in fewer off analysis heats and a better end product.

## CONCLUSIONS

The advantages obtained by the use of the spectrograph are:

1. Relieved the analytical load on the chemical section of the laboratory.
2. Reduced the cost of analysis.
3. Increased the number of analyses that can be made.
4. Decreased the time necessary for accurate calculation of each heat because adjustments can be made after the preliminary analysis.
5. Eliminated the necessity for holding the returns from each heat separate and marked so that heats could be calculated accurately using known returns. This has resulted in a saving in the time required to segregate the returns from each heat, put them in a separate container, and store them for future use.
6. Resulted in a saving of pure alloy additions.
7. Less scrap due to off analysis metal.
8. Enabled the company to accept orders with narrow chemical range specifications which could not have been controlled without a preliminary analysis.
9. Resulted in a better end product, and thus more satisfied customers.

# EXTERNAL CORNER HEAT EXTRACTION

by K. H. Coats


## ABSTRACT

Solidification of metal castings presents a variety of problems including calculation of the rate of heat extraction from the casting by the mold. An important problem in this category is prediction of the rate of heat extraction by a sand mold from a solidifying casting (or part thereof) of rectangular cross-section. The corners of this rectangular cross-section are termed "external" corners.[1] This problem can be treated by numerical solution of the partial differential equation governing unsteady state heat conduction in the infinite planar region surrounding an interior rectangle at constant temperature. No known tools of analytical mathematics are capable of yielding the solution in this geometry. Thus attempts to treat this problem to date have been largely experimental and empirical.

The purpose of this paper is presentation of the results of the numerical solution of the heat equation in the geometry described in the first paragraph. The partial differential equation was replaced by a finite-difference form and solved step-wise on a computer. A dimensionless form of the rate of heat extraction at the casting-mold interface was calculated and numerically integrated with respect to time to yield the dimensionless cumulative quantity of heat extracted as a function of dimensionless time. The results are presented graphically and in tables; they can be used in conjunction with known mold thermal properties and casting dimensions to yield cumulative heat extraction as a function of time or, equivalently, the time of solidification.

Mathematical treatment of heat extraction by sand molds is simplified by the assumption of a virtually constant temperature at the mold-casting interface. Adams and Taylor[2] and Ruddle[1] point out that available data justify this assumption when pure metals or near-eutectic alloys are involved. This assumption should be invalid, however, in cases involving the sand casting of alloys having a wide solidification range. A method is therefore given herein for the treatment of cases where the interfacial temperature varies with time in any manner. The method is applicable to heat extraction from castings of linear, cylindrical and spherical shapes as well as the rectangular shape considered here. Ohira[3] has given an expression for the temperature distribution in a sand mold for the case where the interfacial temperature is a quadratic function of time.

The use of the tabulated dimensionless heat extraction quantity for sand castings of rectangular cross-section is illustrated by example problems. Mention is made of the application of this tabulated quantity to problems in unsteady-state mass diffusion and fluid flow through porous media since the heat equation solved in this study applies equally well to these phenomena.


## PREVIOUS STUDIES

Heat extraction by sand molds has been treated mathematically and experimentally by many authors. The reader is referred to chapter 4 of Ruddle[1] for a more or less complete coverage of the work done in this area. Chvorinov[4] proposed use of an equation applicable to linear heat flow for the calculation of heat extraction in a variety of mold (casting) geometries. Adams and Taylor[2] correctly pointed out that the linear flow solution does not apply to cylindrical and spherical shapes, and gave the different equations relating heat extraction to time for these three geometries. Ruddle and Minchner[5] and other authors[6,7,8] have obtained experimental data pertinent to heat extraction by sand molds.

These data support in general the assumption of a constant mold-casting interface temperature when pure metals and near-eutectic alloys are cast. Ruddle and Skinner[9] treated the problem of heat flow at external corners by integrating the areas between experimentally obtained isotherms in sand molds. They concluded that the difference in cumulative heat extraction by an external corner and by a plane surface of equal perimeter can be represented as a straight-line function of time. The results presented in this paper can be shown to roughly justify their conclusion.

## ENGINEERING CONSIDERATIONS

The rectangle shown in Fig. 1 represents a cross-section of a casting, and the infinite region exterior to the rectangle represents a sand mold. The partial differential equation governing heat conduction in this infinite exterior region has been numerically solved in this study to yield the equation giving cumulative heat extraction $Q$ from the casting (rectangle) as a function of time:

$$Q = \rho \, chab \, (T_S - T_O) \, \overline{Q}_R \qquad (1)$$

where

$\rho, c$ = density, specific heat of sand.
$h$ = height of rectangular section.
$a, b$ = half-sides of rectangular cross-section (Fig. 1).
$T_S$ = constant temperature of mold-casting interface $\cong$ solidification temperature.
$T_O$ = original uniform mold temperature.

The term $\overline{Q}_R$ is a dimensionless heat extraction quantity for rectangular geometry which has been calculated, as described in the Appendix, and is tabulated as a function of dimensionless time in the Table. $\overline{Q}_R$ is also plotted versus time in Fig. 2. The dimensionless time employed here for rectangular heat flow is denoted by $t_R$, where

$$t_R = 0.84 \frac{Kt}{\rho \, ca^2} \qquad (2)$$

$K$ is effective thermal conductivity of the mold and $t$ is real time. A parameter in the tabulation at $\overline{Q}_R$ versus $t_R$ is the ratio of the sides of the rectangle, a/b, denoted here by $L$. Thus the case $L = 1.0$ applies to a square.

K. H. COATS is with University of Michigan, Ann Arbor.



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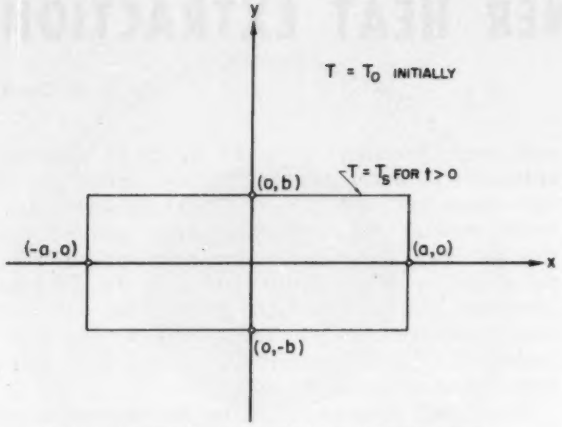



Fig. 1 — Rectangle surrounded by infinite plane.

The applicability of equation (1) is limited by the following conditions:

a) Infinite mold.
b) Constant interface temperature.
c) Two-dimensional heat flow.

These conditions are not as restrictive as they may appear. The practical definition of an infinite mold is one which is sufficiently large or of low enough thermal conductivity that the temperature of the exterior mold surface does not rise appreciably during the solidification of the casting. In most actual sand casting operations the effect of the mold exterior surface on heat extraction from the casting is negligible, and assumption of an infinite mold for calculation purposes introduces negligible error. Chvorinov[4] claims that if the mold wall is at least half the casting thickness the outer mold surface can rise to 500 C (932 F) in steel casting with little effect on heat extraction by the mold.

The restriction of a constant interfacial temperature is unnecessary and is eliminated by use of equation (17). The requirement of two dimensional heat flow simply means that no heat is assumed to be extracted normal to the plane of the cross-section shown in Fig. 1. In cases where heat loss from the riser and/or bottom of a rectangular casting is appreciable use of equation (1) or (17) will yield maximum times of solidification. Adams and Taylor[2] treat heat loss from risers by convection and radiation.

Equation (1) can be combined with the usual heat balance[2] to yield solidification times for rectangular sections. If $\Delta H$ is the specific heat of fusion of the cast metal then at the time of complete solidification

DIMENSIONLESS EFFLUX $\bar{Q}_R$

| Dimensionless Time, $t_R$ | L = 1.0 | L = 1.25 | L = 1.5 | L = 1.75 | L = 2.0 |
|---|---|---|---|---|---|
| 0.00040 | 0.19703 | 0.22166 | 0.24629 | 0.27092 | 0.29555 |
| 0.00080 | 0.27865 | 0.31348 | 0.34831 | 0.38314 | 0.41797 |
| 0.00120 | 0.34127 | 0.38393 | 0.42659 | 0.46925 | 0.51191 |
| 0.00160 | 0.39407 | 0.44333 | 0.49259 | 0.54184 | 0.59110 |
| 0.00200 | 0.44058 | 0.49565 | 0.55073 | 0.60585 | 0.66387 |
| 0.00400 | 0.62308 | 0.70355 | 0.79363 | 0.88130 | 0.96871 |
| 0.00600 | 0.77609 | 0.88308 | 0.99675 | 1.10486 | 1.21035 |
| 0.00800 | 0.91348 | 1.04096 | 1.17201 | 1.29527 | 1.41480 |
| 0.01000 | 1.03805 | 1.18204 | 1.32702 | 1.46289 | 1.59467 |
| 0.02000 | 1.53102 | 1.72986 | 1.92494 | 2.11023 | 2.29184 |
| 0.03000 | 1.90078 | 2.13946 | 2.37446 | 2.60011 | 2.82244 |
| 0.04000 | 2.20712 | 2.48069 | 2.75096 | 3.01219 | 3.27033 |
| 0.05000 | 2.47477 | 2.78023 | 3.08269 | 3.37634 | 3.66703 |
| 0.06000 | 2.71600 | 3.05116 | 3.38356 | 3.70731 | 4.02820 |

*(continued on next page)*

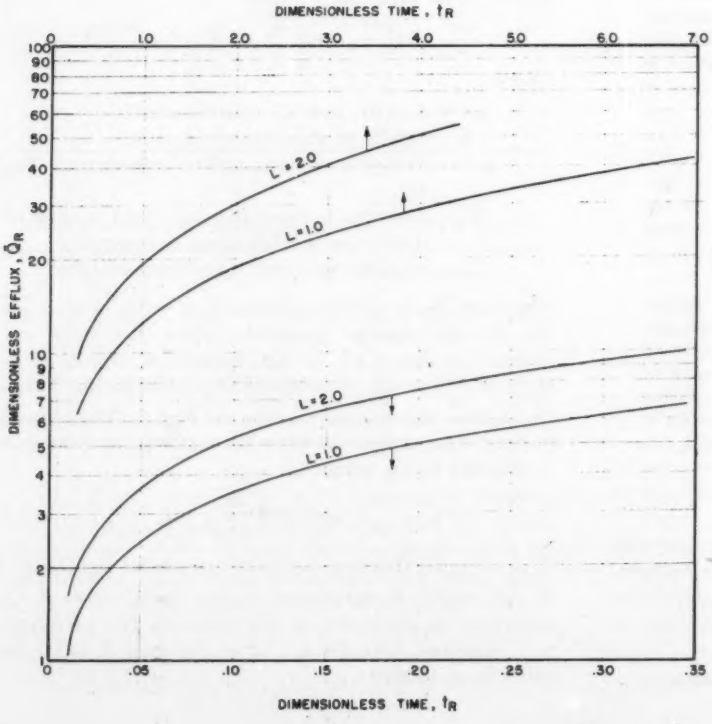



Fig. 2 — Dimensionless efflux $Q_R$ vs. dimensionless time $t_R$.

DIMENSIONLESS EFFLUX $\overline{Q}_R$

| Dimensionless Time, $t_R$ | L = 1.0 | L = 1.25 | L = 1.5 | L = 1.75 | L = 2.0 |
|---|---|---|---|---|---|
| 0.07000 | 2.93777 | 3.30095 | 3.66157 | 4.01363 | 4.36291 |
| 0.08000 | 3.14449 | 3.53434 | 3.92176 | 4.30071 | 4.67692 |
| 0.09000 | 3.33912 | 3.75451 | 4.16759 | 4.57225 | 4.97420 |
| 0.10000 | 3.52379 | 3.96377 | 4.40153 | 4.83090 | 5.25757 |
| 0.11000 | 3.70006 | 4.16382 | 4.62540 | 5.07863 | 5.52916 |
| 0.12000 | 3.86913 | 4.35594 | 4.84063 | 5.31697 | 5.79059 |
| 0.13000 | 4.03194 | 4.54117 | 5.04831 | 5.54710 | 6.04316 |
| 0.14000 | 4.18923 | 4.72033 | 5.24935 | 5.77000 | 6.28789 |
| 0.15000 | 4.34164 | 4.89408 | 5.44446 | 5.98645 | 6.52564 |
| 0.16000 | 4.48966 | 5.06299 | 5.63426 | 6.19711 | 6.75711 |
| 0.17000 | 4.63371 | 5.22752 | 5.81924 | 6.40253 | 6.98290 |
| 0.18000 | 4.77417 | 5.38806 | 5.99986 | 6.60317 | 7.20351 |
| 0.19000 | 4.91135 | 5.54497 | 6.17647 | 6.79944 | 7.41938 |
| 0.20000 | 5.04550 | 5.69852 | 6.34940 | 6.99169 | 7.63088 |
| 0.22000 | 5.30563 | 5.99657 | 6.68528 | 7.36527 | 8.04204 |
| 0.24000 | 5.55614 | 6.28392 | 7.00937 | 7.72597 | 8.43922 |
| 0.26000 | 5.79825 | 6.56193 | 7.32318 | 8.07541 | 8.82417 |
| 0.28000 | 6.03295 | 6.83170 | 7.62788 | 8.41489 | 9.19829 |
| 0.30000 | 6.26104 | 7.09412 | 7.92447 | 8.74549 | 9.56273 |
| 0.32000 | 6.48321 | 7.34992 | 8.21375 | 9.06807 | 9.91844 |
| 0.34000 | 6.70000 | 7.59974 | 8.49640 | 9.38341 | 10.26622 |
| 0.36000 | 6.91191 | 7.84410 | 8.77303 | 9.69212 | 10.60676 |
| 0.38000 | 7.11934 | 8.08345 | 9.04412 | 9.99476 | 10.94063 |
| 0.40000 | 7.32264 | 8.31820 | 9.31011 | 10.29180 | 11.26835 |
| 0.45000 | 7.81471 | 8.88691 | 9.95495 | 11.01219 | 12.06312 |
| 0.50000 | 8.28682 | 9.43332 | 10.57512 | 11.70538 | 12.82797 |
| 0.55000 | 8.74197 | 9.96071 | 11.17421 | 12.37521 | 13.56710 |
| 0.60000 | 9.18245 | 10.47164 | 11.75499 | 13.02467 | 14.28388 |
| 0.65000 | 9.61007 | 10.96816 | 12.31967 | 13.65618 | 14.98108 |
| 0.70000 | 10.02631 | 11.45187 | 12.87003 | 14.27172 | 15.66102 |
| 0.75000 | 10.43237 | 11.92414 | 13.40752 | 14.87295 | 16.32568 |
| 0.80000 | 10.82926 | 12.38608 | 13.93337 | 15.46128 | 16.97672 |
| 0.85000 | 11.21782 | 12.83861 | 14.44862 | 16.03793 | 17.61560 |
| 0.90000 | 11.59878 | 13.28255 | 14.95417 | 16.60397 | 18.24359 |
| 0.95000 | 11.97276 | 13.71859 | 15.45079 | 17.16032 | 18.86178 |
| 1.00000 | 12.34031 | 14.14731 | 15.93916 | 17.70780 | 19.47113 |

DIMENSIONLESS EFFLUX $\overline{Q}_R$

| Dimensionless Time, $t_R$ | L = 1.0 | L = 1.25 | L = 1.5 | L = 1.75 | L = 2.0 |
|---|---|---|---|---|---|
| 1.05000 | 12.70190 | 14.56924 | 16.41988 | 18.24716 | 20.07249 |
| 1.10000 | 13.05794 | 14.98485 | 16.89351 | 18.77905 | 20.66662 |
| 1.15000 | 13.40880 | 15.39453 | 17.36052 | 19.30405 | 21.25418 |
| 1.20000 | 13.75481 | 15.79866 | 17.82136 | 19.82270 | 21.83575 |
| 1.25000 | 14.09627 | 16.19756 | 18.27641 | 20.33547 | 22.41187 |
| 1.30000 | 14.43342 | 16.59153 | 18.72605 | 20.84280 | 22.98299 |
| 1.35000 | 14.76651 | 16.98084 | 19.17060 | 21.34508 | 23.54954 |
| 1.40000 | 15.09575 | 17.36574 | 19.61037 | 21.84266 | 24.11189 |
| 1.60000 | 16.37590 | 18.86285 | 21.32392 | 23.78905 | 26.32257 |
| 1.80000 | 17.60805 | 20.30599 | 22.98224 | 25.68516 | 28.49173 |
| 2.00000 | 18.79929 | 21.70369 | 24.59563 | 27.54208 | 30.62988 |
| 2.20000 | 19.95526 | 23.06276 | 26.17213 | 29.36806 | 32.74460 |
| 2.40000 | 21.08043 | 24.38879 | 27.71807 | 31.16932 | 34.84143 |
| 2.60000 | 22.17850 | 25.68637 | 29.23852 | 32.95063 | 36.92454 |
| 2.80000 | 23.25260 | 26.95938 | 30.73758 | 34.71569 | 38.99714 |
| 3.00000 | 24.30543 | 28.21111 | 32.21860 | 36.46744 | 41.06176 |
| 3.20000 | 25.33932 | 29.44435 | 33.68435 | 38.20820 | 43.12045 |
| 3.40000 | 26.35631 | 30.66152 | 35.13712 | 39.93987 | 45.17487 |
| 3.60000 | 27.35819 | 31.86467 | 36.57883 | 41.66398 | 47.22642 |
| 3.80000 | 28.34657 | 33.05561 | 38.01110 | 43.38178 | 49.27628 |
| 4.00000 | 29.32285 | 34.23589 | 39.43529 | 45.09431 | 51.32550 |
| 4.20000 | 30.28829 | 35.40685 | 40.85257 | 46.80243 | 53.37498 |
| 4.40000 | 31.24401 | 36.56968 | 42.26392 | 48.50685 | 55.42556 |
| 4.60000 | 32.19103 | 37.72540 | 43.67018 | 50.20815 | |
| 4.80000 | 33.13023 | 38.87489 | 45.07206 | 51.90684 | |
| 5.00000 | 34.06241 | 40.01895 | 46.47018 | 53.60332 | |
| 5.20000 | 34.98831 | 41.15824 | 47.86507 | | |
| 5.40000 | 35.90856 | 42.29336 | 49.25718 | | |
| 5.60000 | 36.82374 | 43.42484 | 50.64688 | | |
| 5.80000 | 37.73437 | 44.55313 | 52.03453 | | |
| 6.00000 | 38.64092 | 45.67862 | 53.42040 | | |
| 6.20000 | 39.54380 | 46.80167 | | | |
| 6.40000 | 40.44339 | 47.92257 | | | |
| 6.60000 | 41.34003 | 49.04160 | | | |
| 6.80000 | 42.23401 | 50.15898 | | | |
| 7.00000 | 43.12562 | 51.27493 | | | |

the cumulative heat extraction must equal (2a) (2b) $h\,\rho_m\,[\Delta H + c_m\,(T_p - T_m)]$, where a slight superheat is accounted for by the difference $(T_p - T_m)$ between pouring and solidification temperatures. Thus

$$Q = \rho\,chab\,(T_S - T_O)\,\overline{Q}_R$$
$$= 4\,abh\,\rho_m\,[\Delta H + c_m\,(T_p - T_m)]$$

or

$$\overline{Q}_R = \frac{4\,\rho_m}{\rho c\,(T_S - T_O)}[\Delta H + c_m\,(T_p - T_m)] \qquad (3)$$

In an actual case the right-hand side of (3) can be readily calculated, and the value of dimensionless time $t_R$ corresponding to $\overline{Q}_R$ can be found from the Table. The solidification time is then calculated in equation (2) as

$$t = \frac{\rho\,ca^2}{0.84K}\,t_R$$

### *Example Problem 1*

The rectangular cross-section of a magnesium sand casting is roughly 15 cm x 10 cm. Assuming an interfacial temperature $T_S$ equal to the solidification temperature of 650 C (1202 F), calculate the solidification time for a 30 C superheat. Data for the metal are

$\rho_m = 1.61$ g./cc.
$c_m = 0.319$ cal./g. − C
$\Delta H = 88$ cal./g.
$T_m = 650$ C (1202 F).

Data for the sand mold are

$\rho = 1.46$ g./cc.
$c = 0.251$ cal./g. − C
$K = 0.00075$ cal./cm. − C − sec.
$T_O = 22$ C.

### *Solution*

Following the assumption $T_S = T_m = 650$ C (1202 F) and equation (3) becomes

$$\overline{Q}_R = \frac{4\,(1.61)}{1.46\,(0.251)\,(650 - 22)}\,[88 + 0.319\,(30)]$$

$$\overline{Q}_R = 2.73$$

The casting dimensions give $L = \frac{a}{b} = \frac{15/2}{10/2} = 1.5$

and, from the Table, at L = 1.5, $\overline{Q}_R = 2.73$

$$t_R = 0.0395$$

and

$$t = \frac{\rho\,ca^2}{0.84K}\,t_R = \frac{1.46\,(0.251)\,(15/2)^2}{0.84\,(0.00075)}\,0.0395$$

$$t = 1300\,\text{sec} = 21.6\,\text{min}$$

Note that a must always be taken as the longer half-side of the rectangle since $\overline{Q}_R$ is tabulated only for values of $L$ (a/b) greater than 1.0.

As mentioned, Adams and Taylor[2] presented equations giving cumulative heat extraction as a function of time for linear, cylindrical and spherical geometries. The rectangular case may now be added to these equations to obtain:

$$\frac{Q_L}{A} = \frac{2}{\sqrt{\pi}} \rho c (T_s - T_o) \bar{Q}_L \quad (4)$$

$$\frac{Q_R}{A} = \frac{\rho c a}{4(1+L)} (T_s - T_o) \bar{Q}_R \quad (5)$$

$$\frac{Q_C}{A} = \rho c R_C (T_s - T_o) \bar{Q}_C \quad (6)$$

$$\frac{Q_S}{A} = \rho c R_S (T_s - T_o) \bar{Q}_S \quad (7)$$

Subscripts L, R, C, S here denote linear, rectangular, cylindrical and spherical geometries, respectively. Equation (5) is obtained by dividing equation (1) by area of a rectangular casting, $(4a + 4b)\ h$, and replacing $b$ by a/L. Equations (4), (6) and (7) are equivalent to those given by Adams and Taylor under the definitions:

$$\bar{Q}_L^* = \sqrt{t_L} \quad (8)$$

$$\bar{Q}_C = 0.5\, t_C + \frac{2}{\sqrt{\pi}} \sqrt{t_C} \quad (9)$$

$$\bar{Q}_S = t_S + \frac{2\sqrt{t_S}}{\sqrt{\pi}} \quad (10)$$

where

$$t_L = \frac{Kt}{\rho c} \quad (11)$$

$$t_C = \frac{Kt}{\rho c R_C^2} \quad (12)$$

and

$$t_S = \frac{Kt}{\rho c R_S^2} \quad (13)$$

Equations (4) - (13) facilitate comparison between heat extractive power of molds of different geometries. For equal areas a spherical geometry yields the greatest rate of heat extraction followed by cylindrical and linear geometries in that order. A comparison between the cylindrical and rectangular cases shows that the rate of heat extraction in the latter geometry lies between the linear and cylindrical. Division of equation (5) by equation (6) gives, for equal areas,

$$\frac{Q_R}{Q_C} = \frac{a}{4 R_C (1+L)} \frac{\bar{Q}_R}{\bar{Q}_C} \quad (14)$$

For equal areas, $R_C = \frac{2}{\pi}(a+b) = \frac{2a}{\pi}\left(\frac{L+1}{L}\right)$ so that (14) becomes

$$\frac{Q_R}{Q_C} = \frac{\pi L}{8(L+1)^2} \frac{\bar{Q}_R}{\bar{Q}_C} \quad (15)$$

and this ratio is less than 1.0 for any values of $L$ and time $t$ greater than zero. The ratio $Q_R/Q_L$ is, of course, greater than 1.0 for all $L$ and time.

**$\bar{Q}_L$ is dimensional with unit of length, while $\bar{Q}_R$, $\bar{Q}_C$ and $\bar{Q}_S$ are dimensionless.*

### *Time-Variant Interfacial Temperature*

The case of a time-variant interfacial temperature can be treated by application of Duhamel's superposition theorem[10] to the constant temperature solutions (4) - (7). The general form of these latter equations is

$$Q = Z A (T_s - T_o) \bar{Q} \quad (16)$$

where $Z$ is a coefficient differing for each geometry. Use of the superposition theorem gives

$$Q_j = \frac{Z}{2} A \sum_{i=0}^{i=j-1} \Delta T_i \bar{Q}_{j-i} \quad (17)$$

where

$\Delta T_i = T_{s_{i+1}} - T_{s_{i-1}}, i \neq 0.$

$\Delta T_0 = T_{s_0} + T_{s_1} - 2\, T_o.$

$T_{s_i}$ = interfacial temperature at time $t = i\Delta t$.

$\Delta t$ = time increment chosen to represent continuously changing $T_s$ by a step function.

$\bar{Q}_{j-i}$ = value of $\bar{Q}$ at dimensionless time corresponding to real time of $(j - i)\Delta t$.

$Q_j$ = cumulative heat extraction by mold at time $t = j\Delta t$.

The approximation of a continuously changing interfacial temperature $T_S$ is illustrated in Fig. 3. Equation (17) becomes more accurate as the time increment chosen for this approximation is taken smaller. Equations (4) - (7) now become for the variable temperature case,

$$\frac{Q_{Lj}}{A} = \frac{\rho c}{\pi} \sum_{i=0}^{i=j-1} \Delta T_i \bar{Q}_{L_{j-i}} \quad (18)$$

$$\frac{Q_{Rj}}{A} = \frac{\rho c a}{8(1+L)} \sum_{i=0}^{i=j-1} \Delta T_i \bar{Q}_{R_{j-i}} \quad (19)$$

$$\frac{Q_C}{A} = \frac{\rho c R_C}{2} \sum_{i=0}^{i=j-1} \Delta T_i \bar{Q}_{C_{j-i}} \quad (20)$$

$$\frac{Q_S}{A} = \frac{\rho c R_S}{2} \sum_{i=0}^{i=j-1} \Delta T_i \bar{Q}_{S_{j-i}} \quad (21)$$

In most practical cases involving the sand casting of alloys the interfacial temperature $T_S$ is not known a priori as a function of time. This in no way lessens the usefulness of equations (18) - (21), provided a relationship between Q and $T_S$ is known. Different relationships will hold between Q and $T_S$ (assuming here that at any time the metal temperature is substantially uniform and equal to the interfacial temperature $T_S$) in the three temperature ranges of superheat, solidification and sub-cooling. In the superheat range, the heat balance gives at any time

$$Q = c_m (T_p - T_S) W \quad (22)$$

where $W$ is the casting weight. In the two-phase solidification range

$$Q = c_m (T_p - T_{SI}) W + W \alpha (T_S) \Delta H \quad (23)$$

where $T_{SI}$ is the liquidus temperature and $\alpha(T_S)$ is

a function of temperature ranging from zero at the liquidus to 1.0 at the solidus temperature. The term $\Delta H$ is the total heat evolution/lb cooled from the liquidus temperature $T_{sl}$ to solidus $T_{ss}$. The function $\alpha(T_S)$ can be determined rather simply for a binary noneutectic alloy from the phase diagram and from knowledge of the specific heat in the two-phase range.

In the sub-cooled range

$$Q = c_m (T_P - T_{sl}) W + W \Delta H + c_m (T_{ss} - T_s) W \quad (25)$$

The heat capacity $c_m$ may obviously be allowed to vary with temperature and/or differ from the liquid and solid; only minor modification of (22), (23) and (25) is required in this case. Combination of equations (22), (23) and (25) with the appropriate one of (18) - (21) allows prediction of solidification times for alloys.

### *Example Problem 2*

A magnesium alloy is poured with negligible superheat into a sand mold at an initial uniform temperature of 22 C. The casting is essentially a rectangular cylinder, 15 cm x 10 cm. The alloy liquidus and solidus temperatures are 607 C and 482 C, respectively. The function $\alpha(T)$ in this range can be approximated by

$$\alpha = \frac{607 - T}{607 - 482}$$

Data for the sand mold are the same as given in example problem 1 while the alloy data are

$$\Delta H = 111 \text{ cal./g.}$$
$$c_m = 25 \text{ cal./g.} - \text{C.}$$
$$T_P \cong 607 \text{ C.}$$
$$\rho_m = 1.61 \text{ g./cc.}$$

### *Solution*

Since the casting is of rectangular cross-section, equation (19) applies. Combination of (19) and (23) in the heat balance gives

$$A \frac{\rho c a}{8 (1 + L)} \sum_{i=0}^{i=j-1} \Delta T_i \overline{Q}_{R_{j-i}} = W \alpha (T_{s_j}) \Delta H$$

or, upon substitution for area $A$ and weight $W$,

$$4bh (L + 1) \frac{\rho c a}{8 (L + 1)} \sum_{0}^{j-1} \Delta T_i \overline{Q}_{R_{j-i}} = 4abh \rho_m \Delta H \alpha (T_{s_j})$$

Cancellation and insertion of $L = \frac{a}{b} = 1.5$, $\Delta H = 111$, $\alpha = 0.008 (607 - T_{s_j})$ and other properties gives

$$\sum_{i=0}^{i=j-1} \Delta T_i \overline{Q}_{R_{j-i}} = 31.2 (607 - T_{s_j}) = 18{,}900 - 31.2 T_{s_j}$$

This equation is now written for j = 1, 2, 3, 4, . . . as:

$$\Delta T_o \overline{Q}_{R1} = (T_{s0} + T_{s1} - 2 T_o) \overline{Q}_{R1} = 18{,}900 - 31.2 T_{s1} \quad (26)$$

$$\Delta T_o \overline{Q}_{R2} + (T_{s2} + T_{s0}) \overline{Q}_{R1} = 18{,}900 - 31.2 T_{s2} \quad (27)$$

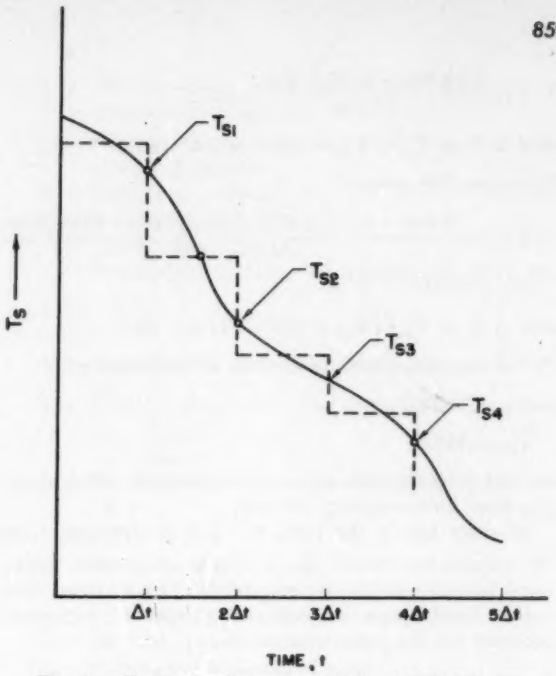



**Fig. 3 — Representation of variable boundary temperature by a step function.**

$$\Delta T_o \overline{Q}_{R3} + \Delta T_1 \overline{Q}_{R2} + (T_{s3} - T_{s1}) \overline{Q}_{R1} = 18{,}900 - 31.2 T_{s3} \quad (28)$$

$$\Delta T_o \overline{Q}_{R4} + \Delta T_1 \overline{Q}_{R3} + \Delta T_2 \overline{Q}_{R2} + (T_{s4} - T_{s2}) \overline{Q}_{R1} = 18{,}900 - 31.2 T_{s4} \quad (29)$$

These equations can be solved explicitly in the order given for $T_{s1}$, $T_{s2}$, $T_{s3}$, . . .

Let a time increment $\Delta t$ of 8 min be chosen. The corresponding dimensionless time increment is then, from equation (2)

$$\Delta t_R = 0.84 \frac{K}{\rho c a^2} \Delta t = 0.84 \frac{0.00075 (8) (60)}{1.46 (0.251) (15/2)^2}$$

$$\Delta t_R = 0.0147$$

From the Table, for L = 1.5

$$\overline{Q}_{R1} = \overline{Q}_R \text{ at } t_R \text{ of } 1 (0.0147) = 1.608$$
$$\overline{Q}_{R2} = \overline{Q}_R \text{ at } t_R \text{ of } 2 (0.0147) = 2.347$$
$$\overline{Q}_{R3} = \overline{Q}_R \text{ at } t_R \text{ of } 3 (0.0147) = 2.887$$
$$\overline{Q}_{R4} = \overline{Q}_R \text{ at } t_R \text{ of } 4 (0.0147) = 3.348$$
$$\overline{Q}_{R5} = \overline{Q}_R \text{ at } t_R \text{ of } 5 (0.0147) = 3.753$$

Rearrangement of (26) gives

$$T_{s1} = \frac{18{,}900 - \overline{Q}_{R1} (T_{s0} - 2 T_o)}{31.2 + \overline{Q}_{R1}}$$

$$T_{s1} = \frac{18{,}900 - 1.608 [607 - 2 (22)]}{31.2 + 1.608}$$

$$T_{s1} = 548 \text{ C.}$$

and $\Delta T_o = T_{s0} + T_{s1} - 2 T_o = 607 + 548 - 44 = 1111$ C

Rearrangement of (27) yields

$$T_{s2} = \frac{18{,}900 - \Delta T_o \overline{Q}_{R2} + T_{s0} \overline{Q}_{R1}}{31.2 + \overline{Q}_{R1}}$$

$$T_{s2} = \frac{18{,}900 - 2610 + 975}{32.808} = 526\,C$$

and $\Delta T_1 = T_{s2} - T_{s0} = 526 - 607 = -81.$

Equation (28) gives

$$T_{s3} = \frac{18{,}900 - 1111\,(2.887) + 81\,(2.347) + 548\,(1.608)}{31.2 + 1.608}$$

$$T_{s3} = 510\,C$$

and $\Delta T_2 = T_{s3} - T_{s1} = 510 - 548 = -38.$

Following the above procedure of calculation:

$T_{s4} = 498\,C$

$T_{s5} = 487\,C$

so that solidification should be complete after about 5½ time increments or 44 min.

In cases where the ratio $L = a/b$ is different from the values for which $\bar{Q}_R$ is tabulated in the Table, cross plotting might be employed. The simplest procedure in this case is, however, to run the computer program for the particular value of $L$ desired.

### *Unsteady State Liquid Flow*

Equations (1) and (19) also apply to the cases of unsteady state liquid flow through porous media and mass diffusion which obeys Fick's law. For the former case, Darcy's flow law for horizontal flow appears as

$$\vec{V} = -\frac{k}{\mu}\vec{\nabla} P \qquad (30)$$

where $\vec{V}$ is the fluid velocity vector, $k$ the permeability of the porous medium, $\mu$ the fluid viscosity and $\vec{\nabla} P$ the gradient $\left(\frac{\partial P}{\partial x}\vec{i} + \frac{\partial P}{\partial y}\vec{j}\right)$ of the pressure. Combination of (30) with the continuity equation and an equation relating liquid density to pressure yields[11] equation (31) for two dimensional flow.

$$\frac{k}{\mu \phi c}\left(\frac{\partial^2 P}{\partial x^2} + \frac{\partial^2 P}{\partial y^2}\right) = \frac{\partial P}{\partial t} \qquad (31)$$

The term $c$ in this equation is the liquid compressibility and $\phi$ is porosity of the porous medium. Except for nomenclature, equation (31) is identical to equation (38), the equation solved in this study. Thus, the cumulative efflux of water (in volume units) away from the boundary of an oil or gas reservoir of approximate rectangular shape is

$$W_e = h\,\phi\,c\,a\,b\,\Delta P\,\bar{Q}_R \qquad (32)$$

The reservoir is assumed, of course, to be situated in a thin water sand of infinite extent. The term $h$ becomes the water sand thickness in this case, and the designation "thin" water sand means that $h$ is small in comparison to the smaller of the sides of the rectangular reservoir (say $h/2b$ and $h/2a < 0.05$). The term $\Delta P$ in (32) is $P_b - P_0$ or reservoir pressure minus original aquifer pressure. Dimensionless time for the fluid flow case is

$$t_R = 0.84\,\frac{k\,t}{\mu\,\phi\,c\,a^2} \qquad (33)$$

For the case of diffusion of a component in a mixture a diffusion coefficient $D$ can often be found such that Fick's law holds.

$$\vec{n} = -D\vec{\nabla} C \qquad (34)$$

Here $\vec{n}$ is the mass flow vector (mass of component per unit area per unit time), $D$ is the diffusion coefficient and $\vec{\nabla} C$ the gradient of the component concentration. Combination of (34) with a continuity equation yields[12] equation (35) for two-dimensional diffusion.

$$D\left(\frac{\partial^2 C}{\partial x^2} + \frac{\partial^2 C}{\partial y^2}\right) = \frac{\partial C}{\partial t} \qquad (35)$$

Again, the similarity between (35) and (38) is obvious. The cumulative efflux of the diffusing component from the rectangle is given by

$$M = h\,a\,b\,\Delta C\,\bar{Q}_R \qquad (36)$$

where $M$ is cumulative efflux in mass units and $h$ is the thickness of the planar system. Equation (36) applies to the case of a component diffusing from a rectangular boundary at concentration $C_b$ into the infinite exterior region of initial concentration $C_0$. The term $\Delta C$ is equal to $C_b - C_0$. Dimensionless time for the diffusion case is given by

$$t_R = 0.84\,\frac{D\,t}{a^2} \qquad (37)$$

## APPENDIX

The geometry considered in this study is shown in Fig. 1. A rectangle of length 2a and width 2b is surrounded by the infinite plane at an initial uniform temperature equal to $T_0$. The temperature at the rectangular boundary is maintained at $T_s$ for time $> 0$. The solution corresponding to the case of a time-variant temperature at the rectangular boundary can be obtained from the solution for a constant boundary temperature, as shown. The cumulative quantity of heat extracted from the rectangle is desired as a function of time.

The partial differential equation governing two-dimensional heat flow appears as

$$\frac{K}{\rho c}\left(\frac{\partial^2 T}{\partial x^2} + \frac{\partial^2 T}{\partial y^2}\right) = \frac{\partial T}{\partial t} \qquad (38)$$

where $t$ is time and $T$ is temperature. This equation was solved numerically by a slightly modified version of the alternating-direction implicit method.[13] The infinite exterior region was transformed to a finite region by defining new spatial variables $u$, $v$, where

$$u = 1 - e^{-\frac{fx}{a}} \qquad (39)$$

$$v = 1 - e^{-\frac{fy}{b}} \qquad (40)$$

and $f$ is a constant scaling factor which simplifies the choice of a u-v grid. An $f$ value of 0.9163 was convenient in this study and therefore $t_R = f^2\,Kt/\rho c\,a^2 = 0.84\,Kt/\rho c\,a^2$.

The cumulative heat extraction $Q$ is obtained from Fourier's law of heat conduction. This law gives the heat fluxes in the x and y directions as

$$q_x = -K\frac{\partial T}{\partial x} \qquad (41)$$

$$q_y = -K\frac{\partial T}{\partial y} \qquad (42)$$

The total flux from the rectangle is obtained by integration (41) and (42) over the half-sides of the rectangle in the first quadrant and multiplying the result by 4.

$$q = -4K\left[\int_0^b \left(\frac{\partial T}{\partial x}\right)_{x=a} dy + \int_0^a \left(\frac{\partial T}{\partial y}\right)_{y=b} dx\right] \qquad (43)$$

The cumulative quantity of heat extracted from the rectangle at time $t$ is

$$Q = h\int_0^t q\,dt \qquad (44)$$

where $h$ is the thickness of the slab or plane and $Q$ is in units of heat energy. Substitution of $u$, $v$ for $x$, $y$ and $t_R$ for $t$ in the above two equations yields

$$Q = h\rho c\,a\,b\,(T_S - T_0)\,\overline{Q}_R$$

where

$$\overline{Q}_R = -\frac{4e^{-f}}{f^2}\int_0^{t_R}\left[\int_0^{1-e^{-f}}\left(\frac{\partial \overline{T}}{\partial u}\right)_{u=1-e^{-f}}\frac{dv}{1-v} + L^2\int_0^{1-e^{-f}}\left(\frac{\partial \overline{T}}{\partial v}\right)_{v=1-e^{-f}}\frac{du}{1-u}\right]d_{t_R} \qquad (45)$$

and $\overline{T} = \dfrac{T - T_0}{T_S - T_0}$

Substitution of $\overline{T}$, u, v, $t_R$ into (38) for T, x, y and t, respectively, gives an equation which was numerically solved to yield $\overline{T}$. These values of $\overline{T}$ were then employed to calculate the derivatives in (45), and Simpson's rule of numerical integration was employed to carry out the double integration in (45). The details of the solution are reported elsewhere.[16]

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## NOMENCLATURE

a, b = Half-sides of rectangular cross-section.
A = Area.
c = Specific heat of mold.
$c_m$ = Specific heat of metal.
d = Spatial increment $\Delta u$, $\Delta v$.
f = Scaling factor in transformation of x, y to u, v, 0.9163 in this study.
h = Thickness of plane or slab containing rectangle.
$\Delta H$ = Heat of fusion.
K = Thermal conductivity of mold.
L = Ratio of half-sides of rectangle, a/b.
Q = Cumulative heat extraction.
$\overline{Q}_L$ = Heat extraction quantity for linear geometry.
$\overline{Q}_C$ = Dimensionless heat extraction quantity for cylindrical geometry.
$\overline{Q}_R$ = Dimensionless heat extraction quantity for rectangular geometry.
$\overline{Q}_S$ = Dimensionless heat extraction quantity for spherical geometry.
$R_C$ = Radius of cylinder.
$R_S$ = Radius of sphere.
t = Time.
$t_C$ = Dimensionless time for cylindrical heat flow, $\dfrac{Kt}{\rho c R_C^2}$.
$t_L$ = Transformed time for linear heat flow $\dfrac{Kt}{\rho c}$.
$t_R$ = Dimensionless time for rectangular heat flow, $0.84\dfrac{Kt}{\rho c a^2}$.
$t_S$ = Dimensionless time for spherical heat flow, $\dfrac{Kt}{\rho c R_S^2}$.
T = Temperature.
$T_0$ = Initial mold temperature.
$T_S$ = Mold-casting interfacial temperature.
$T_{SL}$ = Liqidus temperature of alloy.
$T_{SS}$ = Solidus temperature of alloy.
$\overline{T}$ = Transformed temperature, $\dfrac{T - T_0}{T_S - T_0}$.
$T_p$ = Pouring temperature.
$T_m$ = Melting temperature of pure metal.
u, v = Transformed spatial coordinates.
W = Weight of casting.
x, y = Spatial coordinates.
$\rho$ = Density of mold.
$\rho_m$ = Density of metal.
$\mu$ = Viscosity of flowing fluid.

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# ALUMINUM BRONZES CONTAINING MANGANESE, NICKEL AND IRON

## *Chemical composition effect on structure and properties*

by J. O. Edwards and D. A. Whittaker


### ABSTRACT

The effect of changes in aluminum and manganese, and to a lesser extent iron and nickel contents, was investigated. Eighteen melts over a range of compositions were prepared by induction melting, and these were cast into standard keel block test bars, and into tapered wedge blocks representing castings up to 4 in. thick. The mechanical properties and metallography of these test castings was investigated, and related to the chemical analysis.

The alloys appear to be less sensitive to melting conditions than aluminum bronzes without manganese, and there was no undue melt loss due to oxidation of any of the constituents. All the alloys appeared to retain their properties reasonably well in thick sections, and the pouring temperature did not appear to be critical at least in thin sections. However, there was some indication that Type A alloy may have a higher optimum pouring temperature than Type B.

It was found that the mechanical properties and microstructure were closely related to both the aluminum and manganese contents, aluminum having the predominant effect. For better correlation of properties with analysis, a concept of "equivalent aluminum content" was developed similar to the "equivalent zinc content" of high tensile brass. This showed that manganese had most effect on yield strength, and least effect on elongation, and enabled the properties of any specific alloy to be predicted with some accuracy using a series of curves which were developed. This also allowed the manganese working range to be calculated for a given aluminum content in order to meet specified properties. It appeared that in conjunction with a hardness determination and a rapid bath analysis for aluminum and/or manganese, the developed curves could be used to give effective foundry control of large melts.

The mechanical test results indicate that the copper-aluminum-manganese alloys form a continuous series with properties which change progressively with changes in aluminum or manganese content. While this may not apply to such factors as corrosion-erosion resistance, metallography suggests that here too, changes should be progressive.

A limited study indicated that increasing the nickel content from 2-6 per cent, while at the same time increasing the iron content from 2-4 per cent, caused a serious reduction in ductility, accompanied by heavy precipitation within the grains. Higher iron (6 per cent) in combination with 2 per cent nickel gave normal ductilities. It is possible that the reduction in ductility apparently caused by the high nickel content could be counteracted by deliberately reducing the iron content.

Metallography showed that the proportion of alpha and decomposed beta was dependent on the equivalent aluminum content. The amount of coarse crystallites of intermetallic compound appeared to be related to the iron content, and it was deduced that these separated from the melt while it was still liquid. The physical metallurgy of the finer particles of intermetallic compounds dispersed through the structure is complex, although the amount of such particles appears to increase as the aluminum content increases.


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### INTRODUCTION

The cast aluminum bronzes containing substantial amounts of manganese have been in commercial use for about five years. These alloys possess desirable



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properties as propeller materials in that they apparently have high ultimate tensile strength, yield strength and elongation values, are highly resistant to impingement, cavitation and other types of corrosion and are somewhat lighter than the conventional high strength yellow brass (manganese bronze) propeller alloys. They are also reputed to have better foundry characteristics than aluminum bronzes of lower manganese content.

In making an assessment of these alloys, it was found that apart from the basic patents,[1,2] and a paper by Bradshaw et al,[3] there was relatively little information published on the material, although a number of articles[4,5] quoted specified and actual properties on individual melts. It was decided, therefore, to investigate the effect of variations of aluminum and manganese, and to a lesser extent iron and nickel contents, on the structure and tensile properties of the material.

## ALLOY COMPOSITION CHOICE

Two proprietary alloys to these approximate specification are used commercially:

| | Alloy A | Alloy B |
|---|---|---|
| Manganese, % | 6 - 8 | 11 - 14 |
| Aluminum, % | 7 - 10 | 7 - 9 |
| Nickel, % | 1 - 2 | 1 - 6 |
| Iron, % | 4 - 6 | 2 - 4 |
| Impurities, % | 0.5 max. | 0.5 max. |
| Copper, % | Rem. | Rem. |
| UTS, psi min. | | 90,000 |
| YS (0.5% Offset), psi min. | | 40,000 |
| Elong. in 2 in., % min. | | 20 |

In general, therefore, variations in aluminum and manganese content were kept within the ranges of the two alloys. The basic ingot used was to Alloy A specification, and to avoid confusion, the iron and nickel contents were kept at about 4.5 and 1.8 per cent, respectively, for both alloys in the initial melts. Later the iron and nickel contents were varied within the ranges of the specific alloys.

## MELTING PROCEDURE

Metal for propellers is invariably melted in large reverberatory furnaces, and the melting schedules are relatively long. It was considered impractical to duplicate this on a laboratory scale, and for convenience the high frequency lift-coil induction furnace was used. It was appreciated that this was by no means an ideal melting unit for this type of material, and previous work[6] had shown that oxide dispersions were formed in aluminum bronze melts, reducing ductility. However, it was considered that the relatively severe oxidation for a short period might simulate losses through preferential oxidation in the long time reverberatory melt.

The standard melting procedure used for 230 lb melts of both types of alloy was:

1. Ingot plus large scrap melted down in induction furnace.
2. Light scrap charged and the metal superheated to 30 C (55 F) above the desired pouring temperature.
3. Main alloy additions (manganese, aluminum) stirred in.
4. Superheated metal degassed with dry nitrogen for 15 min at 5 litres/min.
5. Superheat temperature checked and crucible removed from furnace.
6. Poured at required temperature.

No fluxes, melt covers or deoxidants were used with either alloy. The analyses of the various melts are given in Table 1.

### *Test Castings*

In propeller manufacture, the alloy must maintain its properties in thick sections with slow cooling rates. It is, therefore, desirable to assess the properties of massive test castings such as those of Bradshaw,[3] but with only limited melting facilities, this was impossible. Accordingly, from each melt two keel blocks of A.S.T.M. B208-54 design were cast together with two tapered wedge bars of a design previously used.[6] This latter bar simulates castings from ½-in. to 4 in. in section thickness, gated and risered for optimum soundness, and is shown in Fig. 1.

### *Pouring Temperature*

One melt each of Alloys A and B was poured at various temperatures into leg gated keel blocks. Table 2 indicates the composition for these alloys, and shows the resulting mechanical properties as an average of two test bars for each temperature. From these preliminary tests a pouring temperature of about 1200 C (2190 F) was selected as this temperature appeared to be in the middle of a range which gave reasonable properties for both alloys. These test castings and all others were left in the mold for 2 hr subsequent to casting, before shakeout.

## TEST RESULTS

Seven test bars were machined from each wedge block to represent section thicknesses of ½, ¾, 1, 1½, 2, 3 and 4 in., and two bars were machined from each standard keel block.

Tensile test results from these bars are shown in Table 3. Results from the tapered wedge sections are the average of two bars per melt unless otherwise stated, and for keel blocks, four bars were averaged. Brinell hardness tests were carried out on the fractured ends of the keel bars, and these are included in Table 3.

Graphical presentation of the results was rendered difficult because although aluminum had the predominant influence, manganese also had an appreciable effect. To resolve this, a practice was adopted similar to that used in high tensile brass where the equivalent zinc content is calculated from Guillet's[7] coefficients of equivalence. Thus, the properties of the keel block test bars were plotted against an equivalent aluminum content calculated from the actual aluminum content plus some fraction of the manganese content (Al + Mn/x %). Since manganese had more effect on some properties than others, the manganese fraction was

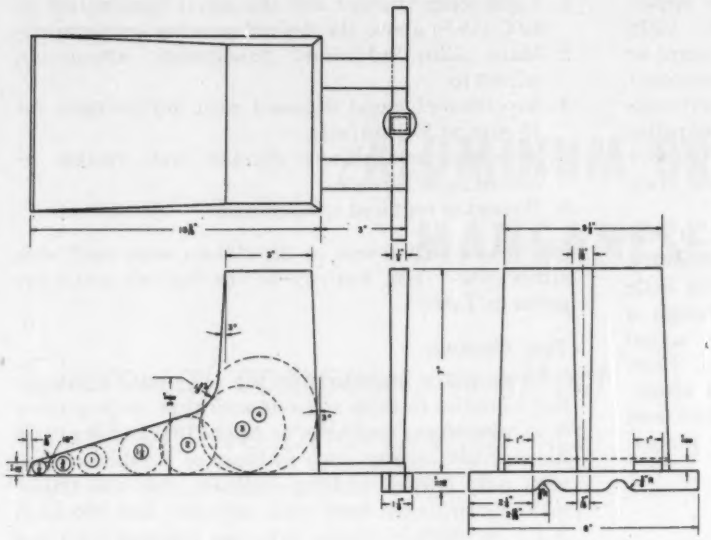


Fig. 1 — Dimensions of tapered wedge bar casting used to relate section thickness and mechanical properties for copper-base alloys. Numbered circles in side view show location and effective section thickness of test bars.

TABLE 1 — CHEMICAL ANALYSIS AND EQUIVALENT ALUMINUM CONTENTS FOR VARIOUS MELTS

| Melt No. | Analysis, % | | | | | Equivalent Aluminum Content | | |
|---|---|---|---|---|---|---|---|---|
| | Cu | Al | Mn | Fe | Ni | $\left(Al+\frac{Mn}{4}\%\right)$ | $\left(Al+\frac{Mn}{6}\%\right)$ | $\left(Al+\frac{Mn}{8}\%\right)$ |
| **Alloy A Type** | | | | | | | | |
| 506 | 80.26 | 7.12 | 6.09 | 4.63 | 1.86 | 8.64 | 8.14 | 7.88 |
| 515 | 78.01 | 7.53 | 7.67 | 4.80 | 1.86 | 9.45 | 8.81 | 8.49 |
| 560 | 74.85 | 8.37 | 9.52 | 5.20 | 1.92 | 10.75 | 9.96 | 9.56 |
| 554 | 77.18 | 8.47 | 7.26 | 5.82 | 1.13 | 10.28 | 9.68 | 9.38 |
| 511 | 77.70 | 8.48 | 7.08 | 4.98 | 1.90 | 10.25 | 9.66 | 9.37 |
| 497 | 76.70 | 8.55 | 7.46 | 5.25 | 1.98 | 10.42 | 9.79 | 9.48 |
| 559 | 76.84 | 9.25 | 7.24 | 4.64 | 1.80 | 11.06 | 10.46 | 10.16 |
| 507 | 77.78 | 10.07 | 5.95 | 4.35 | 1.82 | 11.56 | 11.06 | 10.81 |
| 516 | 75.08 | 10.19 | 8.19 | 4.52 | 1.79 | 12.24 | 11.56 | 11.21 |
| **Alloy B Type** | | | | | | | | |
| 521 | 71.96 | 7.15 | 14.36 | 4.45 | 1.92 | 10.74 | 9.54 | 8.94 |
| 518 | 75.52 | 7.30 | 10.90 | 4.38 | 1.97 | 10.02 | 9.12 | 8.66 |
| 549 | 70.53 | 7.75 | 12.20 | 3.79 | 5.70 | 10.80 | 9.72 | 9.28 |
| 510 | 73.02 | 7.82 | 12.71 | 4.58 | 1.75 | 11.00 | 9.94 | 9.41 |
| 547 | 76.26 | 8.22 | 12.12 | 2.09 | 1.07 | 11.25 | 10.24 | 9.74 |
| 548 | 72.96 | 8.27 | 13.34 | 3.08 | 3.06 | 11.36 | 10.33 | 9.81 |
| 525 | 71.03 | 8.79 | 14.04 | 4.37 | 1.89 | 12.30 | 11.13 | 10.55 |
| 524 | 72.45 | 8.83 | 12.40 | 4.43 | 1.85 | 11.93 | 10.90 | 10.38 |
| 520 | 74.01 | 9.23 | 10.51 | 4.49 | 1.77 | 11.86 | 10.98 | 10.54 |

TABLE 2 — POURING TEMPERATURE EFFECT ON MECHANICAL PROPERTIES

| Pouring Temp. | | Mechanical Properties, kpsi | | | Elong., % |
|---|---|---|---|---|---|
| C | F | UTS | YS, 0.2% | YS, 0.5% | in 2 in. |
| **Alloy A (Cu 76.70%, Mn 7.46%, Fe 5.25%, Ni 1.98%)** | | | | | |
| 1295 | 2360 | 94.8 | 36.5 | 38.7 | 30.0 |
| 1240 | 2265 | 96.0 | 37.4 | 39.7 | 31.5 |
| 1210 | 2210 | 95.7 | 36.9 | 39.1 | 32.0 |
| 1150 | 2100 | 95.3 | 36.9 | 39.0 | 31.5 |
| 1100 | 2010 | 93.1 | 37.5 | 39.3 | 25.0 |
| **Alloy B (Cu 73.02%, Al 7.82%, Mn 12.71%, Fe 4.58%, Ni 1.75%)** | | | | | |
| 1300 | 2370 | 98.0 | 41.7 | 44.4 | 36.5 |
| 1250 | 2280 | 98.0 | 46.5 | 47.1 | 35.5 |
| 1200 | 2190 | 98.2 | 44.1 | 47.6 | 36.0 |
| 1150 | 2100 | 100.0 | 44.3 | 45.2 | 35.0 |
| 1090 | 1995 | 98.5 | 41.1 | 41.9 | 27.0 |
| 1045 | 1910 | 96.2 | 42.4 | 43.5 | 25.5 |

varied depending on the property being plotted, and smoothest curves were obtained by plotting:

$$\text{UTS vs. } Al+\frac{Mn}{6}\%$$

$$\text{YS vs. } Al+\frac{Mn}{4}\%$$

$$E\% \text{ vs. } Al+\frac{Mn}{8}\%$$

$$H_B \text{ vs. } Al+\frac{Mn}{4}\%$$

The numerical values of the equivalent aluminum content for the various alloys are given in Table 1, and the mechanical properties of the keel bars plotted against these are shown in Figs. 2-5.

TABLE 3 — MECHANICAL TEST RESULTS FOR ALL KEEL BLOCK AND TAPERED WEDGE CASTINGS

| | Alloy A Type | | | | | Alloy B Type | | | | |
|---|---|---|---|---|---|---|---|---|---|---|
| Section | UTS kpsi | YS (0.2%), kpsi | YS (0.5%), kpsi | E, % in 2 in. | Bhn, 500/10/30 | UTS, kpsi | YS (0.2%), kpsi | YS (0.5%), kpsi | E, % in 2 in. | Bhn, 500/10/30 |
| **Melt No. 506** | | | | | | **Melt No. 521** | | | | |
| Keel | 65.9 | 28.3 | 29.4 | 63.0 | 104 | 89.0 | 40.9 | 42.3 | 43.0 | 130 |
| Wedge | | | | | | | | | | |
| ½-in. | 62.6 | 26.4 | 27.9 | 63.5 | | 85.5 | 44.8 | 46.4 | 38.0 | |
| ¾-in. | 63.7 | 25.8 | 27.0 | 64.0 | | 84.9 | 37.8 | 40.5 | 38.0 | |
| 1 in. | 64.1 | 26.4 | 27.8 | 65.5 | | 85.5 | 38.6 | 40.9 | 39.0 | |
| 1½-in. | 64.0 | 26.5 | 28.1 | 65.0 | | 85.8 | 37.5 | 39.0 | 34.5 | |
| 2 in. | 64.0 | 26.7 | 27.8 | 66.5 | | 86.4 | 40.5 | 42.6 | 34.5 | |
| 3 in. | 64.2 | 26.9 | 27.7 | 60.0 | | 85.0 | 39.2 | 40.5 | 31.5 | |
| 4 in. | 64.4* | 26.1* | 27.5* | 63.0* | | 84.2 | 40.6 | 42.9 | 27.5 | |
| **Melt No. 515** | | | | | | **Melt No. 518** | | | | |
| Keel | 76.5 | 32.1 | 34.5 | 52.0 | 119 | 80.1 | 86.9 | 38.7 | 47.0 | 119 |
| Wedge | | | | | | | | | | |
| ½-in. | — | — | — | — | | 78.1 | 35.1 | 36.9 | 47.0 | |
| ¾-in. | 74.4* | 35.3* | 37.2* | 55.0* | | 77.7 | 34.3 | 36.5 | 51.0 | |
| 1 in. | 74.4 | 31.9 | 33.7 | 55.0 | | 78.0 | 36.2 | 37.8 | 50.5 | |
| 1½-in. | 74.6 | 32.9 | 34.8 | 54.0 | | 78.3 | 34.3 | 36.3 | 49.5 | |
| 2 in. | 75.8 | 33.0 | 36.1 | 50.0 | | 78.6 | 35.7 | 37.7 | 45.0 | |
| 3 in. | 76.1 | 34.0 | 36.0 | 49.0 | | 79.0 | 35.5 | 37.3 | 43.5 | |
| 4 in. | 75.5 | 33.5 | 35.0 | 47.0 | | 79.0 | 33.9 | 35.6 | 44.0 | |
| **Melt No. 560** | | | | | | **Melt No. 549** | | | | |
| Keel | 96.5 | 39.2 | 40.7 | 24.5 | 126 | 89.4 | 43.9 | 44.6 | 13.0 | 130 |
| Wedge | | | | | | | | | | |
| ½-in. | 100.0 | 43.2 | 44.6 | 26.5 | | 82.6 | 43.0 | 43.5 | 10.5 | |
| ¾-in. | 95.5 | 36.4 | 38.0 | 30.5 | | 72.9* | 42.5* | 43.2* | 10.0* | |
| 1 in. | 95.8 | 37.2 | 39.2 | 29.5 | | 81.1 | 42.6 | 43.3 | 10.0 | |
| 1½-in. | 95.8 | 37.1 | 38.6 | 29.0 | | 79.7 | 40.8 | 41.4 | 9.0 | |
| 2 in. | 95.0 | 37.1 | 39.2 | 25.0 | | 82.6 | 41.0 | 42.3 | 10.5 | |
| 3 in. | 93.3 | 37.3 | 38.9 | 21.0 | | 79.9 | 41.3 | 42.4 | 8.5 | |
| 4 in. | 93.8 | 37.0 | 38.1 | 21.5 | | 79.3 | 41.6 | 42.6 | 8.5 | |
| **Melt No. 554** | | | | | | **Melt No. 510** | | | | |
| Keel | 90.0 | 34.1 | 36.1 | 36.5 | 114 | 99.1 | 44.2 | 46.4 | 35.5 | 126 |
| Wedge | | | | | | | | | | |
| ½-in. | 87.8* | 30.9* | 33.0* | 36.5* | | (Pouring Temperature Experiment) | | | | |
| ¾-in. | 87.5 | 32.5 | 34.0 | 34.5 | | | | | | |
| 1 in. | 87.4 | 31.7 | 34.3 | 33.5 | | | | | | |
| 1½-in. | 87.6 | 32.0 | 34.5 | 30.0 | | | | | | |
| 2 in. | 88.0 | 33.3 | 36.3 | 31.5 | | | | | | |
| 3 in. | 87.7 | 31.4 | 34.5 | 26.5 | | | | | | |
| 4 in. | 87.1 | 32.5 | 34.4 | 24.0 | | | | | | |
| **Melt No. 511** | | | | | | **Melt No. 547** | | | | |
| Keel | 90.2 | 36.8 | 38.9 | 34.0 | 124 | 100.2 | 43.1 | 44.3 | 24.0 | 136 |
| Wedge | | | | | | | | | | |
| ½-in. | 88.6 | 42.4 | 44.7 | 37.5 | | 96.3 | 40.6 | 41.6 | 20.5 | |
| ¾-in. | — | — | — | — | | 93.9 | 40.9 | 42.3 | 21.0 | |
| 1 in. | 88.6 | 34.9 | 36.7 | 37.5 | | 93.7 | 41.1 | 42.5 | 20.0 | |
| 1½-in. | 87.3 | 34.8 | 36.9 | 28.5 | | 96.7 | 40.7 | 42.2 | 18.5 | |
| 2 in. | 87.9 | 35.5 | 37.3 | 28.0 | | 95.5 | 40.9 | 43.1 | 16.5 | |
| 3 in. | 86.4 | 34.9 | 36.9 | 26.0 | | 95.3* | 40.6* | 43.0* | 15.0* | |
| 4 in. | 87.1 | 35.9 | 38.1 | 28.0 | | 97.8 | 40.6 | 42.5 | 22.0 | |
| **Melt No. 497** | | | | | | **Melt No. 548** | | | | |
| Keel | 95.5 | 36.9 | 39.1 | 32.0 | — | 105.4 | 45.6 | 47.1 | 20.0 | 139 |
| Wedge | | | | | | | | | | |
| ½-in. | | | | | | 102.7 | 44.3 | 45.4 | 20.5 | |
| ¾-in. | (Pouring temperature experiment) | | | | | 103.0* | 44.0* | 45.0* | 18.0* | |
| 1 in. | | | | | | 103.8 | 44.1 | 45.1 | 17.0 | |
| 1½-in. | | | | | | 106.5 | 44.8 | 45.7 | 19.0 | |
| 2 in. | | | | | | 107.8 | 45.9 | 46.4 | 16.0 | |
| 3 in. | | | | | | 105.0 | 45.3 | 45.5 | 12.5 | |
| 4 in. | | | | | | 103.9 | 46.9 | 46.1 | 14.0 | |
| **Melt No. 559** | | | | | | **Melt No. 525** | | | | |
| Keel | 107.4 | 44.6 | 45.4 | 19.5 | 133 | 125.1 | 63.0 | 58.8 | 15.0 | 185 |
| Wedge | | | | | | | | | | |
| ½-in. | 99.4 | 45.4 | 47.0 | 23.5 | | 121.1 | 63.4 | 61.0 | 15.5 | |
| ¾-in. | 103.7 | 41.5 | 42.5 | 20.0 | | 122.5 | 59.2 | 55.4 | 14.0 | |
| 1 in. | 103.6 | 42.9 | 43.7 | 19.0 | | 124.1 | 58.7 | 56.2 | 14.0 | |

*(continued on next page)*

TABLE 3 — MECHANICAL TEST RESULTS FOR ALL KEEL BLOCK AND TAPERED WEDGE CASTINGS
(Continued)

| | Alloy A Type | | | | | Alloy B Type | | | | |
|---|---|---|---|---|---|---|---|---|---|---|
| Section | UTS, kpsi | YS (0.2%), kpsi | YS (0.5%), kpsi | E, % in 2 in. | Bhn, 500/10/30 | UTS, kpsi | YS (0.2%), kpsi | YS (0.5%), kpsi | E, % in 2 in. | Bhn, 500/10/30 |
| 1½-in. | 103.5 | 41.2 | 42.7 | 18.0 | | 123.2 | 58.7 | 55.6 | 9.5 | |
| 2 in. | 99.7 | 42.2 | 43.4 | 13.0 | | 126.4 | 60.7 | 56.0 | 8.5 | |
| 3 in. | 98.6 | 41.5 | 43.5 | 12.0 | | 125.0 | 59.4 | 56.5 | 8.0 | |
| 4 in. | 92.3 | 40.4 | 42.3 | 9.0 | | 122.8 | 64.0 | 56.9 | 7.0 | |
| **Melt No. 507** | | | | | | **Melt No. 524** | | | | |
| Keel | 113.8 | 51.1 | 52.3 | 11.0 | 158 | 119.7 | 54.5 | 53.3 | 15.5 | 167 |
| Wedge | | | | | | | | | | |
| ½-in. | 108.5 | 59.2 | 56.4 | 12.5 | | 111.2 | 68.9 | 57.3 | 11.0 | |
| ¾-in. | 108.0 | 43.3 | 44.9 | 11.5 | | 113.2 | 56.0 | 53.5 | 13.5 | |
| 1 in. | 108.0 | 45.0 | 45.5 | 11.0 | | 113.1 | 54.3 | 53.0 | 11.5 | |
| 1½-in. | 108.5 | 44.6 | 45.8 | 11.0 | | 110.6 | 53.3 | 51.4 | 7.5 | |
| 2 in. | 107.3 | 45.1 | 45.5 | 9.5 | | 116.0 | 54.5 | 52.3 | 9.0 | |
| 3 in. | 108.8 | 47.7 | 47.5 | 10.0 | | 113.9 | 56.6 | 52.9 | 7.5 | |
| 4 in. | 107.0 | 48.7 | 48.5 | 8.5 | | 112.2 | 56.9 | 53.8 | 6.0 | |
| **Melt No. 516** | | | | | | **Melt No. 520** | | | | |
| Keel | 127.3 | 66.6 | 60.8 | 10.0 | 183 | 117.0 | 51.4 | 49.5 | 13.5 | 158 |
| Wedge | | | | | | | | | | |
| ½-in. | 127.3* | 80.8* | 63.1* | 9.5* | | 113.8 | 49.0 | 48.7 | 12.5 | |
| ¾-in. | 121.4 | 64.3 | 57.7 | 8.5 | | 113.8* | 50.6* | 49.5* | 13.0* | |
| 1 in. | 117.8 | 63.4 | 58.1 | 7.5 | | 107.5 | 50.7 | 50.4 | 8.5 | |
| 1½-in. | 109.7 | 63.8 | 57.5 | 4.5 | | 107.7 | 54.3 | 51.6 | 8.0 | |
| 2 in. | 105.0 | 64.8 | 58.2 | 4.0 | | 110.0* | 50.6* | 51.0* | 8.0* | |
| 3 in. | 101.9 | 64.6 | 59.0 | 4.0 | | 107.4 | 54.5 | 51.6 | 6.5 | |
| 4 in. | 101.9 | 65.8 | 58.4 | 3.5 | | 104.6 | 54.2 | 50.9 | 6.5 | |

* Value for one bar only.

The structure of the alloys and the constituents present were checked against the work of Cook et al[8] and Panseri and Leoni.[9] No detailed metallographic study was attempted because of the complexity of the alloy system. Representative photomicrographs are given in Figs. 6-15. A variety of etchants was used, depending on the structure which it was desired to resolve. These included alcoholic ferric chloride, aqueous ammonium persulphate and ammonia-hydrogen peroxide-caustic potash solutions.

## DISCUSSION

In the following discussion, it must be remembered that the experiments were limited to induction melts of 230 lb, and the pouring of these into keel blocks and tapered wedge bars. Thus, correlation of these results with large scale propeller manufacturing practice may not be exact. However, the object of the experiments was to develop a rationale whereby the foundryman could control his operations, and it is believed that the principles developed will apply for both large and small melts although the absolute values may be somewhat different. In addition, only mechanical properties have been examined, and there may be significant effects on such factors as erosion and corrosion resistance which are not shown by this investigation.

Considering first melting, it may be mentioned that previous experience[6] with aluminum bronze to A.S.T.M. B148 Alloy 9B indicated that high frequency induction melting gave severe gas absorption and reduced ductility. This was not apparent in the present experiments, which suggests that the high manganese alloys are less sensitive to melting conditions than those without manganese. Because of the large amount of readily oxidized elements in the alloys, it was thought it might be difficult to control the composition. However, in general it was found that the final analyses were close to the calculated amounts. Table 4 shows the difference between the calculated and final analyses of melts from the top and bottom of the two basic composition ranges, and it will be seen that the melting losses were slight.

From the results shown in Table 2, a pouring temperature of about 1200 C (2190 F) was selected for both alloys. This is somewhat higher than that recom-

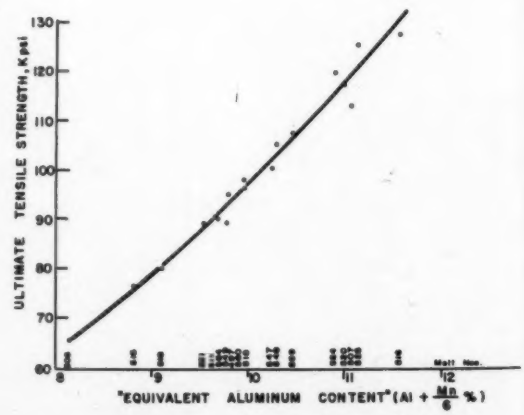



Fig. 2 — Ultimate tensile strength of keel blocks vs. equivalent aluminum content (Al + Mn/6 per cent).

TABLE 4 — CHANGE OF COMPOSITION ON REMELTING

| | Low Alloy Range, % | | | | | High Alloy Range, % | | | | |
|---|---|---|---|---|---|---|---|---|---|---|
| | Cu | Al | Mn | Fe | Ni | Cu | Al | Mn | Fe | Ni |
| **Alloy A** | | | | | | | | | | |
| Ingot Composition | 64.50 | 7.07 | 6.16 | 4.55 | 1.69 | 75.00 | 9.32 | 6.24 | 4.41 | 1.77 |
| Additions Charged | 16.20 | — | — | — | 0.20 | — | 1.00 | 2.10 | — | — |
| Total Elements Charged | 80.70 | 7.07 | 6.16 | 4.55 | 1.89 | 75.00 | 10.32 | 8.34 | 4.41 | 1.77 |
| Final Melt Analysis | 80.26 | 7.12 | 6.09 | 4.63 | 1.86 | 75.08 | 10.19 | 8.19 | 4.52 | 1.79 |
| Change on Melting | | +0.05 | −0.07 | +0.08 | −0.03 | | −0.13 | −0.15 | +0.11 | +0.02 |
| **Alloy B** | | | | | | | | | | |
| Ingot Composition | 66.80 | 7.17 | 10.66 | 4.21 | 1.61 | 70.90 | 6.89 | 11.17 | 4.19 | 1.86 |
| Additions Charged | 8.45 | — | 0.40 | 0.30 | 0.40 | — | 2.21 | 3.13 | 0.20 | — |
| Total Elements Charged | 75.25 | 7.17 | 11.06 | 4.51 | 2.01 | 70.90 | 9.10 | 14.30 | 4.39 | 1.86 |
| Final Melt Analysis | 75.52 | 7.30 | 10.90 | 4.38 | 1.97 | 71.03 | 8.79 | 14.04 | 4.37 | 1.89 |
| Change on Melting | | +0.13 | −0.16 | −0.13 | −0.04 | | −0.31 | −0.26 | −0.02 | +0.03 |

mended by Bradshaw[3] for Alloy B. However, comparison of Table 2 with Bradshaw's results shows that reasonable test bar properties were obtained over a temperature range of 1300-1010 C (2370-1850 F), indicating that, at least in thin sections such as test bars, this alloy is relatively insensitive to variations in pouring temperature. It is reputed that in casting propellers, higher pouring temperatures are required for Alloy A than for Alloy B. There is some indication in Table 2 that the optimum pouring temperature for test bars is higher for Alloy A than for Alloy B.

Table 3 and Figs. 2-5 clearly show the predominant effect of aluminum on the mechanical properties of all the alloys. The effect of manganese depends on the property being considered, as is shown by the manganese fraction used in calculating the equivalent aluminum content. Thus, for a given aluminum content, an increase in manganese has most effect on yield strength and hardness (Mn/4), somewhat less effect in increasing the ultimate tensile strength (Mn/6) and relatively little effect in decreasing elongation (Mn/8). From Figs. 2-4 it can be seen that in order to meet the specified mechanical properties, these requirements are necessary:

UTS, 90,000 psi min. — Al + Mn/6 greater than 9.7%
YS, 40,000 psi min. — Al + Mn/4 greater than 10.5%
Elong., 20% min. — Al + Mn/8 less than 10%

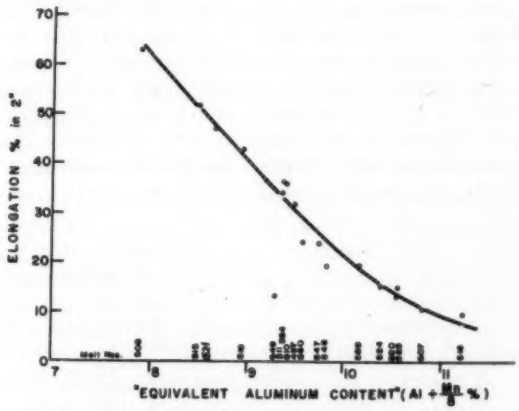



Fig. 4 — Elongation of keel blocks vs. equivalent aluminum content (Al + Mn/8 per cent).

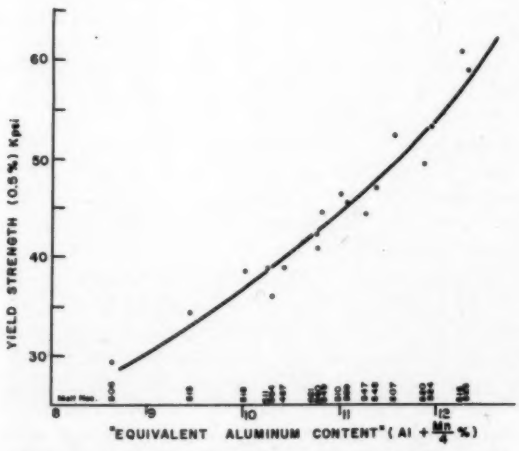



Fig. 3 — Yield strength (0.5 per cent extention under load) of keel blocks vs. equivalent aluminum content (Al + Mn/4 per cent).

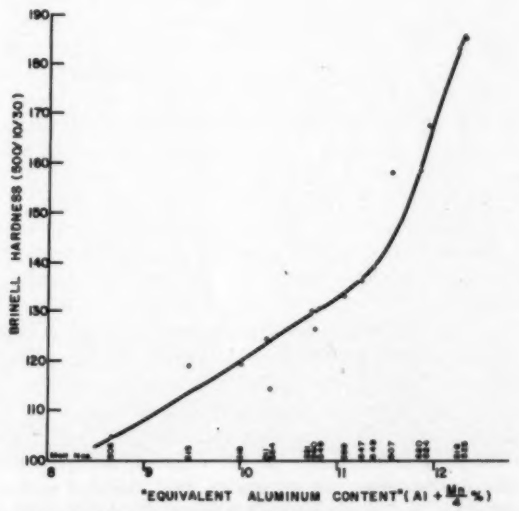



Fig. 5 — Brinell hardness of keel blocks vs. equivalent aluminum content (Al + Mn/4 per cent).

TABLE 5 — MANGANESE CONTENTS REQUIRED AT VARIOUS ALUMINUM LEVELS TO SATISFY CRITICAL (Al + Mn/x) VALUES NECESSARY TO MEET MECHANICAL PROPERTY SPECIFICATIONS

| | Critical Manganese Content, % | | |
|---|---|---|---|
| Aluminum Content, % | UTS — 90 kpsi min. $\left(Al + \frac{Mn}{6}\%\right)$ > 9.7 | YS — 40 kpsi min. $\left(Al + \frac{Mn}{4}\%\right)$ > 10.5 | E % — 20 min. $\left(Al + \frac{Mn}{8}\%\right)$ < 10 |
| 7 | 16.5 | 14 | 24 |
| 7.5 | 13.5 | 12 | 20 |
| 8 | 10.5 | 10 | 16 |
| 8.5 | 7.5 | 8 | 12 |
| 9 | 4.5 | 6 | 8 |
| 9.5 | 1.5 | 4 | 4 |
| 10 | 0.0 | 2 | 0 |

### *Critical Manganese Content*

Thus, at any given aluminum content, the manganese content just necessary for the alloy to meet the requirements can be calculated, and this is done in Table 5. From this table it can be seen that for each aluminum content there is a minimum manganese content below which the alloy will fail to meet the UTS or YS requirements, and a maximum manganese content above which it will not have the necessary elongation. These critical values in effect form the manganese control range, and are:

| Al Content, % | Critical Mn Range, % |
|---|---|
| 7 | 16.5 - 24 |
| 7.5 | 13.5 - 20 |
| 8 | 10.5 - 16 |
| 8.5 | 8 - 12 |
| 9 | 6 - 8 |
| 9.5 | 4 - 4 |
| 10 | 2 - 0 |

It is interesting to note that as the aluminum content increases, the control range for manganese decreases steadily so that at 9.5 per cent Al, 4 per cent Mn is necessary to meet the YS requirements, but more than 4 per cent Mn will cause the alloy to fail on elongation. Thus, about 9 per cent Al appears to be the practical upper limit for these alloys. At less than 8 per cent Al, the minimum manganese content is governed by the UTS requirement, while at more than 8.5 per cent Al, the minimum manganese is governed by YS. In all cases the maximum manganese which can be tolerated is dependent on the elongation value.

From all this it will be seen that within the limits of the composition ranges investigated, the alloys behaved as a continuous series from 7-10 per cent Al and 6-14 per cent Mn, and not as two alloys (Alloy A and Alloy B) with different characteristics.

The effect of iron and nickel was not investigated in detail; however, it appeared that as the nickel content was increased beyond 2 per cent, the ductility began to fall. It is not known whether the poor ductility of Melt 549 is due to the high nickel (5.70 per cent Ni) alone, or in combination with the 3.79 per cent Fe present. However, higher iron in combination with 2 per cent Ni gives normal ductilities, and it is interesting to note that one of the patents[2] specifies that the iron content shall be three times the nickel content.

All the alloys examined appeared to retain their properties quite well up to a section thickness of at least 4 in. Comparing the ¾-in. and the 4 in. sections from the wedge bars, it will be seen from Table 3 that the UTS and YS values are about the same, elongation being 5 to 10 per cent lower in the thick sections.

Although hardness values are not a specified requirement, it would appear that hardness determinations used in conjunction with the other curves could be used as a rapid control measure to predict the other mechanical properties likely to be obtained from a specific melt.

### *Mechanical Properties and Microstructure Correlation*

As noted, the metallography of these alloys was not studied in detail, but it was immediaely apparent that there was some correlation between mechanical properties and microstructure. The lower strength alloys of high ductility had a predominantly alpha structure, as shown in Fig. 6, while as the strength increased and ductility dropped, the proportion of alpha in the structure decreased, and its distribution altered, as shown in Figs. 7 and 8.

In binary Cu-Al alloys, the amount of alpha and decomposed beta is dependent on the amount of aluminum present. With the manganese containing alloys, while aluminum had the predominant effect, manganese also modified the structure so that, as with mechanical properties, the combined effect was proportional to an equivalent aluminum content. In general, the amount of alpha in the structure was inversely proportional to an equivalent aluminum content calculated from Al + Mn/4%. Some idea of the relative effects of aluminum and manganese can be obtained by comparing Figs. 6, 7, 8, and 9.

At higher magnifications, the alloys exhibited a variety of structures. In general, the basic structure was alpha crystals formed from beta at elevated temperatures, and decomposed beta. According to Panseri and Leoni, whose terminology has been used throughout as they were working with the same system, beta decomposes to a ternary eutectoid of alpha + $Cu_3Mn_2Al$ + NiAl. In addition, NiAl is said to be precipitated within the alpha, and rosettes of $FeAl_3$ are distributed throughout the structure. A typical structure is shown in Fig. 10.

The structures so far described agree closely with those of Cook[7] and Panseri.[9] However, these workers examined the structures of heat treated samples rather than as-cast ones, and it is apparent that there were some additional effects in the as-cast structures. In the first instance, while many of the fine rosettes of $FeAl_3$ in the structure were probably formed from reactions in the solid state, the size and shape of many of the larger particles suggests that they were crystallites formed from the liquid during freezing.

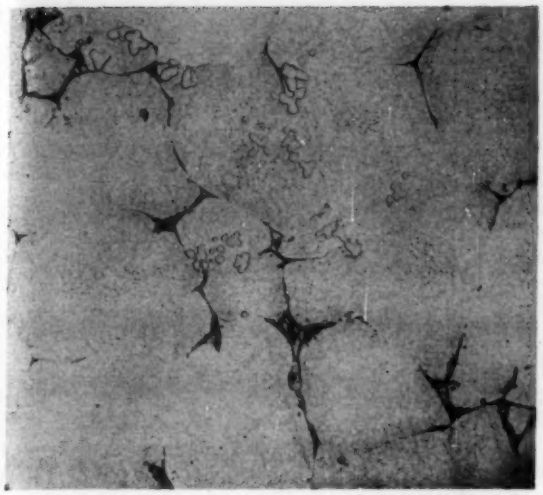


Fig. 6 — Melt 506 (Cu-80.26 per cent, Al-7.12 per cent, Mn-6.09 per cent, Fe-4.63 per cent, Ni-1.86 per cent). Mainly alpha rosettes of $FeAl_3$ and decomposed beta. 150 ×.

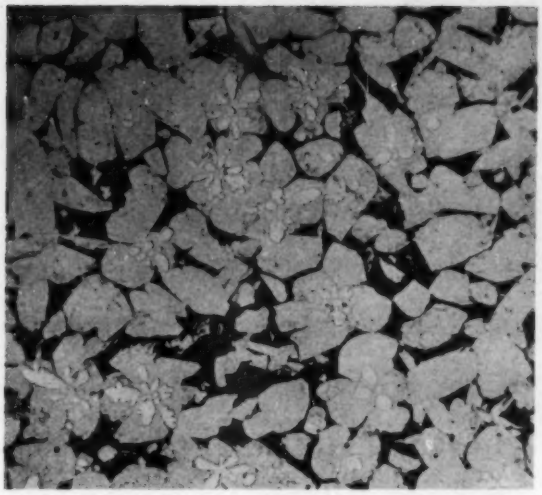


Fig. 7 — Melt 521 (Cu-71.96 per cent, Al-7.15 per cent, Mn-14.36 per cent, Fe-4.45 per cent, Ni-1.92 per cent). Note how the same aluminum content, increased manganese has modified structure, as compared to Fig. 6. 150 ×.

These larger crystallites may appear to be similar to $FeAl_3$, such as in Fig. 7, but, in many cases, it can be shown by different etching techniques that these coarser dendrites appear to be somewhat different from the small $FeAl_3$ rosettes. Figure 11 shows a structure in which at least two compounds are apparent in the dendrites, and Fig. 12 is an enlarged view of the same field. Figures 13 and 14 illustrate other typical forms of these larger crystallites, which obviously can not be the single compound $FeAl_3$ but are probably closely related to it. It was noted that in almost every case, these larger crystallites were surrounded by substantial alpha crystals, Fig. 7 being typical.

### *Manganese Aluminum Content Changes Effect*

It was found that changes in the manganese or aluminum content had little effect on the incidence of the coarse crystallites of compound, but in general they did increase the total amount of compound in the structure in the form of finer particles. This was particularly true of aluminum. In addition, as the aluminum and manganese increased, the complexity of the $FeAl_3$ type compounds also increased in that



Fig. 9 — Melt 516 (Cu-75.08 per cent, Al-10.19 per cent, Mn-8.19 per cent, Fe-4.52 per cent, Ni-1.79 per cent). Similar structure to Fig. 8, but slightly more alpha despite higher aluminum content. 150 ×.



Fig. 8 — Melt 525 (Cu-71.03 per cent, Al-8.79 per cent, Mn-14.04 per cent, Fe-4.37 per cent, Ni-1.89 per cent). Alpha needles in decomposed beta matrix. Increased aluminum has reduced amount of alpha, compared to Fig. 7. 150 ×.



Fig. 10 — Melt 547 showing massive alpha containing precipitates of NiAl. Decomposed beta consisting of eutectoid alpha plus $Cu_3Mn_2Al$ plus NiAl. Rosettes of $FeAl_3$ throughout the structure. 1000 ×.

duplexing occurred, and the compounds adopted definite crystalline shapes other than the typical rosettes.

It was noted that the melts high in iron had a larger amount of coarse compound dendrites in the structure. These compounds might be expected to affect such factors as foundry fluidity and feeding, and might be responsible for the higher pouring temperatures reputed to be necessary for Type A alloy, which in general has a higher iron content than the other.

The main effect noted with the melts of high nickel content was the heavy precipitaion (NiAl) which occurred within the alpha grains. This is illustrated in Fig. 15.

## CONCLUSIONS

Results of mechanical tests on 18 induction melted aluminum bronze alloys with 7-10 per cent Al, 6-14 per cent Mn, 2-6 per cent Ni and 2-6 per cent Fe indicate:

The alloys appeared to be less sensitive to melting conditions than aluminum bronzes without manganese, and no abnormal loss of aluminum or manganese occurred.

The pouring temperature of the alloys did not appear to be critical, and all the alloys retained their properties reasonably well in sections up to 4 in. thick.

By adopting a concept similar to that of the equivalent zinc content in high tensile brass, an equivalent aluminum content could be calculated from the actual aluminum content plus some fraction of the manganese content. It was shown that this equivalent aluminum content could be used to predict the properties of any of the alloys studied with relatively high accuracy. This concept, coupled with determinations of hardness, and possibly rapid analysis for aluminum and/or manganese, should give effective foundry control for large melts.

The mechanical test results and metallography indicate that the copper-aluminum-manganese alloys form a continuous series with properties which change progressively with changes in aluminum or manganese content.

It appeared that increasing the nickel content from 2 to 6 per cent caused a steady deterioration in ductility, but it is possible that this could be counteracted by deliberately lowering the iron content as the nickel content was increased. This will be investigated further.

## ACKNOWLEDGMENTS

The authors wish to acknowledge material assistance and helpful discussion provided by various members of William Kennedy & Sons Ltd., Owen Sound, Ontario.

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2. G. T. Callis, *New Aluminum Bronzes,* British Patent No. 762,235 (Nov. 28, 1956).



Fig. 11 — Melt 560 showing crystallites of $FeAl_3$ and a dark compound. Note that the two occur jointly in many dendrites suggesting that the two compounds are isomorphous. 150 ×.

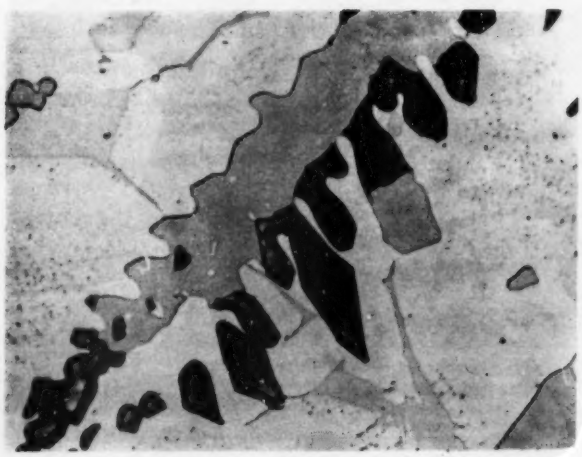


Fig. 12 — Melt 560 showing detail of dendrite arm shown in Fig. 11. Note also precipitation in alpha and smaller particles of $FeAl_3$, probably formed by reaction in the solid state. Dark phase shows coring or precipitation effects. 1500 ×.

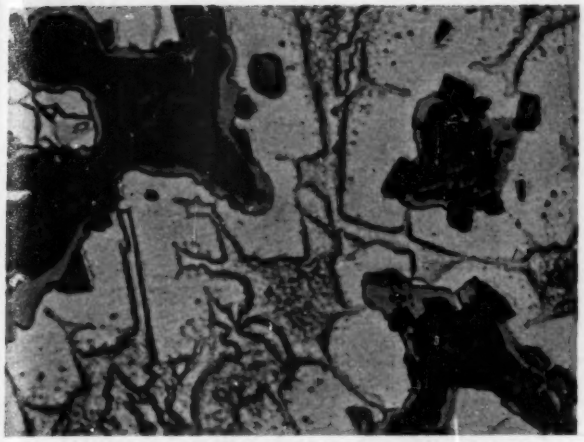


Fig. 13 — Melt 520 showing dark compound surrounded by reaction fringe of $FeAl_3$. 1500 ×.

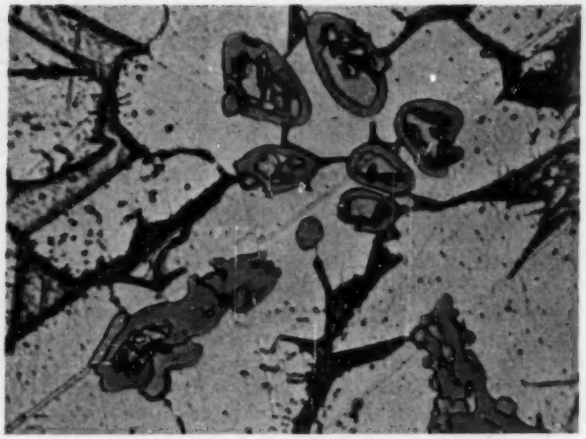


Fig. 14 — Melt 507 showing $FeAl_3$, probably containing three phases. 1500 ×.


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## APPENDIX

## SUPPLEMENTARY INFORMATION ON METALLOGRAPHY AND THE EFFECT OF IRON AND NICKEL

Previous work had shown that to obtain properties of 90 kpsi ultimate tensile strength, 40 kpsi yield strength and 20 per cent elongation, the composition of about 8 per cent Al and 12 per cent Mn appeared to be easiest to control in the foundry. The bulk of the previous work had been done on alloys with 2 per cent Ni and 4.5 per cent Fe, but three melts (547, 548 and 549) had been prepared with 2 per cent Fe, 1 per cent Ni; 3 per cent Fe, 3 per cent Ni and 3.8 per cent Fe, 5.7 per cent Ni, respectively. In these melts it was noted that there was a marked drop in elongation, at the highest iron and nickel combination. This was attributed to the high nickel content, since higher iron in combination with only 2 per cent Ni behaved normally. It was noted in the microstructures that in Melt 549 (3.8 per cent Fe, 5.7 per cent Ni) there was an almost complete absence of the typical rosettes of $FeAl_3$ and a heavy precipitation of dark etching compounds presumed to be $NiAl_3$ within the grains.

### *Melt Preparation*

Since the fact that iron and nickel had been increased simultaneously confused the interpretation, additional melts were prepared. In the first series, three iron additions were made to one melt, and test bars were poured after each iron addition. This suffered from two disadvantages

a) Test bars were poured over the range 1280-1200 C (2340-2190 F) instead of at a constant temperature of 1200 C (2190 F).
b) Even with rapid stirring, all the iron added did not go into solution, and there was additional danger that stirring might spoil melt quality by entraining oxides.

Consequently three additional melts were prepared in which two keel block test bars were poured at 1200 C (2190 F) and the pot was then returned to the furnace, make up metal and iron additions were made, the metal brought to 1200 C (2190 F), degassed with nitrogen for a further 5 min and the next two keel bars poured.

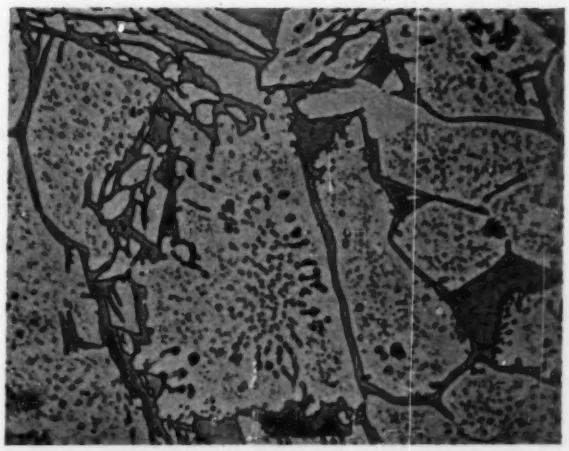


Fig. 15 — Melt 549 showing heavy precipitation of dark compound in alpha. 1500 ×.

TABLE A — CHEMICAL ANALYSIS AND "EQUIVALENT ALUMINUM CONTENTS" OF VARIOUS MELTS SHOWING THE INFLUENCE OF IRON AND NICKEL

| Melt No. | Composition, % Cu | Al | Mn | Fe | Ni | Equivalent Aluminum Content $\left(Al + \frac{Mn}{4}\right)$ | $\left(Al + \frac{Mn}{6}\right)$ | $\left(Al + \frac{Mn}{8}\right)$ |
|---|---|---|---|---|---|---|---|---|
| 547 | 76.26 | 8.22 | 12.12 | 2.09 | 1.07 | 11.25 | 10.24 | 9.74 |
| 548 | 72.96 | 8.27 | 13.34 | 3.08 | 3.06 | 11.36 | 10.33 | 9.81 |
| 549 | 70.53 | 7.75 | 12.20 | 3.79 | 5.70 | 10.80 | 9.72 | 9.28 |
| 565 A-B | | | | 2.02 | | | | |
| 565 C-D | 72.26 | 8.05 | 11.27 | 2.30 | 5.80 | 10.87 | 9.93 | 9.46 |
| 565 E-F | | | | 3.02 | | | | |
| 584 | 71.57 | 7.70 | 12.45 | 2.38 | 5.90 | | | |
| 585 | 70.44 | 7.95 | 12.54 | 3.28 | 5.79 | 11.23 | 10.22 | 9.65 |
| 586 | 69.53 | 7.99 | 12.56 | 4.13 | 5.79 | | | |
| 510 | 73.02 | 7.82 | 12.71 | 4.58 | 1.75 | 11.0 | 9.94 | 9.41 |

*Test Results*

The chemical analysis and mechanical properties of these melts and those of some of the earlier melts are given in Tables A and B, respectively, together with the equivalent aluminum contents, and the mechanical properties predicted from these.

*Discussion*

It will be seen from these tables that in all melts high in nickel, the elongation values are low, and much below those predicted from the equivalent aluminum content.

It was thought initially that this deleterious effect might be overcome by simultaneously reducing the iron content. However, it is apparent that this is not the case. Low iron in conjunction with high nickel appears to give lower elongation values, and lower ultimate tensile strength, yield strength and hardness values than when a higher iron content is present (say 4 per cent).

Metallography showed that high nickel was accompanied by the almost complete absence of $FeAl_3$ in the structure, and compositions with high nickel and low iron were almost simple two phase structures of $\alpha + \beta$, with a small amount of precipitate within the $\alpha$ grains. At a constant high nickel content, as the iron content increased the amount of dark etching precipitate within the grains increased, and dark crystallites or dendrites were observed in the structures. In no case, however, were the typical massive rosettes of $FeAl_3$ formed to any extent when nickel was above 3 per cent.

The apparent reduction in the amount and number of precipitating phases with high nickel content, and the increase in the amount of precipitate with increasing iron content could explain the observed effects on ultimate tensile strength, yield strength and Brinnel hardness number, but it is difficult at the same time to account for the elongation values which would normally react in opposition to the rest with regard to precipitation effects.

In conclusion, it was noted that the high nickel alloys tended to be whiter in color than corresponding low nickel alloys, and that the tensile fractures had a characteristic appearance. High nickel-low iron specimens broke with a rough fracture with no necking of the test bar.

With increasing iron content the fractures were almost flat, showed the crystalline structure of the bar in a manner similar to etching, and had a fine matte sheen over the surface, the whole giving the appearance of a brittle fracture.

While doing this additional work, the opportunity was taken to check some of the earlier metallographic findings by examining samples quenched after holding at different temperatures. This indicated that in this class of alloys $\beta$ is retained at room temperatures and does not undergo eutectoidal transformation. Also the massive iron rich dendrites which were thought to freeze out of the still liquid alloy were shown to form from the $\beta$ at about 900 C (1650 F), and presumably owe their characteristic shape to orientation of the $\beta$ grains or $\beta$ grain boundaries.

TABLE B — ACTUAL AND PREDICTED MECHANICAL PROPERTIES FROM MELTS SHOWING THE INFLUENCE OF IRON AND NICKEL

| Melt No. | Mechanical Properties Obtained UTS, kpsi | 0.2% YS, kpsi | 0.5% YS, kpsi | Elong., % in 2 in. | Bhn, 500/10/30 | Mechanical Properties Predicted from Equivalent Aluminum Content UTS, kpsi | 0.5% YS, kpsi | Elong., % in 2 in. | Bhn, 500/10/30 |
|---|---|---|---|---|---|---|---|---|---|
| 547 | 100.2 | 43.1 | 44.3 | 24.0 | 136 | 102 | 47 | 25 | 137 |
| 548 | 105.4 | 45.6 | 47.1 | 20.0 | 139 | 103 | 48 | 24 | 139 |
| 549 | 89.4 | 43.9 | 44.6 | 13.0 | 130 | 92 | 43 | 33 | 130 |
| 565 A-B | 78.3 | 40.2 | 41.5 | 8.0 | 136 | | | | |
| 565 C-D | 78.9 | 40.5 | 42.3 | 8.0 | 136 | 98 | 45 | 30 | 132 |
| 565 E-F | 83.8 | 40.6 | 41.9 | 12.0 | 136 | | | | |
| 584 | 76.7 | 40.1 | 41.8 | 9.0 | 112 | | | | |
| 585 | 93.0 | 46.2 | 47.0 | 11.0 | 136 | 102 | 47 | 28 | 137 |
| 586 | 96.7 | 47.2 | 48.1 | 12.0 | 143 | | | | |
| 510 | 99.1 | 44.2 | 46.4 | 35.5 | 136 | 98 | 43 | 31 | 135 |

# CASTING COST SURVEY

by J. A. Westover


## ABSTRACT

**The results of a survey on casting costing of the AFS Industrial Engineering and Cost Committee are presented. The importance of the Industrial Engineering Department in supplying the necessary means of accurate measurement is stressed as being of utmost value in the determination of casting prices.**


## INTRODUCTION

The AFS Industrial Engineering and Cost Committee has long felt the need for action on the matter of promoting understanding of the latest pricing concepts from both the engineering and the accounting standpoints. Because it was believed that most foundries would have an interest in this subject, it was decided to conduct a survey, analyze the results and report the findings to all foundrymen.

## MATERIALS SENT OUT

Materials sent out by AFS are shown in Figs. 1, 2, 3, 4 and 5:

*Fig. 1* — Covering letter signed by S. C. Massari, AFS Technical Director.

*Fig. 2* — Photographs of pattern equipment available and being used by the foundry making (a) one simple job and (b) one more complex job with cores required.

*Fig. 3* — Drawing of the simple job (throwout CAM M10183).

*Fig. 4* — Drawing of complex job (part A-557M).

*Fig. 5* — Cost Estimating Forms for various metals used.

Five different forms (aluminum, brass, iron, malleable and steel) were sent to different shops, each receiving those forms which were applicable to that particular shop's operations. Information specified on the Cost Estimate Sheets included:

1. Actual costing weights to be used.
2. Material costs (in the case of metal, ingot price of aluminum and brass, and metal market price of scrap.
3. Core weights.
4. Pattern equipment.

The specified information as to weight of the castings, metal materials costs, core weights, etc., was given in order that all survey replies would be on a consistent basis. All estimate cost figures are thus to be considered comparable.

## COST SHEET ANALYSIS

The response was good, and such that it was possible to analyze the results as anticipated. The findings of the analysis were discussed in detail in a report given at the Industrial Engineering & Cost Committee session during the Castings Congress at San Francisco on May 9, 1961.

From the analysis, it was evident that the committee has much work yet to do in this field. Help of those present at the Congress was sought, as members were urged to voice suggestions which could be answered to clarify matters. Foundrymen present were also asked to add their suggestions as to good points to be included in the recommendations.

## METHOD OF ANALYSIS

Some of the firms reporting were helpful in sending their own completed estimating forms, which made it easier to compile the desired information. This permitted a more constructive analysis of methods being used, as it was then possible to ascertain the basis of the cost estimate figures.

Some of the replies (Fig. 6) did not reflect setup or order costs. For discussion purposes at the Congress, the assertion was made that in cases where a company did not vary its costs with the order size, someone was kidding himself and his company. On the other hand, one case was found where a reporting company reflected order size with such an apparent high setup cost that they have or will undoubtedly suffer on small

**Subject: Casting Costing Survey**
**AFS Industrial Engineering & Cost Committee**

**Gentlemen:**

**We are taking this opportunity to solicit your cooperation in a survey which is being made by our Industrial Engineering & Cost Committee as a means for ascertaining the various methods utilized by different foundries for developing their actual cost for the production of a given casting. We wish to emphasize that we are only interested in the method and the final cost figure and in no way concerned with the selling price of the casting. We are attaching hereto copies of the reply form, the detailed drawing of the two castings involved, as well as photographs of the pattern and core box equipment available for the production of the casting. In addition to filling in the information requested on the questionnaire form, the committee is likewise interested in having copies of your own estimate sheets for these two jobs so that they will understand your estimating procedure.**

**All replies will be kept in strictest confidence and, should you so desire, there is no need to indicate in your reply as to the source of the information. In any event, after a thorough analysis is made of the replies, the committee will report all information in a strictly anonymous manner.**

**After you have prepared your estimates and executed the forms, please return them to the writer so that we may further make certain that all information is kept in strict confidence.**

**Thanking you for your cooperation and prompt reply, I am**

**Sincerely yours,**
**S. C. MASSARI**
**Technical Director**

**Fig. 1 — Text of letter sent out in connection with the survey.**

J. A. WESTOVER is Pres. and Treas., Westover Corp., Milwaukee.

quantity inquiries by losing orders that could be theirs. If this high setup cost is an actual cost to be absorbed, it would be well for someone to do something about seeing whether it can be reduced before it is detrimental to this company's sales efforts.

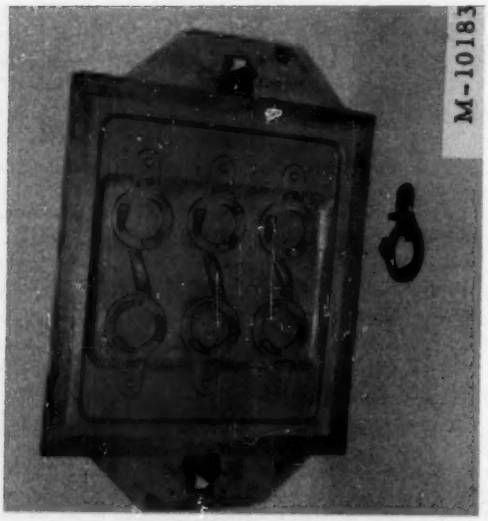
## NEED FOR INDUSTRIAL ENGINEERING

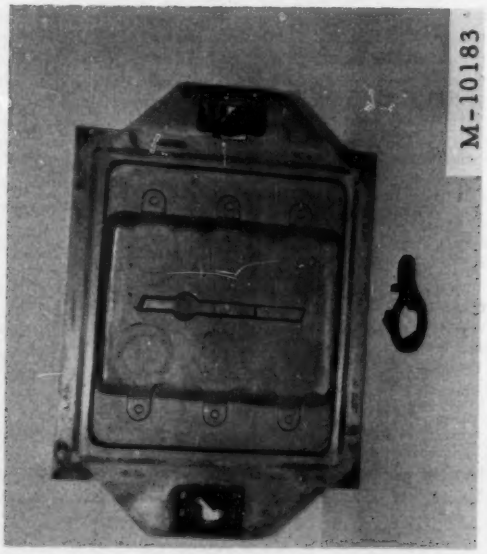
The Industrial Engineer and the Timestudy man must think of more than just a piece price or an incentive standard on foundry jobs. They should keep in mind the results of work measurement efforts for estimating and product pricing. The amount of work involved in making a particular casting in the Core, Molding and Cleaning departments can be made, no matter what the type of casting is – iron, malleable, steel or nonferrous.

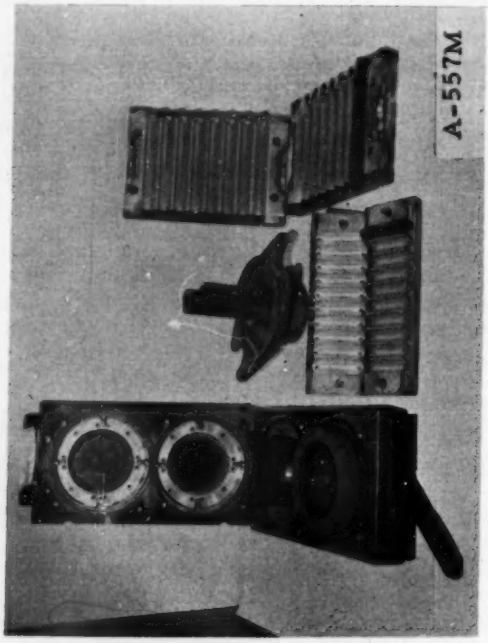
Let us consider two castings or identical weight. The first is a block casting, and little work must be done in its production. No core work is necessary. In the Cleaning Room a big snagging wheel grinder may be used. The second is an intricate casting. The Core Department must do much work on time consuming cores. In the Cleaning Room hand grinders with small mounted points must be used. More time is expended in the various departments for the production of the second casting. It can readily be understood how both the simple and intricate casting cannot have costs averaged on weight alone. Consideration must be given to the simplicity of the one, and the intricacy of the second.

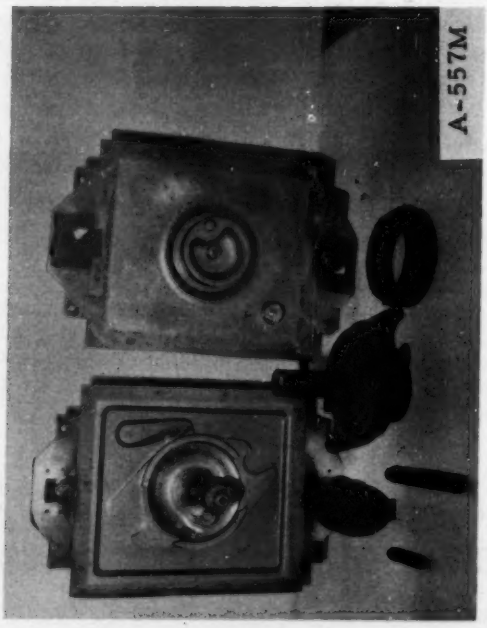
Some of the estimating forms returned reflected an average cost per pound, particularly in the Cleaning

Fig. 2 — *Top* — Matchplate pattern equipment for casting M-10183. *Bottom* — *left* — matchplate and *right* core boxes for casting A-557M.



Department. This could be the work of an accountant who cannot or will not use units of measure as provided by the Industrial Engineer.

The Industrial Engineer, unlike the accountant, can originate an accurate measure to reflect the work content of an operation without relying on historical records. In the Core Room, the Industrial Engineer can measure the actual work involved in making an individual core, and justify the expense of determining the ounces and pounds of sand needed as well as

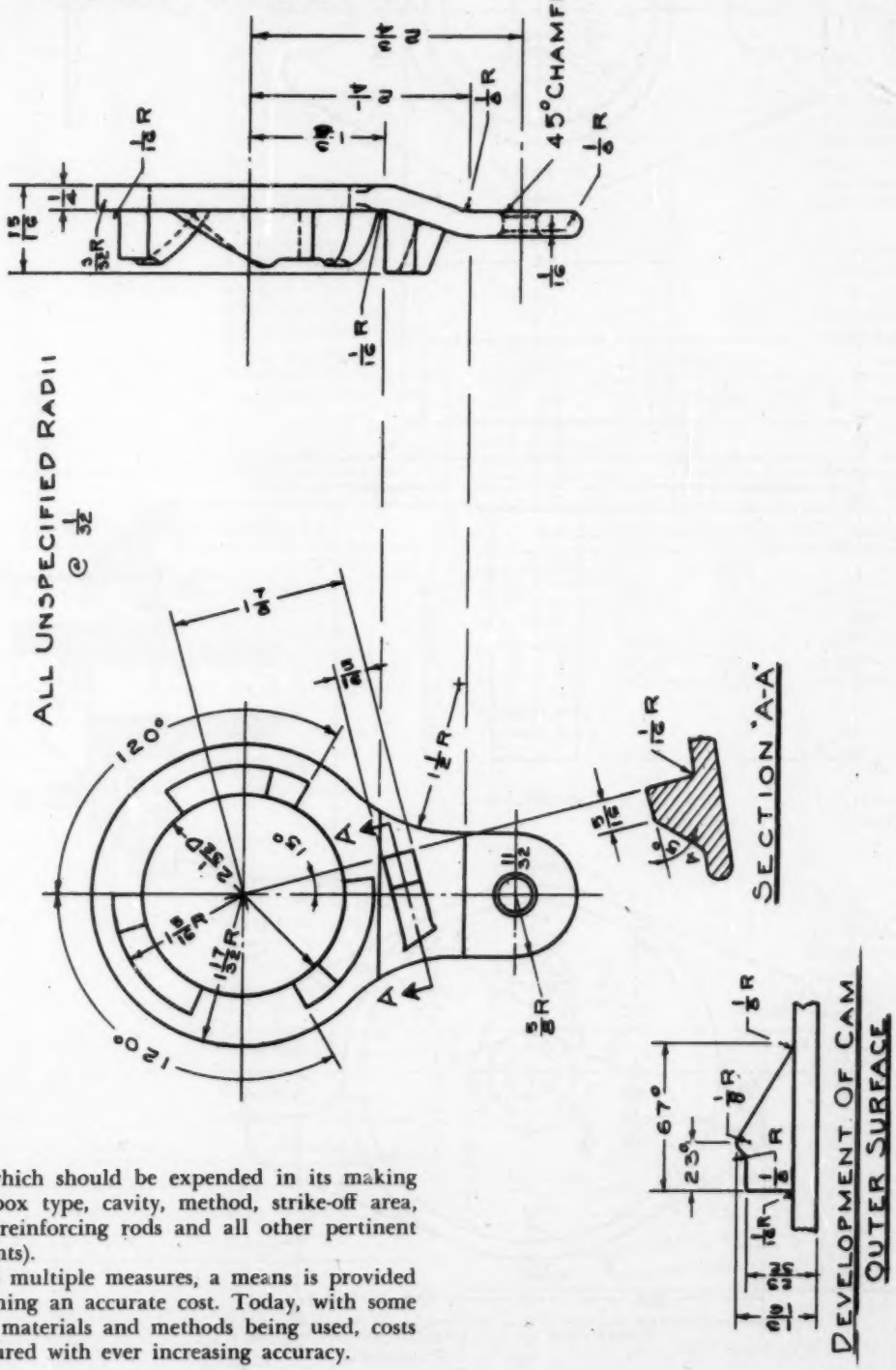



Fig. 3 — Schematic of throwout cam part M10183.

the labor which should be expended in its making (based on box type, cavity, method, strike-off area, number of reinforcing rods and all other pertinent work elements).

Using the multiple measures, a means is provided for determining an accurate cost. Today, with some of the new materials and methods being used, costs must be figured with ever increasing accuracy.

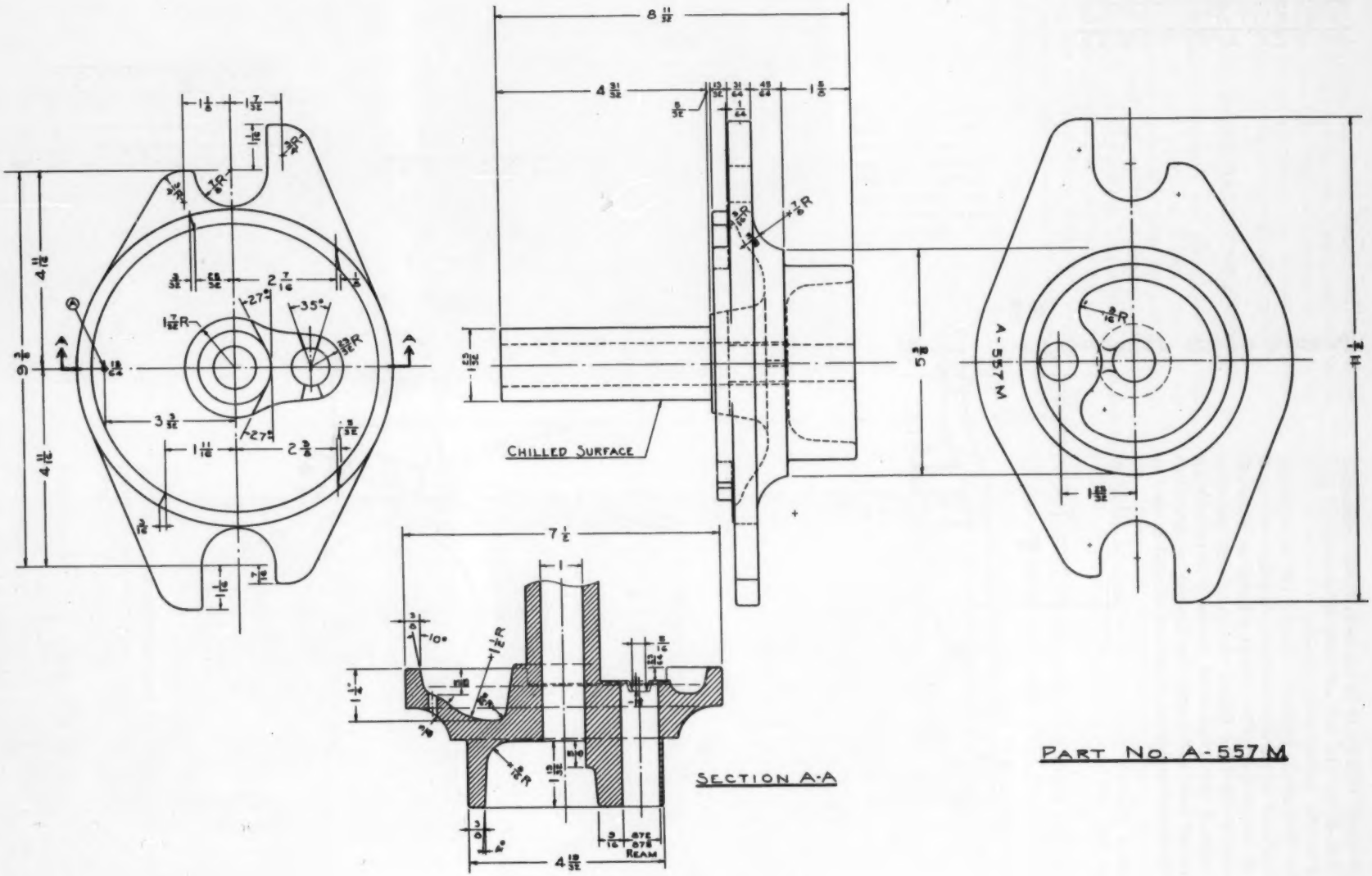



Fig. 4 — Schematic of A-557M.

### ALUMINUM FOUNDRY COST ESTIMATE

| Job Identification | M 10183 | A 557 M |
|---|---|---|
| Metal | 108 | 108 |
| Base Metal Price, lb | Ingot $0.21 | Ingot $0.21 |
| Rough Casting Weight Each, lb | 0.194 | 4.75 |
| Pattern Equipment | 6 on 14 x 16 plate Alum. Plate | 1 on 12 x 14 Alum. Plate |
| Cores and Equipment | None | Core<br>No. 1<br>2 per box 1 per cstg.<br>1.31 lb Core Mtl. per core<br>0.30 lb Arbor weight<br>No. 2<br>8 per box 1 per cstg.<br>0.34 lb Core Mtl. per core<br>No. 3<br>10 per box 1 per cstg.<br>0.09 lb Core Mtl. per core<br>No. 4<br>10 per box 1 per cstg.<br>0.06 lb Core Mtl. per core |

| Quantity | Your Cost | Your Cost |
|---|---|---|
| 500 | $____ each & $____ /lb | $____ each & $____ /lb |
| 1000 | $____ each & $____ /lb | $____ each & $____ /lb |
| 5000 | $____ each & $____ /lb | $____ each & $____ /lb |

### BRASS FOUNDRY COST ESTIMATE

| Job Identification | M 10183 | A 557 M |
|---|---|---|
| Metal | 85-5-5-5 | 85-5-5-5 |
| Base Metal Price, lb | Ingot $0.35 | Ingot $0.35 |
| Rough Casting Weight Each, lb | 0.583 | 14.25 |
| Pattern Equipment | 6 on 14 x 16 Alum. Plate | 1 on 12 x 14 Alum. Plate |
| Cores and Equipment | None | Core<br>No. 1<br>2 per box 1 per cstg.<br>1.31 lb Core Mtl. per core<br>0.30 lb Arbor weight<br>No. 2<br>8 per box 1 per cstg.<br>0.34 lb Core Mtl. per core<br>No. 3<br>10 per box 1 per cstg.<br>0.09 lb Core Mtl. per core<br>No. 4<br>10 per box 1 per cstg.<br>0.06 lb Core Mtl. per core |

| Quantity | Your Cost | Your Cost |
|---|---|---|
| 500 | $____ each & $____ /lb | $____ each & $____ /lb |
| 1000 | $____ each & $____ /lb | $____ each & $____ /lb |
| 5000 | $____ each & $____ /lb | $____ each & $____ /lb |

### IRON FOUNDRY COST ESTIMATE

| Job Identification | M 10183 | A 557 M |
|---|---|---|
| Metal | Class 30 Iron | Class 30 Iron |
| Base Metal Price, $ | 42.00 Steel Scrap<br>39.00 Cast Scrap | 42.00 Steel Scrap<br>39.00 Cast Scrap |
| Rough Casting Weight Each, lb | 0.583 | 14.25 |
| Pattern Equipment | 6 on 14 x 16 Alum. Plate | 1 on 12 x 14 Alum. Plate |
| Cores and Equipment | None | Core<br>No. 1<br>2 per box 1 per cstg.<br>1.31 lb Core Mtl. per core<br>0.30 lb Arbor weight<br>No. 2<br>8 per box 1 per cstg.<br>0.34 lb Core Mtl. per core<br>No. 3<br>10 per box 1 per cstg.<br>0.09 lb Core Mtl. per core<br>No. 4<br>10 per box 1 per cstg.<br>0.06 lb Core Mtl. per core |

| Quantity | Your Cost | Your Cost |
|---|---|---|
| 500 | $____ each & $____ /lb | $____ each & $____ /lb |
| 1000 | $____ each & $____ /lb | $____ each & $____ /lb |
| 5000 | $____ each & $____ /lb | $____ each & $____ /lb |

### MALLEABLE FOUNDRY COST ESTIMATE

| Job Identification | M 10183 | A 557 M |
|---|---|---|
| Metal | 32015 | 32015 |
| Base Metal Price, $ | 45.00 Cast Scrap<br>65.00 Pig Iron<br>40.00 Steel Scrap | 45.00 Cast Scrap<br>65.00 Pig Iron<br>40.00 Steel Scrap |
| Rough Casting Weight Each, lb | 0.583 | 14.25 |
| Pattern Equipment | 6 on 14 x 16 Alum. Plate | 1 on 12 x 14 Alum. Plate |
| Cores and Equipment | None | Core<br>No. 1<br>2 per box 1 per cstg.<br>1.31 lb Core Mtl. per core<br>0.30 lb Arbor weight<br>No. 2<br>8 per box 1 per cstg.<br>0.34 lb Core Mtl. per core<br>No. 3<br>10 per box 1 per cstg.<br>0.09 lb Core Mtl. per core<br>No. 4<br>10 per box 1 per cstg.<br>0.06 lb Core Mtl. per core |

| Quantity | Your Cost | Your Cost |
|---|---|---|
| 500 | $____ each & $____ /lb | $____ each & $____ /lb |
| 1000 | $____ each & $____ /lb | $____ each & $____ /lb |
| 5000 | $____ each & $____ /lb | $____ each & $____ /lb |

**Fig. 5 — The five different cost estimate forms.**

*(continued on next page)*

**STEEL FOUNDRY COST ESTIMATE**

| Job Identification | M 10183 | A 557 M |
|---|---|---|
| Metal | Grade B — Plain Carbon Steel | Grade B — Plain Carbon Steel |
| Base Metal Price, $ | 40.00 Steel Scrap | 40.00 Steel Scrap |
| Rough Casting Weight Each, lb | 0.583 | 14.25 |
| Pattern Equipment | 6 on 14 x 16 Alum. Plate | 1 on 12 x 14 Alum. Plate |
| Cores and Equipment | None | Core<br>No. 1<br>2 per box 1 per cstg.<br>1.31 lb Core Mtl. per core<br>0.30 lb Arbor weight<br>No. 2<br>8 per box 1 per cstg.<br>0.34 lb Core Mtl. per core<br>No. 3<br>10 per box 1 per cstg.<br>0.09 lb Core Mtl. per core<br>No. 4<br>10 per box 1 per cstg.<br>0.06 lb Core Mtl. per core |

| Quantity | Your Cost | Your Cost |
|---|---|---|
| 5 | $____each & $____/lb | $____each & $____/lb |
| 25 | $____each & $____/lb | $____each & $____/lb |
| 100 | $____each & $____/lb | $____each & $____/lb |

Fig. 5 — *continued*

It must be kept in mind that an unnecessary amount of detail in determining casting cost will induce unjustified overhead costs. Also, if there is superfluous detail as far as the people using the formula are concerned, they will lapse into "guesstimating" casting costs instead of following the established formula.

How much detail should be considered in an estimate? In the production shop, infinite detail is necessary for cost determination, because the difference of 0.1 cent can mean either receiving or losing a job. Considerable cost dollars can be at stake because of the large quantities which could be involved in such a case.

For survey purposes, it is to be noted that only the cost, and not the selling price, is of interest. This is because selling prices include markup and profit considerations which vary widely with managements and locality.

## CONCLUSIONS

In the analysis of the replies received, one important point seems to stand out above all others — those companies using standard costing methods of cost centers were quoting costs that varied greatly. It is the writer's opinion that these more truly reflect the actual costs involved than do those costs which were from companies not using cost center variances. In other words, a simple job with little work should carry a low cost, and a more complex job should carry a higher overall cost. It is recommended that cost centers be set up, and that cost center costs be used.

| | Part No. M-10183, $ | | | Part No. A-557M, $ | | |
|---|---|---|---|---|---|---|
| Order Quantity (pos.) | 500 | 1000 | 5000 | 500 | 1000 | 5000 |
| Non-Ferrous — Aluminum | 0.235 | 0.216 | 0.198 | — | — | — |
| | 0.293 | 0.262 | 0.262 | 3.12 | 3.12 | 3.12 |
| Brass | 0.435 | 0.418 | 0.399 | — | — | — |
| | 0.451 | — | — | 3.17 | 3.17 | 3.08 |
| | 0.511 | 0.482 | 0.482 | 8.55 | 8.55 | 8.55 |
| Iron | 0.113 | 0.101 | 0.092 | — | — | — |
| | 0.152 | 0.139 | 0.126 | 1.87 | 1.87 | 1.87 |
| | 0.156 | 0.143 | 0.136 | 1.81 | — | — |
| | 0.156 | 0.152 | 0.149 | 1.73 | 1.725 | 1.72 |
| | 0.192 | 0.190 | 0.188 | 2.29 | 2.20 | 2.12 |
| | 0.257 | — | — | 2.32 | — | — |
| | 0.31 | 0.31 | 0.31 | 3.73 | 3.73 | 3.73 |
| | 0.34 | — | — | 3.47 | — | — |
| Malleable | 0.16 | — | — | 2.12 | — | — |
| | — | — | 0.169 | — | 1.87 | — |
| | 0.175 | 0.171 | 0.171 | 2.46 | 2.46 | 2.46 |
| | 0.188 | 0.173 | 0.155 | 2.338 | 2.324 | 2.314 |
| | 0.200 | 0.181 | 0.173 | 2.40 | 2.39 | 2.38 |
| | 0.222 | 0.202 | 0.184 | 1.885 | 1.885 | 1.885 |
| | 0.256 | 0.230 | 0.213 | 2.70 | 2.70 | 2.70 |
| | 0.171 | 0.161 | 0.161 | 2.79 | 2.79 | 2.79 |
| Order Quantity/Steel: | 5 | 25 | 100 | 5 | 25 | 100 |
| Steel | 0.315 | 0.315 | 0.315 | 3.73 | 3.31 | 3.09 |
| | 0.341 | 0.316 | 0.275 | 5.20 | 4.82 | 4.19 |
| | — | — | 0.279 | — | — | 3.04 |
| | 0.382 | 0.364 | 0.347 | 4.16 | 3.97 | 3.78 |
| | 0.67 | 0.55 | 0.27 | 10.69 | 6.56 | 5.27 |

**Fig. 6 — Cost of castings as submitted in survey.**

It is essential that a company keep material costs separate from other costs where they become a percentage of the cost center costs. This is the only way to keep from overpricing some items and underpricing others. Such over and under pricing can easily occur due to the wide variance in material used per pound of casting or per direct labor hour.

Foundries have much constructive work to do to improve pricing methods, which will, in turn, insure obtaining a fair share of available business at a reasonable profit.

Although this survey has been helpful, the Committee is in no position to say there is one right way to cost a casting. However, the analyzed sheets have pointed out certain specific items which can and should be brought to future meetings of AFS so interested personnel can be in a better position to evaluate cost and pricing concepts that will be applicable to any work from the most simple to the most complex castings.

Without question the Industrial Engineering concept of using whatever units of measure which reflect variations in time required to perform an operation, and for the amount of material which may be used, must be the basis for determining casting cost. In those companies where it is an accounting function to supply casting costs, the Industrial Engineering Department must supply the means of accurate measurement.

It is hoped that this report has served the purpose to clarify the thinking of foundrymen on the subject of the determination of casting costs accurately and profitably.

# RAPIDLY SOLIDIFIED ALLOYS STRUCTURES

by P. E. Brown and C. M. Adams, Jr.


## ABSTRACT

Understanding the metallurgy of cast structures requires detailed consideration of the freezing process. The soundness, strength, directionality and response to heat treatment of solidified alloys are all a direct consequence of the basic mechanism of freezing. From a practical standpoint, some aspects of solidification are almost inescapable. For example, the dendritic character of the freezing process in most alloys is an inevitable feature of castings or ingots as well as arc or spray deposits. However, structures and properties can be significantly influenced by manipulation of certain control variables, and one of the most important of these is freezing rate.

By studying and comparing the structures produced in alloys solidified rapidly and slowly, and observing their similarities and differences, not only can information of practical significance be gained, but such study sheds light on the fundamental processes which accompany solidification. Since arc deposits are castings of a sort which freeze at particularly high rates under conditions which can be closely controlled, the arc has been used to study, in one extreme, the effect of freezing rate on cast structure, to check certain theoretical concepts of solidification and to further understanding of the unusual combinations of properties simetimes exhibited by arc-deposited alloys.

Freezing of a liquid solution is basically an unmixing process whereby a chemically homogeneous liquid transforms into a heterogeneous solid, a consequence of the fact that solute elements have different solubilities in liquids and solids. When a solution starts to freeze the composition of the first solid which forms is substantially different from that of the liquid, which means that solidification must be accompanied by mass transport, at least on a local scale, resulting in what is generally termed microsegregation. At freezing rates which are of particular interest, this mass transport process takes place by diffusion in and through the liquid phase.

Since solid diffusivities are much lower than liquid diffusivities, mass transport in the solid is practically negligible in almost any real solidification process. Consequently, concentration differences which are established by the freezing process are to some extent "locked in"[1,2] and can only be modified by subsequent processing at elevated temperature.

Since solute concentrations influence freezing temperatures, the chemical changes which accompany solidification have important thermal effects, and it is the interplay between chemical and thermal events which largely governs the form and dimensional character of dendritic structures. It has been a principal purpose of this investigation to define and explore the nature of dendritic solidification as it obtains in practical freezing processes involving relatively nondilute engineering alloys. Hence this study is complementary to the work of other investigators who have directed their attention to relatively low rates of solidification, and/or the freezing of nearly pure metals.


## THEORETICAL CONSIDERATIONS

To visualize the cause and effect development of the dendritic freezing pattern, consider a binary alloy of 4 per cent copper in aluminum. As the alloy starts to solidify, the first solid which forms contains only about 0.7 per cent copper; part of the copper is literally pushed aside by the growing solid, and tends to accumulate in the liquid immediately adjacent to the liquid-solid interface. Since liquid diffusion is quite slow, a concentration gradient is thereby established, with the liquid near the interface being somewhat enriched in copper.[3,4] In this circumstance the liquid near the interface has a lower freezing temperature than that which is further away. This situation is depicted schematically in Fig. 1.

In this kind of a freezing process, it is important to recognize that both heat and mass flow are involved. The rate with which a crystal can grow de-

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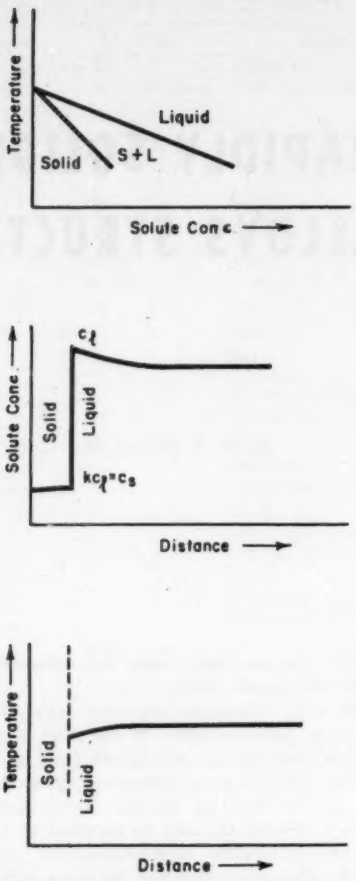



Fig. 1 — Relationships among phase diagram, solute concentration, distance from a freezing interface and liquidus temperature.

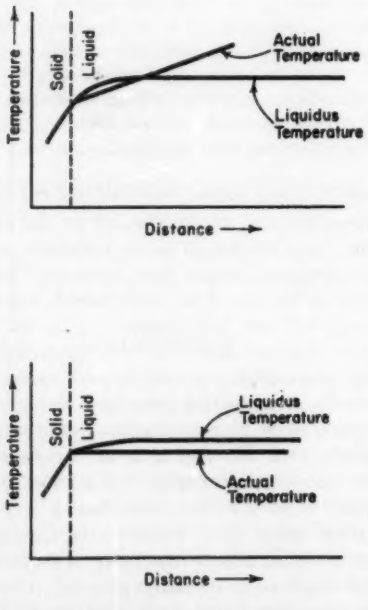



Fig. 2 — Actual and liquidus temperature gradients near a freezing interface.

pends both upon the rate of heat abstraction from the crystal and upon solute migration in the liquid. However, thermal diffusivities are much higher than mass diffusivities, so that temperature gradients are negligible when compared to concentration gradients in any ordinary freezing process (This may not always be the case, and in certain special processes involving the supply of auxiliary heat to the liquid during freezing, steep temperature gradients can be produced artificially, with profound effects on the pattern of solidification[5]).

Using the pertinent phase diagram, concentration distributions in the liquid can be expressed in terms of liquidus temperature gradients. The classical picture showing this comparison is given in the top of Fig. 2. In view of the great difference between thermal and mass diffusivities, the lower half of Fig. 2 more accurately reflects the distribution of liquidus and actual temperatures in real solidification systems. In either case, the essential boundary condition is that the actual and liquidus temperatures must conform at the liquid-solid interface. The tendency for actual temperature gradients to be flat results in a region in the liquid near the interface being slightly supercooled.

There is a tendency for the interface to break down with the growth of protrusions, which literally reach out for that liquid which is most ready to solidify. As the solidification process continues by the thickening of these protrusions, concentration gradients develop in the liquid regions between the protrusions and in some systems under proper freezing conditions there will be a further breakdown of the liquid-solid interface by the growth of secondary protrusions from the sides of the original ones. The spacing of these protrusions or dendrite arms depends, among other things, upon the amount of supercooling which the liquid can sustain during freezing.

If this amount of supercooling is small, the spacing between dendrite arms will likewise be small. The formation of dendrite arms in close proximity to one another has the effect of reducing concentration differences and therefore the magnitude of supercooling in the liquid. In fact, the spacing which nature establishes between dendrite arms is just small enough that the interdendritic liquid pools are being subjected to as much supercooling as they can sustain.

### *Dendritic Arm Spacing*

The dendrite arm spacing is a quantity which can be observed metallographically, and is an important structural feature of cast metal. Since mass transport governs this characteristic dimension and is time dependent, it would be expected that the more rapid solidification, the finer the dendrite structure. This result can be expressed quantitatively using a theoretical model to derive the governing mass transport relationships. One model, which is satisfactory for many purposes, represents the dendrite arms as parallel, closely-spaced flat plates. The concentration distribution in the vicinity of two such plates is shown in Fig. 3.

In this case, solidification takes place by the thickening of the plates with continuous rejection of solute

into the interdendritic liquid. The solute concentration in the liquid is a maximum at the liquid-solid interface, where the rejection process is taking place. When concentration differences in the liquid phase are small compared to the absolute concentrations, the concentration gradient prevailing at any point in the liquid, during solidification, is given by:

$$\frac{\partial C}{\partial X} = \frac{(1-K)\,C_l\,S}{D}\,\frac{(X)}{(l)} \qquad (1)$$

where:

$\frac{\partial X}{\partial C}$ = concentration gradient.

$X$ = distance from the center of the liquid zone (Fig. 3).

$l$ = the half thickness of the liquid zone.

$C_l$ = the concentration in the liquid at the liquid solid interface; in other words, the maximum concentration in the liquid.

$S$ = velocity of liquid solid interface.

$D$ = diffusivity of solute in the liquid.

$K$ = partition ratio.

$1-K$ = segregation coefficient.

Equation (1) shows that the concentration gradient is zero at the middle of the liquid zone, increasing to a maximum value at the liquid-solid interface. Integrating equation (1) gives an expression for the overall concentration difference in the liquid between the liquid-solid interface and the middle of the liquid zone:

$$\Delta C = \frac{(1-K)\,C_l\,S\,l}{2\,D} \qquad (2)$$

where

$\Delta C$ = the surface to center concentration difference in the liquid zone (in Fig. 3, $\Delta C = C_l - C_e$).

The slope of the liquidus line on the phase diagram can be used to convert the concentration difference into a temperature difference, which will equal the amount of supercooling prevelant in the center of the liquid zone:

$$\Delta C = \frac{\Delta T}{M} \qquad (3)$$

where:

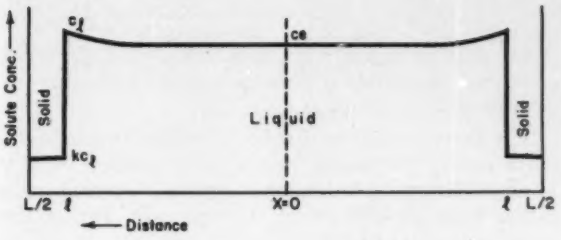
$\Delta T$ = amount of supercooling in the center of the liquid zone.

$M$ = slope of the liquidus line on the phase diagram.

The validity of equation (3) depends upon the fact that, on the local scale represented by the spacing between dendrite arms, temperature differences in metallic systems are completely negligible. This means if the liquid-solid interface is at the liquidus temperature, the center of the liquid zone, at that same temperature, is supercooled an amount given by equation (3).

### *Concentration Differences*

In most systems, the concentration differences are a maximum in the liquid near the beginning of solidification. In this case the width of the liquid zone will equal the spacing between dendrite arms:



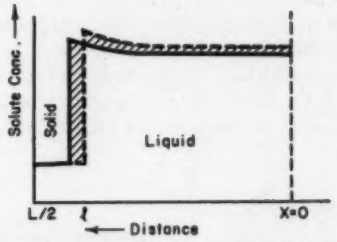




**Fig. 3 — Solute concentration vs. distance for plate-shaped dendrites thickening toward the midplane of the interdendritic liquid (a) at a given instant of time and (b) redistribution of solute as the interface advances a small distance.**

$$l = \frac{L}{2} \qquad (4)$$

where:

$L$ = dendrite arm spacing.

In general there is no knowledge of the velocity with which the surface of an individual dendrite arm propagates into the liquid. However, there is usually some means for measuring, calculating or estimating the time required for solidification, which is simply related to the freezing velocity and spacing of the dendrite arms:

$$S = \frac{L}{2\,t_o} \qquad (5)$$

where:

$t_o$ = solidification time, the lapse of time between the beginning and end of solidification.

The concept of solidification time is vital to understanding the freezing process. The more commonly used phrase, freezing rate, can be quite misleading, and is most difficult to express without ambiguity. The parameter, freezing time, has meaning even when solidification takes place under gradient conditions, if one considers the time which a tiny volume of metal spends between the liquidus and (non-equilibrium) solidus temperatures.

Substituting the above quantities into equation (2), gives an expression for dendrite arm spacing as a function of freezing time, maximum amount of supercooling which the liquid can tolerate, initial concentration of solute in the liquid, mass diffusivity and phase diagram characteristics:

$$L^2 = \frac{8\,D\,\Delta T\,t_o}{M\,(1-K)\,C_o} \qquad (6)$$

where:

$C_o$ = overall concentration of solute in alloy.

(Note that when concentration differences are small compared to absolute concentrations, near the beginning of solidification, $C_l \cong C_o$).

The dendrite model can be improved and brought more nearly into conformance with physical reality by considering the dendrite arms as parallel rods rather than parallel plates. In this case, during the early stages of solidification the rods would be of cylindrical cross-section, but ultimately become polygonal when they impinge upon one another at the end of solidification. A particular advantage of the rod model is that conditions at the beginning of solidification can be assessed with accuracy. A disadvantage is that resulting expressions do not show the effects of alloy properties and freezing conditions on dendrite spacing as simply and clearly as does equation (6).

### *Crystal Growth Factors*

In dendritic solidification under conditions of slight supercooling there are two underlying factors governing the rate and form with which the solid crystals grow. The overall rate of transportation is dictated by heat abstraction from the metal. The local growth rate of a given dendrite arm is dictated by diffusion of solute. Thus, if the overall transformation rate is increased by amplifying the rate of heat withdrawal, but the mass transport problem remains unchanged, more dendrite arms must be produced in a given space.

Consider first the growth of a single, long cylindrical rod in a supercooled alloy. It has been shown that if the alloy is initially at some uniform temperature slightly below its liquidus, and at zero time a long thin rod is precipitated so that mass flow is of cylindrical symmetry, than the radius of the rod increases parabolically with time:[6]

$$r = 2U\sqrt{Dt} \qquad (7)$$

where:

$r$ = radius of the rod.
$D$ = diffusivity of solute in the liquid.
$t$ = time after nucleation of the rod (from the equation it can be seen at zero time the radius of the rod is zero).
$U$ = dimensionless parabolic rate constant.

The parabolic rate constant can be calculated (by trial and error) from the transcendental equation:

$$2U^2 e^{U^2} \int_U^\infty \frac{e^{-y^2}}{y}\, dy = \frac{C_l - C_o}{(1-K)C_l} \qquad (8)$$

where:

$C_o$ = the initial uniform concentration of solute in the liquid, also equal to the concentration at large distances from the surface of the rod, also equal to the overall concentration in the alloy.

When the amount of supercooling is relatively small equation (8) becomes:

$$U^2 \log \frac{0.56}{U^2} = \frac{\Delta T}{2.3(1-K)MC_o} \qquad (9)$$

The parabolic rate equation can be used to relate total solidification time to interdendritic spacing:

$$L = 4U\sqrt{Dt_o} \qquad (10)$$

Whichever model is used to represent the form of the dendrites, in the evaluation of cast structures it is necessary to develop some expression for solidification times from knowledge of the thermal events which control overall transformation rate. In this investigation an arc which moves in a straight line on the flat surface of a thick plate has been used as the controlled solidification system. The thermal properties of moving heat sources as used in welding have been set forth elsewhere,[8] and the freezing time experienced by a tiny volume of metal in a continuous arc deposit can be derived from the cooling rate of the deposit:

$$t_o = \frac{Hq}{2\pi C_p K V (T_m - T_o)^2} \qquad (11)$$

where:

$H$ = heat of fusion.
$q$ = arc power.
$C_p$ = specific heat of solid metal.
$K$ = thermal conductivity of solid metal.
$V$ = arc travel speed.
$T_m$ = melting point (solidus temperature of the metal.
$T_o$ = initial uniform temperature of the plate before arc deposition.

### *Arc Deposition*

At first consideration, the technique of arc deposition would not appear to lend itself to basic solidification studies. However, when geometric and boundary conditions are kept simple in the experimental arrangement, a moving arc becomes a fairly precise kinetic calorimeter. The governing heat conduction relationships are well established and the important energetic quantities, arc power and travel velocity, can be measured with precision. There is wetted contact between the casting and its mold, which removes one of the big imponderables so often present in chill castings.

In studying the relationship among structures, properties and solidification rates in chill castings, the greatest difficulty in analyzing results comes from the important variable of mold-metal interface conductance, which cannot be measured, estimated or even reproduced with accuracy. Also, of course, such an interface condition imposes an upper limit to the solidification rate.

Another valuable feature of arc deposition as an experimental tool is that solidification occurs under steady conditions. In a chill casting the freezing rate varies continuously with distance from the chill. In an arc deposit, the freezing rate is constant along the length of the deposit, and can be calculated with accuracy, using equation (11).

Equation (11) in conjunction with either equation

(6) or equation (10) leads to the prediction that dendrite spacing should be proportional to $\sqrt{q/V}$. The ratio q/V, is usually referred to as arc energy input, and has dimensions of energy per unit length.

## RAPIDLY SOLIDIFIED STRUCTURES

Arc deposits were produced with inert gas welding apparatus using consumable and nonconsumable (tungsten) electrodes. In all cases the chemical composition of the arc deposit was identical with that of the plate material. When a permanent electrode was used, no metal was added to the plate; the travelling arc was merely used to accomplish controlled melting and resolidification in a straight line. Consumable electrode studies permitted a somewhat wider variation of arc energy input, but were limited to deposition of 2014 aluminum alloy wire on 2014 plate.

The permanent electrode process was used to study those alloys which were not available in wire form. A typical arc deposit is shown in Fig. 4. Transverse and longitudinal macrosections are shown in Fig. 5. Each of the columnar grains in Fig. 5 is, in a sense, a single crystal, but the crystal is far from homogeneous, and contains a dispersion of dendrite arms and interdendritic secondary phases, such as Cu $Al_2$. It is this substructure within single grains which has been under scrutiny in this investigation. The terminology assumes some importance, since much is written and said about the effects of grain size on the properties of cast metal.

There are, in rapidly solidified alloys, at least two important characteristic dimensions, one of which is grain size and the other dendrite arm spacing. The second seems to have more influence on mechanical properties than the first, and is much more systematically dependent upon freezing rate. With increasing freezing rate, dendrite spacing becomes smaller but grain size may not change. Grain size seems to depend more upon melt history, such as the presence or absence of grain refiners, the amount of initial superheat, etc. Since in arc deposition, the amount of superheat is likely to be quite high, it is not surprising that arc deposits can exhibit large grain size, even though solidification is rapid.

The microstructures typical of aluminum-rich alloys are shown in Figs. 6, 7 and 8 for 2014 alloy. In general, the sections shown embrace one or at most two grains, but, as can be seen, many dendrites. Under these conditions of rapid solidification, the dendrite arms in aluminum-rich alloys tend to be quite regularly spaced and parallel, and accurate measurements of the spacing between dendrites can be made. The microsections show, qualitatively, the effects of freezing rate on dendrite spacing. When dendrite spacing is plotted as a function of $\sqrt{q/V}$, a correlation is obtained (Fig. 9) which is essentially linear, in conformance with equations (6) or (10).

### *Supercooling Estimation*

In Fig. 9, the dimensions of dendrite spacing are in microns, and of arc energy input in Btu/ft. The $(\text{slope})^2$ of the straight line in Fig. 9 has the dimensions of volume per unit energy, and is equal to $4.8 \cdot 10^{-12}$ $ft^3/Btu$. This quantity, in conjunction with

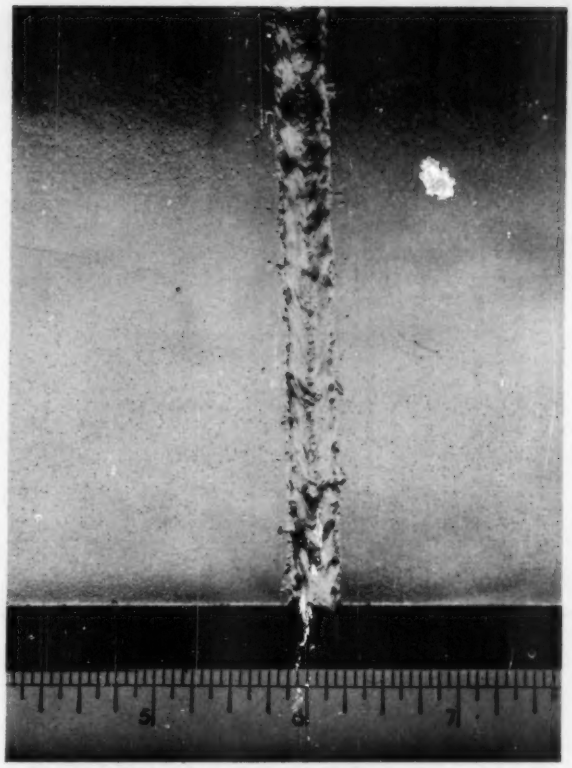


**Fig. 4 — Plane view of typical arc-deposit (106 Btu/ft).**

thermal and chemical values can be used to estimate the amount of supercooling which prevails in interdendritic pools during solidification:

$$M(1-K) = 5.05\ F/Wt\ \%\ Cu \text{ from Hanson.}^{9}$$

$$C_o = 4.4\%\ Cu.$$

$$2\pi = 6.28.$$

$$K = 79\ \frac{Btu}{(hr)\,(ft)\,(F)}.$$

$$C_p = 0.21 \frac{Btu}{lb\,F}.$$

$$(T_m - T_o)^2 = 880^2 = 7.74 \times 10^5\ F^2.$$

$$D = 2.5 \times 10^{-5}\ cm^2/sec = 1 \times 10^{-4}\ ft^2/hr^{10,11}.$$

$$H = 168\ \frac{Btu}{lb}.$$

Equation (6), derived on the basis that adjacent dendrites are parallel flat plates, gives 0.04 F for the magnitude of supercooling. Using equations (9) and (10), based on rod-shaped dendrites, the value obtained is roughly 3 times greater, or about 0.12 F. The important thing is not the precise value of this interdendritic undercooling but its small magnitude, and even more important, its apparent independence of deposition conditions; otherwise the data in Fig. 9 could not be correlated by a straight line. Such a correlation depends upon the presumption that $\Delta T$ is a constant.

When different alloy elements are dissolved in the same base material, equation (6) can be used to

suggest a generalized solidification parameter for preparing a common correlation of dendrite spacings on a single set of axes, such as that shown in Fig. 10 for three aluminum alloys exhibiting close conformance with a single straight line. The important tentative conclusion from this is that the magnitude of interdentritic supercooling is the same for all aluminum-rich alloys, even though different alloy elements may be present.

Electron micrographs of structures, such as those shown in Figs. 6, 7 and 8, were taken to detect any important substructure (i.e., dendrites within dendrites) which could not be observed at 300 magnifications. One such electron micrograph is shown in Fig. 11, the area of which covers slightly more than one dendrite arm. There was no indication of any structural feature which had not been evident at relatively low magnification.

### *Dendrite Fineness or Spacing*

Since dendrite fineness or spacing depends upon freezing rate in a way which appears to be quite

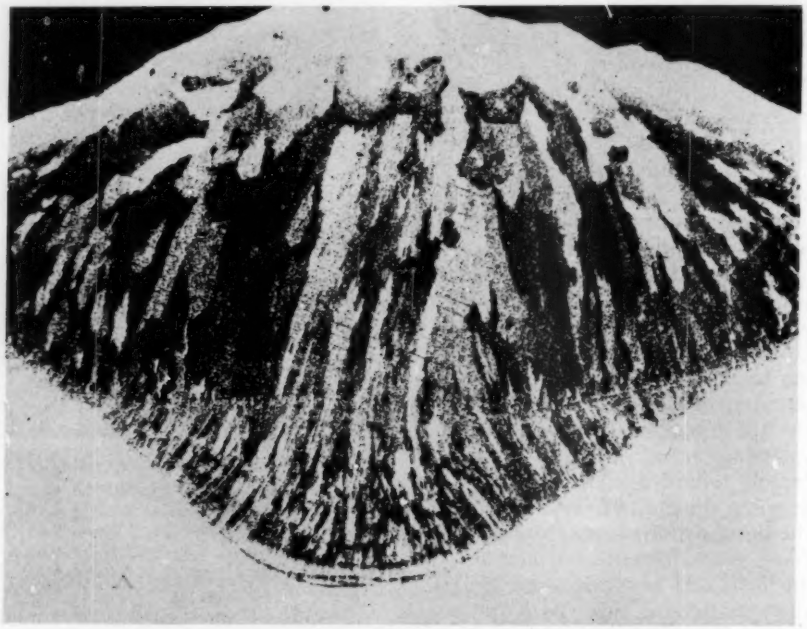




Fig. 5 — Transverse (top) and longitudinal (bottom) views of an arc deposit (106 Btu/ft). Keller's double etch. 16 ×.

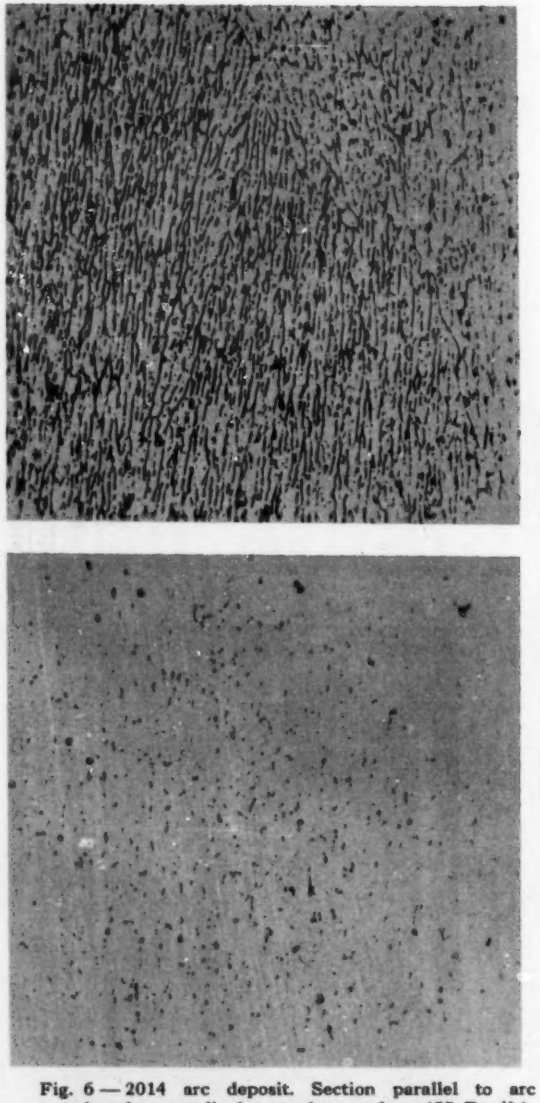


Fig. 6 — 2014 arc deposit. Section parallel to arc travel and perpendicular to plate surface (57 Btu/ft). *Top* — as-deposited, *bottom* — solutionized 940 F one hr, quenched, aged 340 F 10 hr. Keller's double etch. 300 ×.

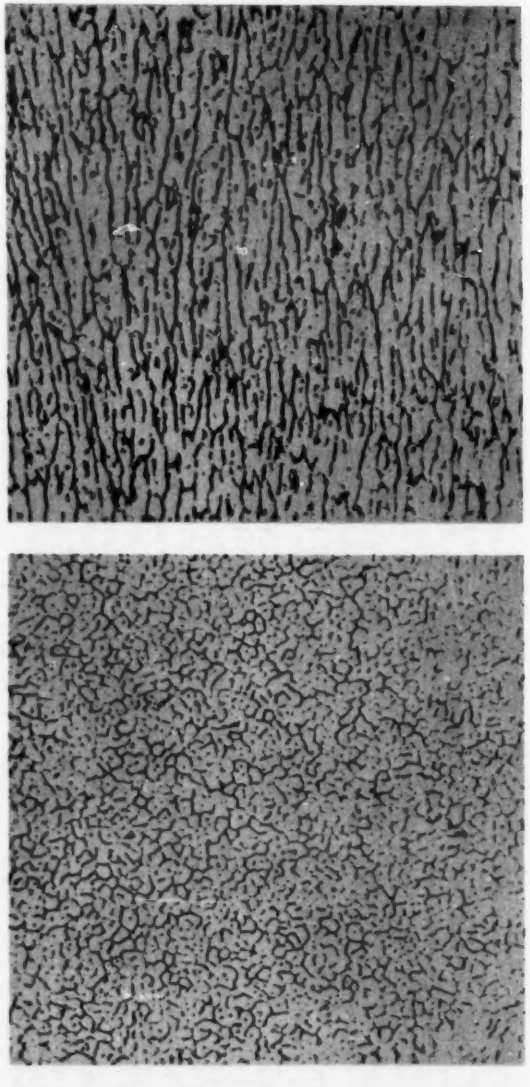


Fig. 7 — 2014 arc deposit, as-deposited (106 Btu/ft). *Top* — parallel to arc travel and perpendicular to plate surface, *bottom* — parallel to plate surface. Keller's double etch. 300 ×.

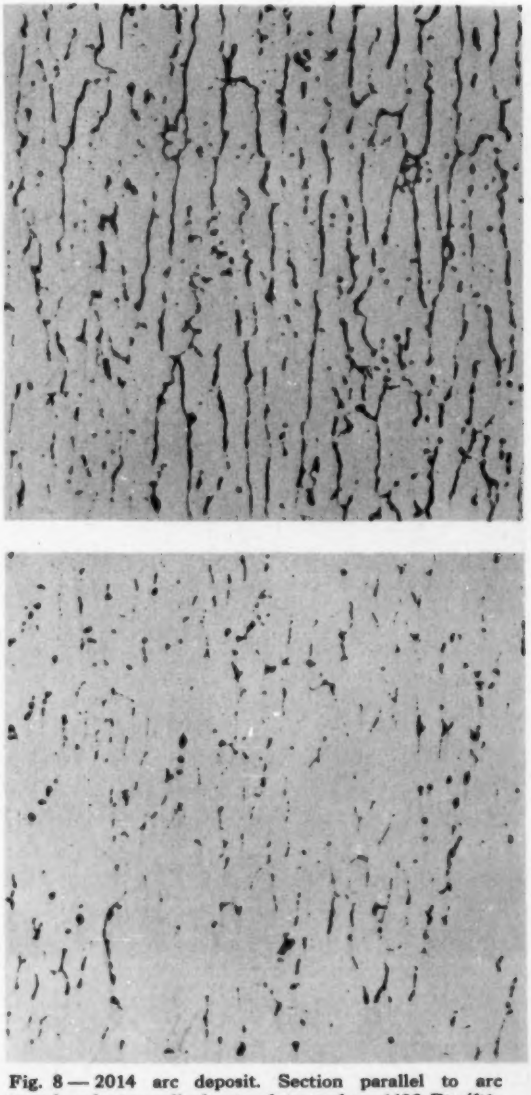


Fig. 8 — 2014 arc deposit. Section parallel to arc travel and perpendicular to plate surface (430 Btu/ft). *Top* — as-deposited, *bottom* — solutionized 940 F one hr, quenched, aged 340 F 10 hr. Keller's double etch. 300 ×.

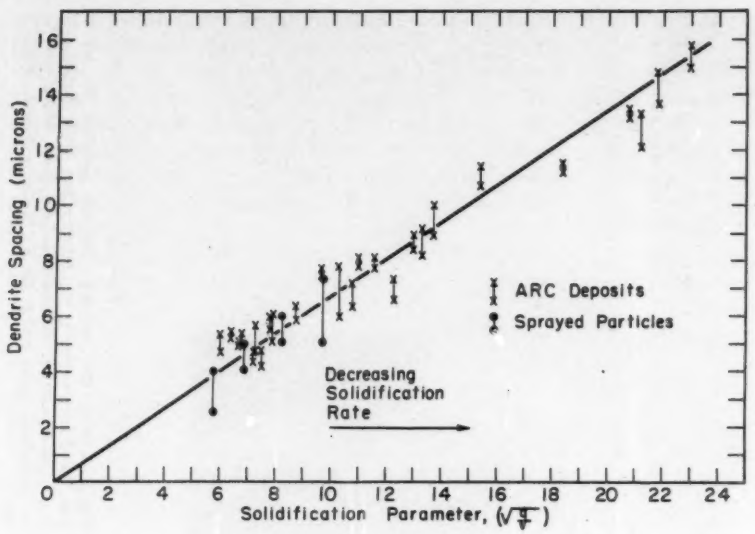



Fig. 9 — Dendrite spacing of 2014 deposits and sprayed particles vs. square root of energy input.

systematic, there is some reason to hope that freezing rates produced under different solidification conditions can be compared by interpretation of metallographic structure. This, as well as thermal considerations, immediately yields the conclusion that arc deposits freeze more rapidly than even the most severely chilled casting. For example, the only castings in which it was found possible to develop freezing rates equivalent to those observed in arc deposits were tiny spray droplets.

The freezing time of a spray droplet depends in a regular way upon its size, decreasing with decreasing droplet diameter. The structure of such a droplet 100 microns in diameter, which freezes about as rapidly as a low energy arc deposit, is shown in Fig. 12. The section in Fig. 12 contains 3 or 4 grains in different orientations, each grain containing many dendrite arms which are roughly parallel. By assuming a reasonable coefficient of heat transfer between a spray droplet and a nitrogen environment (50 Btu/ft$^2$ hr·F), a freezing time can be calculated and converted into the proper units for direct comparison with arc deposits.

Some points calculated in this way for sprayed particles are included in Fig. 9. Substantially the same dependence of structure on freezing time was found for these two widely different solidification techniques.

The work of other investigators can be used to extend the correlation to the relatively low freezing rates normally encountered in castings, and even much lower freezing rates. Work on this phase is just beginning, but some of the data presently available are correlated in Fig. 13. Points near the left of Fig. 13 correspond to severely chilled castings of aluminum alloys studied by Bever et al,[1] and points to the right reflect data obtained by Read and Flemings on castings solidified at low rates under conditions of controlled furnace cooling.[12]

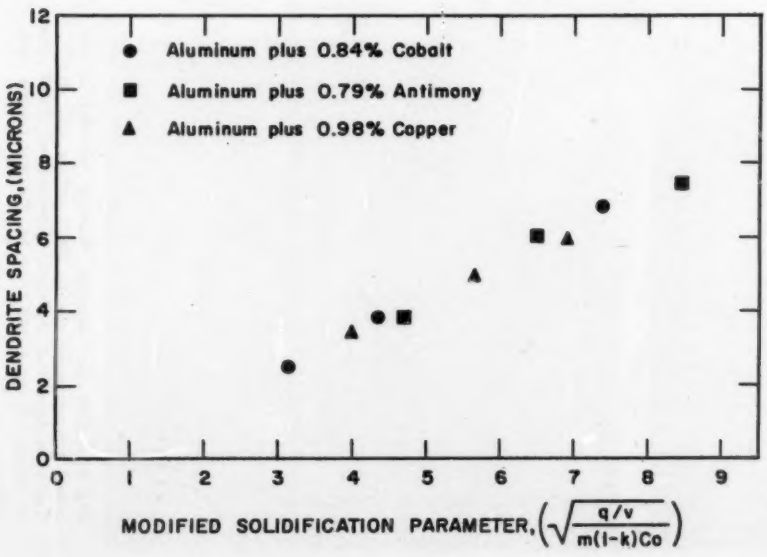



Fig. 10 — Dendrite spacing in arc deposits of three binary aluminum alloys.

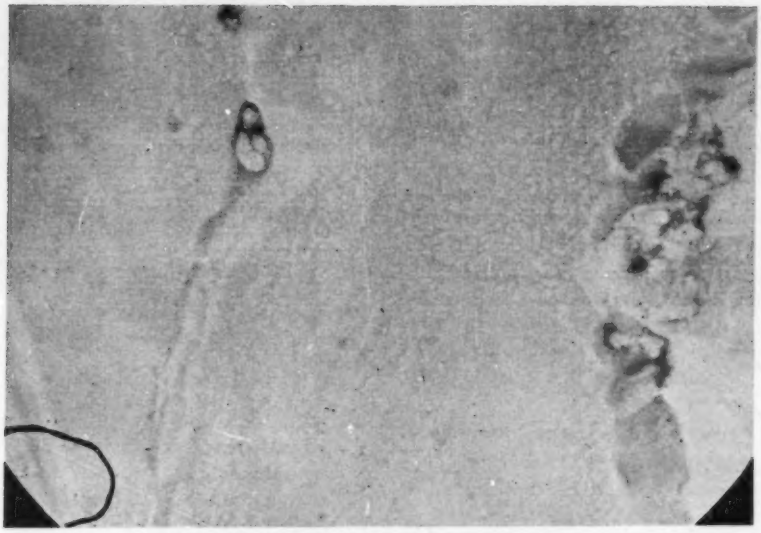


Fig. 11 — 2014 arc deposit. Section parallel to arc travel direction and perpendicular to plate surface (106 Btu/ft). Keller's double etch. 10,000 ×.

Although these data do not conform perfectly with the predicted linear correlation, the general trend is quite clear. It is entirely possible that further work on slow solidification rates may reveal mechanisms which are not accounted for in equations (1) through (11). Thus, freezing is done slowly enough, the magnitude of interdendritic undercooling might decrease below the constant value observed at higher freezing rates. Note the difference in magnitude of the vertical scale in Fig. 13 as compared with Fig. 9, both expressed in microns. All the data in Fig. 9 would fit within the extreme lower lefthand corner of Fig. 13.

### *Freezing Time Concept*

The concept of freezing time is easy enough to visualize when a small discreet volume of metal such as a spray droplet is considered, but is somewhat less clear under gradient freezing conditions. When freezing takes place progressively, as in an arc deposit or a chill casting, during solidification there is an all-liquid, an all-solid and mixed liquid-solid region. Under these conditions, in normal freezing:

1) The velocity with which the liquid-solid zone propogates through the metal will be directly proportional to the temperature gradient in the solid.
2) The width of the liquid-solid zone will be inversely proportional to the temperature gradient.
3) Combining these two facts, the time which a tiny reference volume of metal spends within the liquid-solid zone will be inversely proportional to the square of the temperature gradient.

When solidification data are interpreted in terms of the prevailing temperature gradient, a correlation is obtained as shown in Fig. 14, which is of more general interest than that shown in Fig. 9, since it is not confined to the circumstance of arc deposition.

As can be seen in Figs. 6, 7 and 8, the dendrite structure in dilute aluminum alloys solidified rapidly is quite simple, and can be well represented as an array of parallel rods. This is emphasized by the appearance of the transverse section in Fig. 7 with the plane of polish perpendicular to the axes of the dendrite arms. This was observed to be a characteristic of all dilute aluminum alloys studied. In other alloy systems, based on nickel or copper, the breakdown of the liquid solid interface caused by solute concentration gradients is more complicated, and seems to take place in two or more steps, leading to the development of primary and secondary dendrite arms.

This structure is found in rapidly solidified bronze (Fig. 15). In such a structure there are two important characteristic dimensions, the spacing between primary dendrite arms, and the spacing between secondary dendrite arms. The ratio of these spacings appears to be constant for any given alloy, and the spacings have been found to vary linearly with square root of arc energy input. Correlations for copper-nickel alloys are shown in Fig. 16.

The theoretical considerations represented by equa-

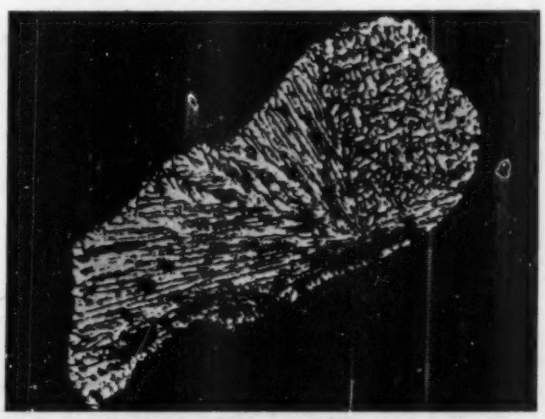


Fig. 12 — Section of 2014 droplet cooled in nitrogen. Keller's double etch. 300 ×.

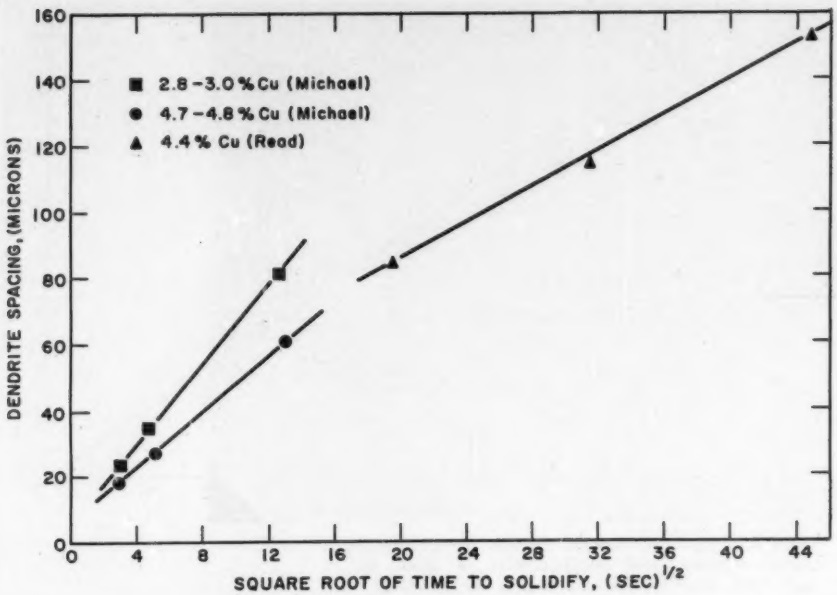



Fig. 13 — Correlation between dendrite spacing and square root of solidification time at low freezing rates. Data from Bever et al[11] and Read.[12]

tions (1) through (11) have been based on binary alloys. When several alloy elements are present, the situation is more complex. However, it is probable that one of the many elements present in a given commercial alloy will have a predominating effect in establishing the fineness of dendrite structure. The element which should be of controlling importance in this sense would be that which presents the greatest mass transport problem, i.e., which gives the lowest value of $L$ in equation (6) or (10). The vital element may or may not be the one present in greatest concentration. For example, in 2014 the 0.9 per cent silicon is probably almost as influential as the 4.4 per cent copper.

*Bronze Studies*

Studies made on a multi-component bronze have led to the correlations shown in Fig. 17. These systems exhibited primary and secondary dendrites. In spite of the fact they are neither dilute nor binary alloys, dendrite spacing was found proportional to $\sqrt{\frac{q}{V}}$, and ratios of primary to secondary dendrite arms spacings were found independent of solidification rate.

It is of interest that eutectic alloys freeze dendritically at high rates of solidification, in spite of the fact that the phase diagram suggests freezing should take place at a discreet temperature. Structures ob-

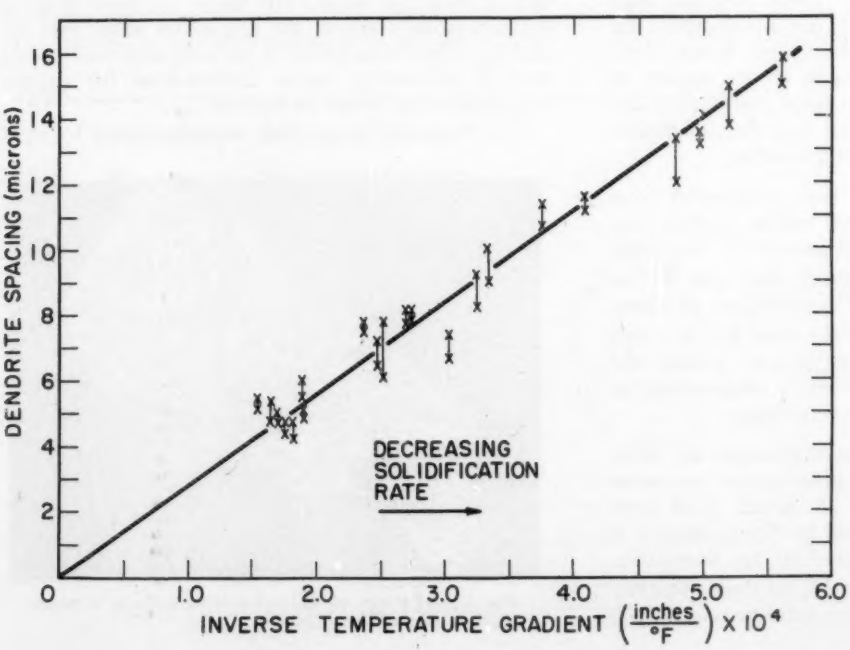



Fig. 14 — Dendrite spacing of 2014 arc deposits vs. inverse temperature gradient.

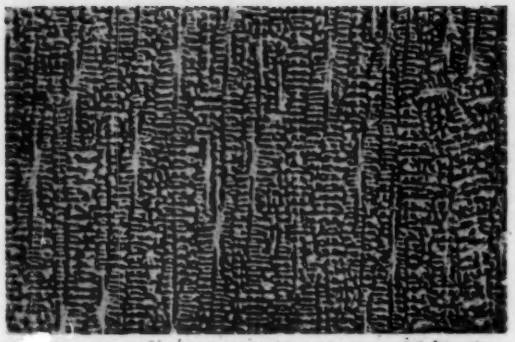


Fig. 15 — As-cast arc deposit of tin bronze (85 Btu/ft). Rosenhain's etch. 300 ×.

served in aluminum-copper alloy (33 per cent copper) are shown in Figs. 18 and 19 at two different magnifications. In addition to the rod-type of dendrite arm structure, which has to date been observed in all aluminum alloys, there is within each dendrite arm a subsidiary eutectic structure.

Although sufficient data are not yet available to warrant a firm conclusion, it appears that the spacing of the dendrite arms, as well as interlamellar spacing of the eutectic within the arms, depend in linear fashion upon the square root of arc energy input. The form of the structures observed in Figs. 18 and 19 indicates crystallization from the liquid takes place in the usual way. This is by formation of aluminum-rich columnar grains, with eutectic depositing within the crystals by an all-solid diffusion and precipitation mechanism, consistant with that suggested by Fisher and Darken for the precipitation of eutectoid pearlite from austenite in iron carbon alloys.[13]

In a preliminary study of mechanical properties, subsize test bars were machined from arc deposits. The heat flow pattern is such that the dendrite orientation in an arc deposit is nearly perpendicular to the path of arc travel, and dimensional limitations dictated the test bars to be machined with their axes parallel to the linear deposits. Thus, tensile properties could only be measured across rather than parallel to the dendrite arms.

### *Aluminum-Copper Alloy Studies*

Earlier work on chill cast aluminum-copper alloy indicated marked mechanical anisotropy of columnar structures; strength and ductility measured parallel to columnar grains were substantially higher than those determined in the transverse direction.[14] The mechanical properties of 2014 arc deposits produced at relatively high energy input are shown in Table 1. Plate and wire compositions are presented in Table 2.

The general level of properties compares favorably with those which can be realized in cast structures produced using foundry procedures. Especially noteworthy is the tensile strength after heat treatment and the ductility before and after heat treatment.

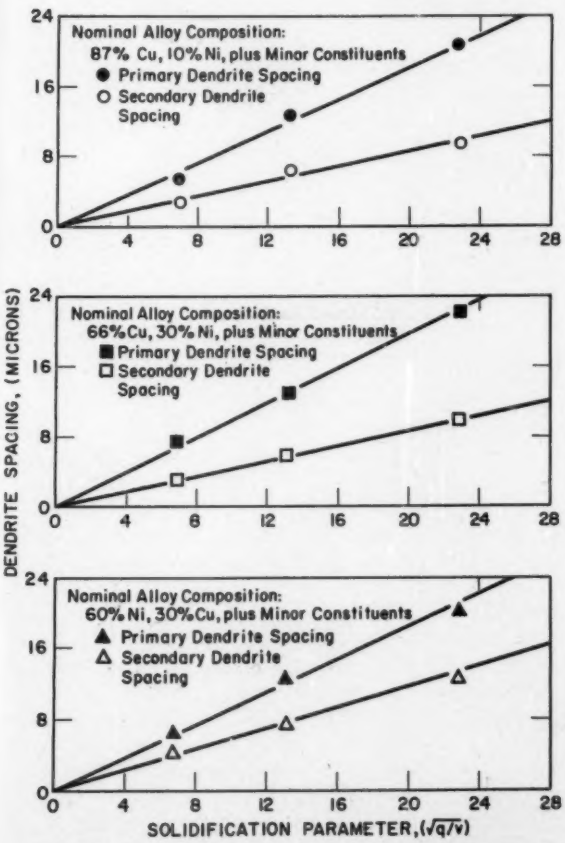



Fig. 16 — Correlation between dendrite spacing and the square root of energy input for arc deposits of three copper-nickel alloys.

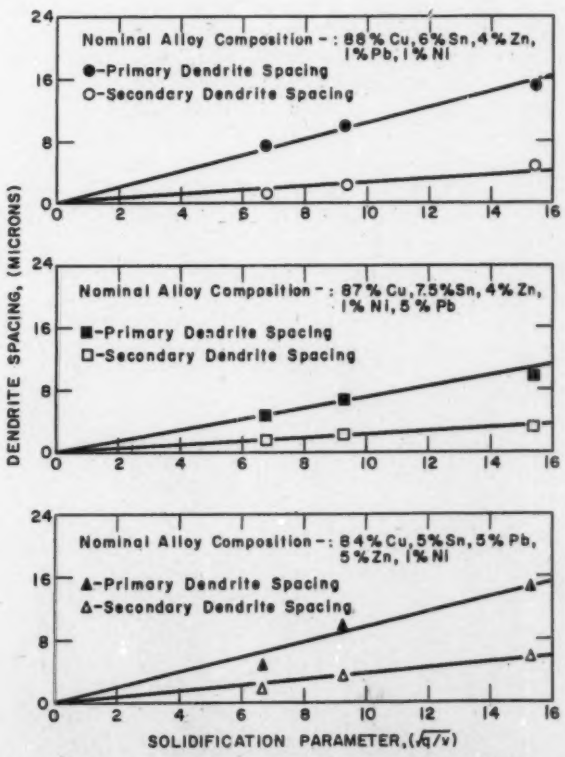



Fig. 17 — Correlation between dendrite spacing and the square root of energy input for arc deposits of three copper-tin alloys.

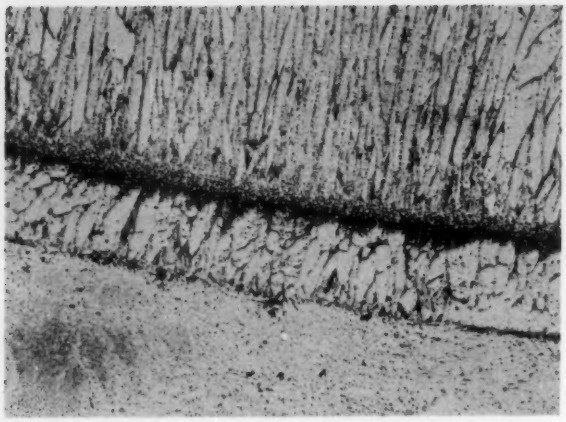


Fig. 18 — Microstructure of aluminum-copper eutectic arc deposit showing substrate and fusion zone. Section perpendicular to arc travel (170 Btu/ft). Keller's double etch. 100 ×.

TABLE 1 — MECHANICAL PROPERTIES OF 2014 ARC DEPOSITS

| | As Deposited | Solution 940 F and Quench | Solution Quench and Age* | |
|---|---|---|---|---|
| | | | 5 min | 15 min |
| Tensile Strength, psi | 45,000 | — | 63,000 | 63,000 |
| Elongation, % | 10 | — | 6.5 | 9 |
| Hardness, Rb | 50 | 42 | 77 | 85 |

*340 F, 10 hr.

TABLE 2 — COMPOSITIONS OF WIRE AND PLATE

| Composition, % | Plate | Wire |
|---|---|---|
| Cu | 4.4 | 4.4 |
| Si | 0.90 | 0.84 |
| Fe | 0.55 | 0.55 |
| Mn | 0.8 | 0.8 |
| Mg | 0.56 | 0.46 |

The effect of heat treating time on properties and structures in arc deposits is directly related to solute distribution. The separation between adjacent regions of maximum and minimum solute concentration in a cast structure is equal to half the dendrite arm spacing. It has been shown that this spacing decreases with increasing solidification rate. In the solid state process of solution heat treatment, the time required to achieve homogenization should be directly proportional to the square of dendrite arm spacing, since this is a measure of the diffusion distance involved.

For this reason, arc deposits can be fully heat treated using solutionizing times measured in minutes whereas castings require hours. It has been demonstrated that impracticably long solutionizing times (more than one week) may be required to achieve optimum properties in a sand casting, simply because interdendritic diffusion paths are so great.[15]

From this point of view it has been possible to detect differences in minimum solution time between arc deposits made at high energy input, and those with finer dendrite spacing produced at low energy input. Since precipitation heat treatment depends upon the homogeneity resulting from the solutionizing process, fully heat treated hardness is influenced by arc energy input and solutionizing time, as shown in Fig. 20. For a low energy deposit, solutionizing is essentially complete after 5 min, and longer solutionizing times have little influence on properties. High energy deposits required 15 min for fully effective solutionizing.

Another consequence of freezing rate on cast structures and properties has to do with the dispersion and influence of relatively insoluble phases. Alloy 2014, like most commercial wrought or cast aluminum alloys, has sufficient concentrations of elements such as iron, manganese and silicon, to form intermetallics much more reluctant to dissolve than $CuAl_2$. The dispersion of these relatively insoluble compounds is rather permanently established during solidification, and is substantially unalterable by solid state heat treatment. The more finely dispersed are these intermetallics, the less deleterious is their influence on ductility. These effects are shown metallographically in Figs. 6 and 8, which show structures in the solutionized condition for two different freezing rates.

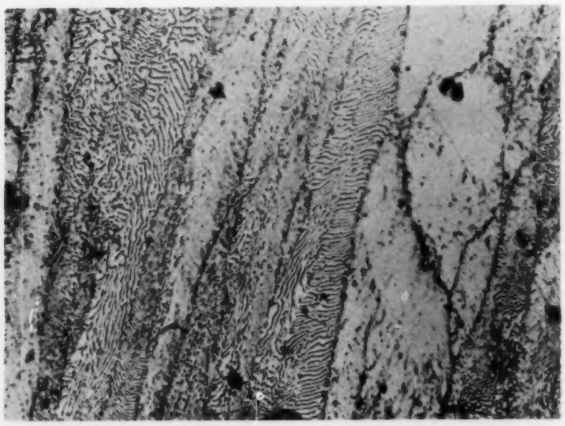


Fig. 19 — Microstructure of aluminum-copper eutectic arc deposit. Section perpendicular to arc travel (170 Btu/ft). Keller's double etch. 500 ×.

## FUTURE STUDIES

Future work will be directed to (a) extending the present study to a wide variety of alloys, (b) further exploration of the detailed way in which solidification relates to the phase diagram, (c) closer examination of the mass transport phenomena which characterize the late stages of solidification when interdendritic solute concentrations become quite high, (d) the extension of arc deposition procedures to the generation of shapes other than straight line deposits, particularly shapes which permit measurement of mechanical properties parallel to the axes of dendrite arms and (e) broadening the theoretical and experimental studies to include the duplex mass transport events which take place during solidification of polyphase structures such as eutectics.

It is also felt that careful scrutiny of results being obtained in a parallel study on structures produced at low freezing rates will ultimately lead to a far more generalized picture of the solidification process

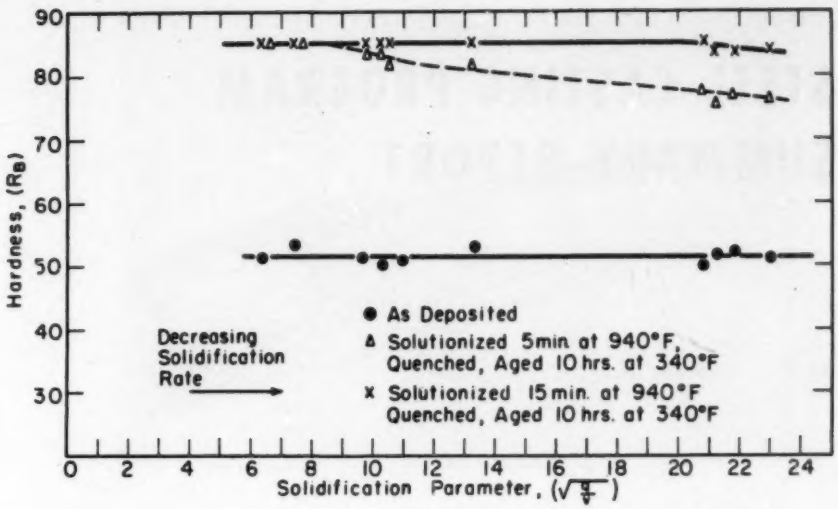



Fig. 20 — Energy input effect on response to heat treatment of 2014 arc deposits.

than is presently available. This is badly needed in order to systematize and advance the art and technique of solidification processing, as in shape casting, ingot manufacture or welding, and should provide a more fundamental basis for the development of alloys especially suited to processing by solidification.

## CONCLUSIONS

In arc deposits, dendrite spacing increases parabolically with energy input, in conformance with mass transport theory. Interdendritic undercooling in the order of 0.1 F, in dilute alloys appears to be a property of the solvent metal, does not vary significantly from one solute element to another, and is remarkably independent of solidification rate.

Arc deposits freeze much faster than chill castings and exhibit smaller dendrite arm spacings than would be expected in any casting process. The minimum time required for effective solution heat treatment of arc deposits, as reflected in final properties, depends in simple fashion on dendrite spacing (energy input).

In arc deposits, there are many dendrite arms per grain. In aluminum alloys these dendrite arms are all parallel to the principal direction of heat flow. In copper and nickel base alloys the primary dendrite arms parallel the direction of heat flow, and secondary dendrite arms are perpendicular to this direction. The ratio of primary to secondary dendrite arm spacings does not vary with solidification rate. Alloys of near eutectic composition freeze dendritically, with primary crystals apparently freezing directly from the melt and secondary phases precipitating within the primary crystals.

## ACKNOWLEDGMENT

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# STEEL CASTING PROGRAM SUMMARY REPORT

by Y. J. Elizondo


## ABSTRACT

A joint program between an airframe producer and three steel foundries was conducted under the sponsorship and direction of the Basic Industry Branch, Manufacturing Methods Div., A.M.C. The purpose of the program was to develop reliable airframe steel castings of configurations, alloys and strength levels not considered economically feasible with the existing state of the art. Alloys with the potential to meet the environmental and strength goals of the program were selected for evaluation by means of small specimens machined from keel blocks. Melt practices, process controls and inspection requirements were established to obtain optimum results. The alloys finally selected were modifications of the S.A.E. 4340 and the A.M.-355 alloys.

Three complex airframe components being produced by other fabrication processes were selected for casting by the compatible foundry process. Design modifications were made to achieve the optimum compromise between airframe design and casting design requirements. The primary objective of these modifications was to produce castings requiring little or no machining. The castings ranged in size from a 4 x 5 x 20 in. to a 6 x 49 x 59 in. envelope; in weight from 20 to 105 lb; and in thickness from 0.10 to 2.00 in. Significant changes were required in the established melt practices, composition and heat treat in order for the castings to duplicate the strengths obtained on the alloy specimens. Evaluation was made by testing specimens removed from the castings after foundry control was established. Castings were considered acceptable when they met all existing nondestructive inspection requirements.

Following this, a trial production run was made. These castings were evaluated by means of full scale static tests on randomly selected castings. Initial results indicated that several of these castings were grossly inadequate. Closer examination revealed that defects undetectable by the conventional nondestructive inspection procedures were causing significant strength reductions. The castings were improved by incorporating changes to eliminate the deficiencies and then employing an improved acceptance procedure. The acceptance criteria employed was to use a full scale static test in addition to the conventional inspection criteria. Furthermore it was required that no changes be made in any of the processes employed once the initial qualification was successful. Static tests of randomly selected castings from the trial production phase after employing this acceptance procedure indicated acceptable results and satisfactory reproducibility.

A pilot production run was then made. Randomly selected castings were subjected to full scale static and fatigue tests. These results were acceptable and indicated good reproducibility. Evaluation of dimensional tolerances and surface finish showed the castings capable of meeting the dimensional surface finish requirements for the production parts. The only exception being one or two critical areas required for mating lugs from adjacent parts. Comparison of the casting costs developed from this program and the actual costs of the production parts shows the castings to be economically feasible where production quantities of parts are required.


## INTRODUCTION

The author's company, under the sponsorship and direction of the Basic Industry Branch, Manufacturing Methods Div. of A.M.C., undertook a program to develop reliable airframe steel castings of configurations, alloys and strength levels which were not considered economically feasible with the existing state of the art. The specific objectives were:

a. To exploit the advantages of steel castings for highly loaded airframe parts operating under stringent environmental conditions.
b. To effect maximum use of advanced foundry techniques for producing high integrity castings.
c. To minimize machining and assembly costs by meeting rigid surface finish and dimensional tolerance requirements.
d. To develop necessary strength design data to permit their use in advanced weapons systems.

In order to achieve these objectives the program was divided into five phases:

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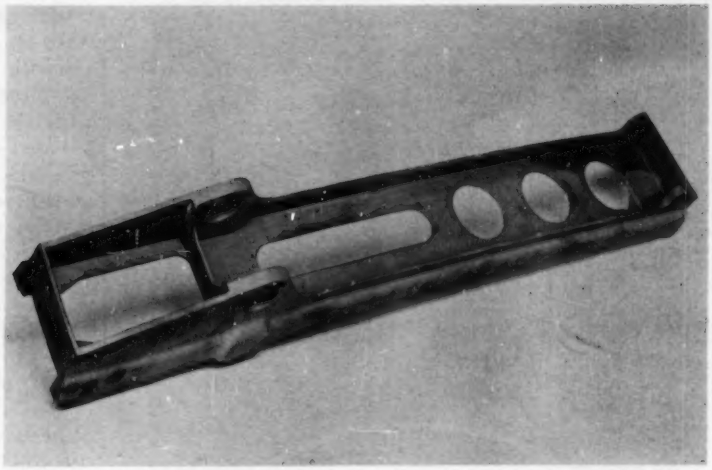


Fig. 1 — Feed back beam fitting.

*Phase I* – Selection of foundries, casting processes and configurations to be made.
*Phase II* – Alloy selection and design data testing.
*Phase III* – Casting design.
*Phase IV* – Trial production.
*Phase V* – Pilot production.

The purpose here is to summarize the problems encountered, the solutions attempted and the results obtained. In this manner it is hoped that the lessons learned as a result of this program will be of mutual benefit to both the aircraft and foundry industries, thereby, resulting in improved weapons systems for use by the services.

## PHASE I

### *Selection of Foundries and Processes*

An extensive survey was made of foundries currently supplying aircraft steel castings. The results of this survey in conjunction with discussions with cognizant personnel in the foundry industry, airframe industry and A.M.C. established a detail statement of work. Those foundries indicating an interest, which the survey revealed had the necessary capabilities and facilities, were requested to submit proposals for technical and cost evaluation. These proposals were reviewed and three foundries selected.

In making the selection consideration was given to obtaining three representative casting processes. The three foundries selected employed the three major casting processes used to produce aircraft steel castings; the sand process, the shell process and the ceramic process. The latter being an investment process employing a frozen mercury investment and a ceramic mold composition.

### *Selection of Configurations*

Once the foundries and processes had been selected, attention was directed to selection of configurations compatible with the foundry, process and program requirements. The basic requirements established general size and complexity. The three mutually agreed upon configurations were – a structural feedback beam (Fig. 1) to be produced by the ceramic process, an arresting gear drag link (Fig. 2) to be produced by the shell process and a large fin beam and bulkhead section (Fig. 3) to be produced by the sand process. These three parts represented a wide variety of foundry problems, thickness, tolerance ranges, finish requirements and sizes. Some of these items were unique to a given part or process, but generally they were common to all.

In addition these parts represented highly loaded primary structural components in an existing airframe design that were being produced by other fabrication processes. As a result direct comparisons could be made between the existing production process and the parts as produced by the casting process.

## PHASE II

### *Selection of Alloys*

An extensive literature survey was conducted to select casting alloys capable of meeting the strength goals established for the program. These goals were to develop an alloy having a tensile strength of 260,000

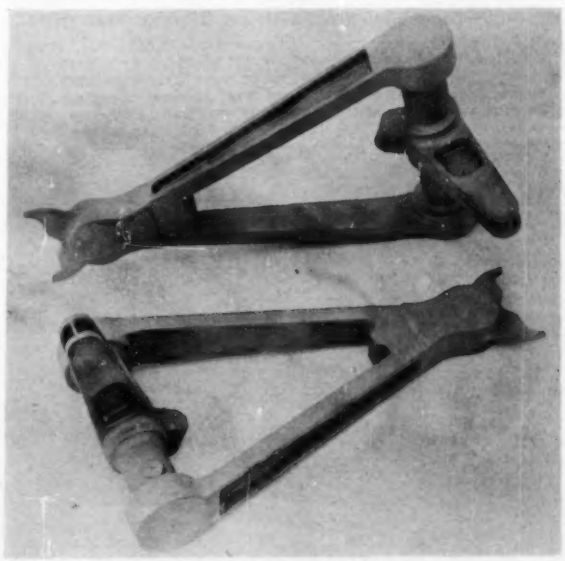


Fig. 2 — Drag link casting.

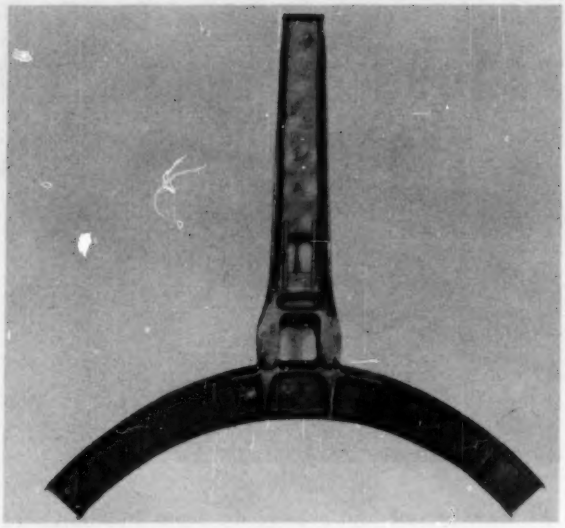


Fig. 3 — Fin beam and bulkhead fitting.

psi with a minimum elongation of 4 per cent to be used at operating temperatures up to 600 F, and one having a tensile strength of 180,000 psi with a minimum elongation of 6 per cent to be used at operating temperatures up to 1000 F.

A number of alloys were selected for evaluation. A preliminary screening was performed, and three basic alloy systems selected for further investigation. This screening was primarily based on availability of data from the basic alloy producers, foundry experience and airframe design experience. The three selections were – the H-11 alloy tool steels, the S.A.E. 4340 alloy and the A.M.-355 alloy.

Final evaluation was then made on the basis of ½-in. diameter test bars machined from cast keel blocks. Figure 4 illustrates a typical keel block and test bar. A variety of heat treatments, compositional changes and process variables were employed by the foundries to obtain optimum results. It was determined that although the H-11 alloy had the highest strength capabilities, in some cases exceeding 300,000 psi, it exhibited the most erratic results. This indicated that considerable effort on alloy and process refinement would be required to produce reliable results. The 4340 alloy exhibited strength levels exceeding 260,000 psi with significantly less scatter. The A.M.-355 alloy exhibited strength levels on the order of 200,000 psi with a reasonable degree of reliability.

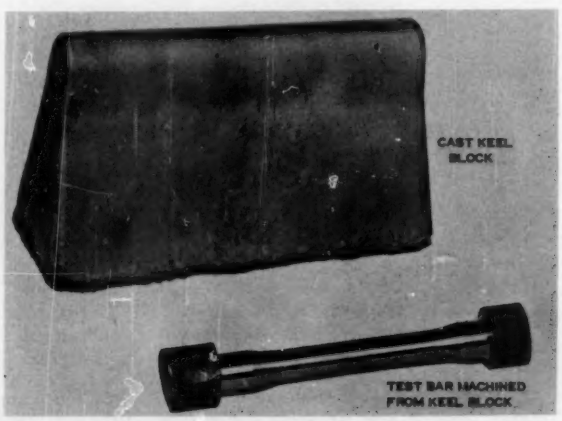



Fig. 4 — Typical keel block and test bar.

Based on the fact that the primary objectives of the program dealt with the development of the castings and the casting processes, the amount of alloy development work that could be supported was limited. For this reason, the final selection of alloys was made although the full potential of the individual alloys was not known. The two alloys selected were S.A.E. 4340 and A.M.-355. Modifications were made to these alloys to make them compatible with the environmental operating conditions required of the castings.

Once the final selection of alloys was made a series of heats were poured by the foundries under the control conditions set up on the basis of the preliminary evaluation. One set of keel blocks were poured at the assumed optimum pouring temperature – one set at 100 F higher and one set at 100 F lower. These results are summarized in Figs. 5 and 6. From these results it can be noted that pouring temperature was a significant factor for the modified 4340 alloy but not significant for the A.M.-355. In addition it can be noted that the effect varied from one process to the next. Part of this difference between processes was later traced to the practice of measuring tap temperature and recording this as pouring temperature by one of the foundries.

### *Design Data Testing*

Based on the results obtained previously, final controls were mutually agreed upon and work initiated to produce the test bars required to obtain design data for the alloys. The compositions and heat treatments agreed upon are shown in Figs. 7, 8 and 9.

One of the first problems encountered was the discrepancies in chemistry from one heat to the next, in several instances resulting in analyses exceeding the limits set up for this portion of the program. An elaborate check analysis system was set up, and individual samples were sent to recognized laboratories for determination. This only led to the conclusion that the major factors affecting the reported target chem-

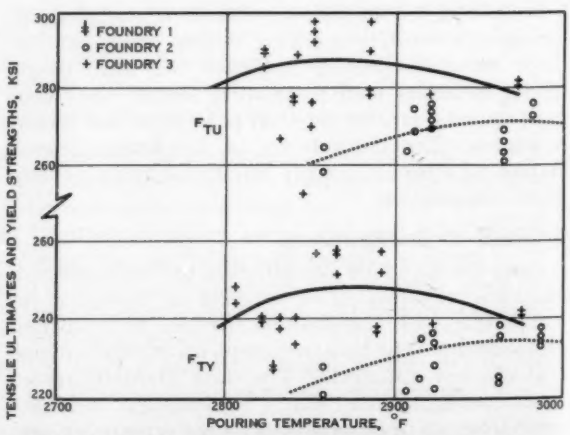



Fig. 5 — Pouring temperature effect on the properties of modified S.A.E. 4340 at room temperature.

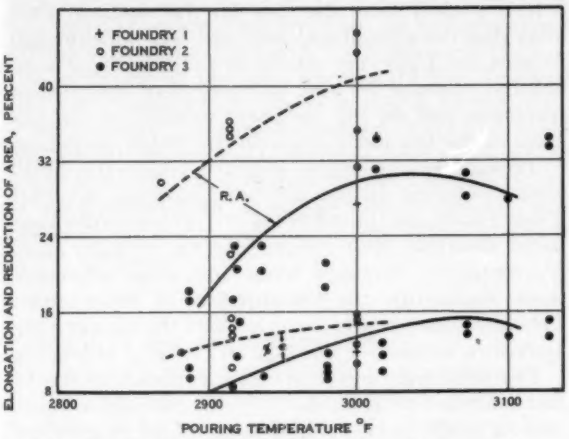



Fig. 6 — Pouring temperature effect on the properties of A.M. 355 at room temperature.

| ELEMENT | 4340 | ELEMENT | AM 355 |
|---|---|---|---|
| C | 0.40–0.45 | C | 0.10–0.14 |
| Mn | 0.50–0.75 | MN | 0.65–0.85 |
| P | 0.025 MAX. | P | 0.025 MAX. |
| S | 0.025 MAX. | S | 0.025 MAX. |
| Si | 1.50–1.80 | Si | 0.65–0.95 |
| CR | 0.9–1.1 | CR | 14.7–15.4 |
| MO | 0.65–0.80 | MO | 2.35–2.65 |
| Ni | 1.95–2.10 | Ni | 3.75–4.25 |
| V | 0.14 MAX. | N | 0.07–0.11 |
| Al | 0.05 MAX. | C+N | 0.20–0.25 |

Fig. 7 — Target chemistry.

istry were the errors in analysis by the laboratories and variations in foundry practices in pouring the chemistry slug.

A second approach was used where a given laboratory was selected to make the determinations for a given foundry. When a change in the analysis occurred a careful review was made of the melt practices to determine if any possible change should be expected. Almost without fail, when a significant change in reported chemistry occurred it could be traced to a change in melt practice, such as the time at which an addition was made, the temperature at which an addition was made or a change in producer of the additive. The exact magnitude of the change could not be predicted in all cases, but the relative change was always consistent with the predicted change based on the change in melt practice.

### *Uniform Melt Establishment*

A concerted effort was then made to establish a fixed melt practice at each foundry. The results of this effort can best be demonstrated by showing the level of sulfur and phosphorous measured over a period of time starting when standard melt practices were set up. This is shown in Fig. 10. Similar improvements were noted for all analyses variables. The differences from one laboratory to another, and from one analyses to the next, were still evident, but the differences were now significantly smaller.

Three basic specimens were used to obtain the design data – the standard ½-in. round tensile specimen, a ½-in. round notched fatigue specimen and a

| | | |
|---|---|---|
| ANNEAL: 1650°F 1200° F | 1 HOUR HOLD 3 HOURS | FURNACE COOL TO AIR COOL |
| ROUGH MACHINE | | |
| HOMOGENIZE: 2000°F | 2 HOURS | AIR COOL |
| AUSTENITIZE: 1600°F | 2 HOURS | OIL QUENCH |
| TEMPER: 625°F 625°F | 3 HOURS 3 HOURS | AIR COOL AIR COOL |

Fig. 8 — Standardized heat treatment for 4340 modified.

| AM 355 | | |
|---|---|---|
| HOMOGENIZE: 2000°F | 2 TO 4 HOURS | AIR COOL |
| SOLUTION ANNEAL: 1850°F | 2 HOURS | WATER QUENCH |
| ROUGH MACHINE | | |
| SOLUTION ANNEAL: 1850°F | 2 TO 3 HOURS | WATER QUENCH |
| SUB-ZERO: −100°F | 6 HOURS | |
| TEMPER: 850°F 950°F | 3 HOURS 3 HOURS | AIR COOL AIR COOL |
| NOTE: BASED ON EXPERIENCE 1 OF THE FOUNDRIES REPLACES THE 1850°F TREATMENT JUST PRIOR TO ROUGH MACHINING WITH −100°F, 3 HOURS AND 1125°F, 3 HOURS. THIS PRACTICE GIVES THEM BETTER MACHINING CHARACTERISTICS WITH NO APPARENT EFFECT ON THE MECHANICAL PROPERTIES. | | |

Fig. 9 — Standardized heat treatment for A.M. 355.

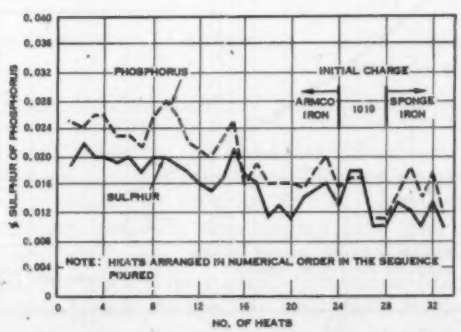



Fig. 10 — Level of impurities in modified 4340 heats.

Table 1 — Tensile Test Results

| ALLOY | TEST TEMP °F | $F_{TU}$, KSI | | $F_{TY}$ | | % e | | % RA | |
|---|---|---|---|---|---|---|---|---|---|
| | | AVG. | MIN(1) | AVG. | MIN(1) | AVG. | MIN(1) | AVG. | MIN(1) |
| 4340 MODIFIED | 80 | 275.0 | 258.7 | 237.1 | 220.1 | 1.8 | 0.5 | 3.7 | 0.5 |
| | 600 | 247.3 | 233.8 | 183.7 | 179.3 | 4.9 | 3.3 | 6.3 | 1.8 |
| AM 355 | 80 | 218.8 | 205.8 | 155.5 | 136.3 | 12.3 | 8.3 | 25.7 | 11.7 |
| | 1000 | 152.8 | 135.9 | 108.1 | 90.0 | 11.8 | 9.7 | 35.6 | 20.4 |

(1) MINIMUM BASED ON 90 PERCENT PROBABILITY WITH 90 PERCENT CONFIDENCE LEVEL.

Table 2 — Bend Test Results

| ALLOY | TEST TEMP. °F | BENDING MODULUS AUG. TEST VALUE KSI | TEST / PREDICTED |
|---|---|---|---|
| 4340 MOD. | RT | 385.0 | 0.73 |
| | 600 | 397.3 | 0.83 |
| AM 355 | RT | 396.3 | 1.0 |
| | 900 | (1) | |

NOTE: (1) DEFLECTION OF AM – 355 BARS TESTED AT 900 ° F WAS OF SUCH A MAGNITUDE THAT THE BARS EXCEEDED THE JIG CAPACITY. AT TIME TESTS WERE DISCONTINUED THE BENDING MODULUS WAS EQUAL TO OR EXCEEDED THE PREDICTED VALUE.

2 x ¾ x 22 in. lug and bend specimen. The tensile test results obtained are summarized in Table 1. A total of 40 specimens representing a minimum of five heats were tested for each alloy and temperature. These data represent the optimum tensile properties expected for the two alloys. Due to the tempering temperature required to obtain high strength levels with the A.M.-355 alloy, the environmental temperature requirement was lowered from 1000 F to 900 F.

It is evident from the data on the modified 4340 alloy that the elongation developed was not sufficient to meet the 4 per cent minimum required. It was decided to continue the tests, and depend on the notched specimens and the lug specimens to evaluate the true effect of the low elongations on the design properties.

Tension-tension fatigue tests at a stress ratio of 0.1 were conducted on the notched specimens, $K_t = 3.0$. These results are shown in Fig. 11. These results compared favorably with typical data for wrought alloys of comparable strength levels and stress concentrations. Apparently the low ductility of the modified 4340 alloy did not significantly affect the fatigue characteristics, at least for this type and severity of loading.

The bend tests were conducted, as shown in Fig. 12, and summarized in Table 2. In this instance the presence of small surface cracks, later traced to grinding, accounted for the majority of the values below the theoretical minimum based on the tensile test results. It was of interest to note that although these cracks were readily visible when the bar was under load, and after failure they were not detected by any of the normal inspection techniques including magnetic partical inspection, dye check or x-ray.

The lug tests were conducted, as shown in Fig. 12, and summarized in Table 3. In this case the results

Table 3 — Lug Test Results

| ALLOY | TEST TEMP. °F | LUG $F_{TU}$ AVG. TEST VALUE KSI | % EFF. (1) |
|---|---|---|---|
| 4340 MOD. | RT | 203.4 | 74 |
| | 600 | 212.3 | 86 |
| AM 355 | RT | 202.9 | 93 |
| | 900 | 145.9 | 95 |

(1) EFFICIENCY = $\frac{\text{AVERAGE LUG TEST VALUE}}{\text{AVERAGE } F_{TU} \text{ FROM TENSILE TESTS}}$

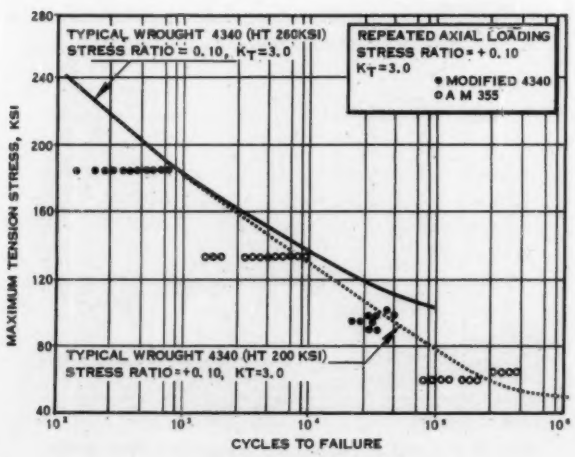
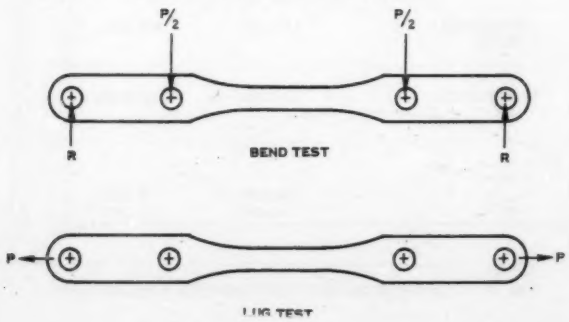



Fig. 11 — Fatigue life of cast 4340 modified and A.M. 355.

Fig. 12 — Bend and lug test loading systems.



reflect a significant effect on the modified 4340 alloy as a result of its low ductility. Tests on wrought alloys of this same general strength level produced efficiencies exceeding 95 per cent, whereas the low ductility cast alloy indicated efficiencies as low as 60 per cent.

In summary, the tensile strength levels achieved with the two alloys were sufficient to meet the program objectives with the exception of the low elongation of the modified 4340. It was decided to continue the development of this alloy during the following phase of the program.

## PHASE III

### *Foundry Control and Process Control*

Each foundry was supplied a drawing of the part to be produced and requested to submit a marked up drawing indicating recommended changes. These changes were reviewed with respect to the part function, space available in the airframe and program objectives. Those changes representing the best compromise between the above factors and foundry process limitations were incorporated. Tooling and pattern work was then initiated. The primary changes requested by the foundries amounted to a general increase in thicknesses to provide a taper or padding in order to promote directional solidification, avoid misruns and provide for machining in the critical tolerance and finish areas.

There were several reasons why all of the changes desired by the foundries could not be incorporated. In addition to the physical limitations, two basic factors had to be considered – (1) A major program objective was to produce castings requiring a minimum of machining and (2) the production parts to be replaced by the castings were machined forgings, therefore any extensive machining on the castings would seriously reduce, if not eliminate the economic advantage of the castings.

The final configurations agreed upon contained several padded areas where the foundries felt that soundness would be impossible without the padding and/or a taper. In the case of the fin beam and bulkhead casting the complete casting was essentially beefed up on its external surfaces. The idea being to establish foundry control and then remove material gradually on succeeding trials.

### *Foundry Control Problems*

The major problems encountered in establishing foundry control were the usual ones of shrinkage, hot tears, cracks and mis-runs. It would be difficult to designate which defect was more prevalent, or that any one defect was more prevalent in one process or the other. The shape and complexity of the castings was the primary factor rather than the process. In general a sound casting, i.e., one acceptable by x-ray standards and meeting all other standard inspection criteria, was obtained within six to ten trials.

The procedure followed was to pour a heat containing one to three castings maintaining a given set of variables constant and including a set of keel blocks, one to three, with each casting. The castings were processed only to the point where an obvious unacceptable defect was discovered. The specimens machined from the keel blocks were fully processed in order to maintain control of the mechanical properties. In this manner, it was also possible to continue the alloy investigation of the 4340 alloy to improve its ductility during the time period used to establish foundry control.

The first major problem that required solution was the inadequacy of the pouring temperatures and melt practices established for the previous phase. Usually a higher pouring temperature and/or a higher mold preheat temperature was required to prevent mis-runs. This resulted in a change in chemistry which resulted in a change in mechanical properties. This is illustrated in Table 4.

**Table 4 — Tensile Specimens from Castings Compared to Tensile Specimens from Keel Blocks**

| 4340 MODIFIED | | | | | | | |
|---|---|---|---|---|---|---|---|
| CASTINGS | | | | | KEEL BLOCKS(1) | | |
| CASTING NO. | SPECIMEN NO. (2) | $F_{TU}$ KSI | $F_{TY}$ KSI | %e | $F_{TU}$ KSI | $F_{TY}$ KSI | %e |
| 37 | 1 | 269.2 | 230.9 | 4.0 | 275 | 237.1 | 1.8 |
| | 2 | 252.4 | 228.9 | 1.0 | | | |
| | 3 | 210.7 | | | | | |
| 40 | 1 | 268.2 | 229.2 | 4.0 | | | |
| | 2 | 254.0 | 226.2 | 4.0 | | | |
| | 3 | 214.1 | | | | | |

(1) AVERAGE TEST VALUES.
(2) TENSILE SPECIMENS 1, 2 AND 3 TAKEN FROM CASTING SECTIONS WHICH WERE THICK TO THIN RESPECTIVELY

The second problem encountered was quite similar to the first, in that heat treat practices required modification to adjust for size and complexity of the castings. Again a trial and error solution was used until the final product was the same as the test bars from a metallographic and a mechanical property viewpoint. Essentially the final chemistry and heat treat procedure reported previously for the alloys were those established in this portion of the program rather than those found adequate for the test bars. Closely associated with the above was the problem of contamination, oxidation, carburization and decarburization due to inadequate control of heat treat atmosphere. There seemed to be a prevalent idea among the heat treat shops that all castings were machined therefore accurate controls were not essential.

### *Fin and Bulkhead Castings*

In the case of the fin beam and bulkhead casting which was beefed up, once a sound casting was produced it became readily evident that economic factors were going to prevent the removal of sufficient material to meet the program objectives. The first sound casting weighed on the order of 250 lb, approximately 150 lb more than the final desired part. There was a natural reluctance on the part of the foundry to remove any significant amount of material from any one trial. It was decided to take the bull by the horns and produce a pattern that would result in a



Fig. 13 — Carburized surface of A.M. 355 in shell mold.

casting conforming to the final part dimensions. This would use the opposite philosophy of adding material only where it was needed.

In only one trial it became obvious that this latter approach was going to produce the desired result with much less effort than the former. In the final analysis the casting weighed on the order of 125 lb. This means approximately 125 lb would have to be removal using the first approach whereas only 25 lb had to be added using the second approach.

The major delays from this point forward were associated with one of two factors – (1) mold-metal reaction and (2) mechanical property variations resulting from process changes required to correct physical defects such as tears and mis-runs. The term sound casting has been used to describe a casting whose only defect was visible and x-ray quality was otherwise acceptable. The first of these two problems quickly became the greatest single deterrent encountered in the program.

As a matter of interest this problem was not discovered until full scale testing of the castings was initiated, or when coupons cut from the castings were tested without machining off the cast surface. Early in the program, the problem of decarburization when the 4340 alloy was cast in ceramic molds was discovered by hardness checks made on the early castings. This was not as simple as it sounds, since the use of Rockwell hardness readings on an as-cast surface were continually suspect. The first checks made were obtained by machining smooth surfaces on the part and taking hardness readings. This of course did not reveal the deficiency. After visual examination of several premature failures on as cast test bars a metallurgical examination was made. This revealed that decarburization was present to various depths, dependent on location and thickness.

### *Carburization Problem*

A second similar problem was the carburization that occurred when the A.M.-355 alloy was cast in shell molds (Figs. 13 and 14 illustrate the problem encountered). In both of these cases, attempts were made to correct the condition by use of carburizing or decarburizing heat treat atmospheres. In both cases, this solution was not successful. For the 4340 alloy the carbon was restored to the desired levels but the mechanical properties were not recovered. This is illustrated in Table 5. For the A.M.-355 the pick up in carbides and impurities during the decarburizing heat treat cycle prevented any significant improvement in mechanical properties. It is of interest that in both cases the condition was corrected to the point where the normal inspection criteria did not indicate any deficiency.

Table 5 — Test Results of Cast Test Bars with Carbon Restoration Cycle

| | $F_{TU}$ KSI | $F_{TY}$ KSI | % e | HARDNESS ROCKWELL C | | | |
|---|---|---|---|---|---|---|---|
| | | | | SURFACE | .005" | .01" | .015" |
| AS CAST SURFACE | 232.5 | — | 0 | 49 | 49 | 49 | 49 |
| ON TEST BARS | 194.1 | — | 0 | | | | |
| WITH .006" | 207.6 | — | 0 | 49 | 49 | 49 | 49 |
| MACHINED OFF SURFACE | 226.1 | — | 0 | | | | |
| WITH .05" | 278.5 | 240.2 | 2.0 | 49 | 49 | 49 | 49 |
| MACHINED OFF SURFACE | 276.8 | 248.4 | 2.0 | | | | |

(1) ALL BARS WERE SUBJECTED TO CARBURIZING HEAT TREAT AS RECOMMENDED FOR DECARB CONDITION PRESENT IN ORIGINAL TEST BARS.

(2) CARBON ANALYSIS FROM DRILLINGS TO DEPTH OF .005 WAS MAX. 0.44, MIN. 0.38; FOR DRILLINGS FROM .005 TO .010 IT WAS MAX. 0.44, MIN. 0.39 FOR CORE MATERIAL MAX. 0.44, MIN. 0.38.

Attempts were then made to correct the deficiency by reducing the time during which the molten metal is in contact with the mold. This was attempted by reducing the pouring temperature to an absolute minimum, by reducing the mold mass to a minimum and by variation of mold pre-heat where applicable. Partial success was achieved, but mechanical properties were still low. This led to the necessity of developing new mold compositions to eliminate or reduce the elements contributing to the reaction. Due

to the time schedule, funds available and processes used this effort was quite limited. Some improvements were achieved but not complete elimination.

In the case of the sand process the major problem was the tearing or cracking. The solution to this appeared to be in developing a mold with sufficient collapsibility to prevent the tearing or cracking. Attempts to achieve this disclosed that the collapsibility could be increased to the point where the mold could not be handled without eliminating all of the tearing. Some limited work was done on modification of the binder to improve the hot strength or cold strength of the mold depending on the analysis of the resulting castings. This was continued to the practical limit without complete elimination of the problem.

At this time it became evident that if this casting was going to be produced such that it would meet the program objectives, some process other than the sand process would be required. An attempt was made to produce this casting by means of a ceramic mold rather than a sand mold. One trial using this process was made with apparent success in eliminating the tears and/or cracks. A second approach would be to increase the casting thicknesses so that enough strength could be developed to prevent tearing. Unfortunately, further work had to be terminated due to lack of funds.

### *Vacuum Degassing*

During this phase of the program, vacuum degassing equipment was installed at two of the foundries. This resulted in a program to modify the melting and pouring techniques to incorporate vacuum degassing and still obtain the same alloy limits, as established in the preceding phase for the two alloys involved.

Basically the installations consisted of a large vacuum chamber with an induction furnace mounted inside. The order of melt additions, the temperature at which additions were made, the time at which the vacuum was applied, the temperature during the vacuum cycle and the length of time for the vacuum cycle were varied to obtain trends from which an optimum cycle could be obtained. For each heat during this evaluation a group of keel blocks was poured to obtain mechanical property data for comparison with the data obtained during the preceding phase.

Two limiting factors were discovered – (1) the rapidity with which the chamber could be evacuated without exploding the melt out of the furnace and (2) the vacuum level that could be achieved for the A.M.-355 alloy without lowering the nitrogen level below the limits required to obtain optimum strength levels. The first of these limits could be overcome by practice, but the latter limited the vacuum (approximately 5 to 10 mm of Hg) that could be employed with A.M.-355 alloy. For the 4340 alloy vacuum levels as low as 100 microns could be reached without difficulty. However, consideration of the time the melt was at temperature led to the choice of a 5 min cycle (10 to 20 mm of Hg) for the former alloy and a 20 min cycle (20 mm of Hg) for the latter.

Results of mechanical property tests did not disclose

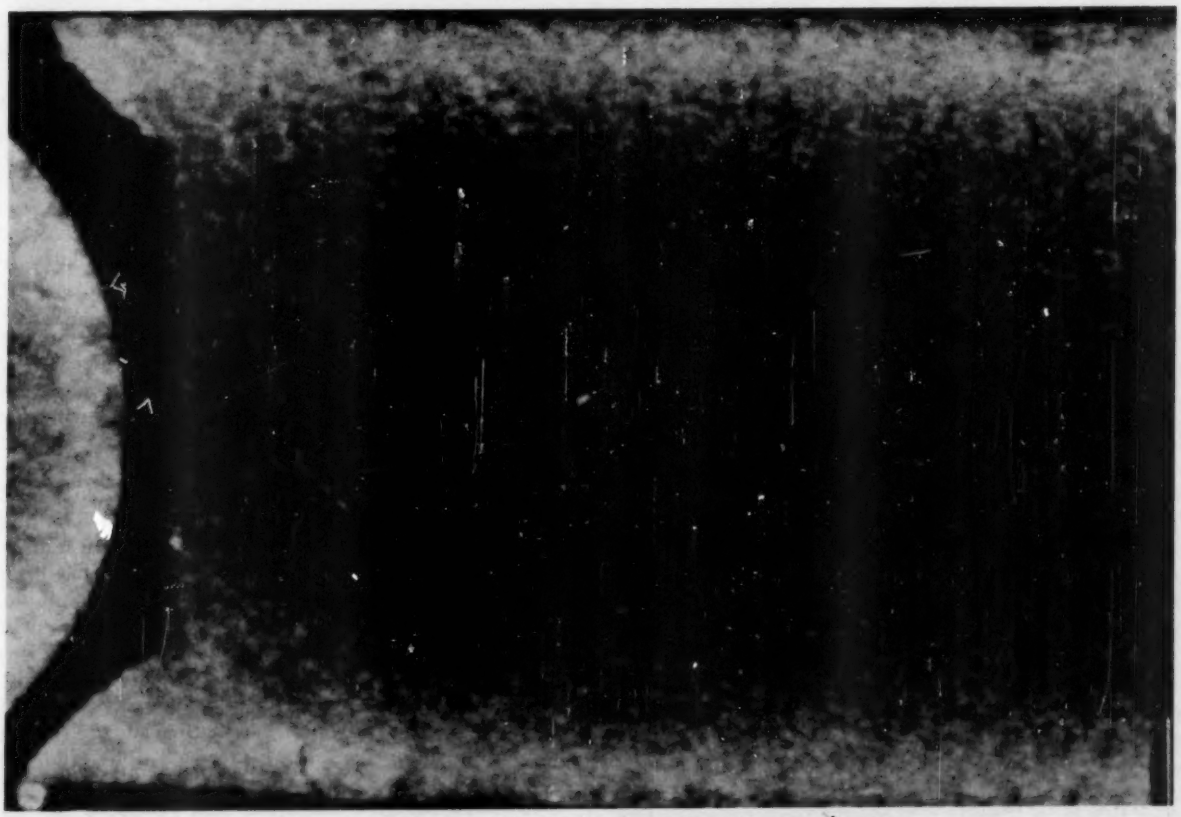


Fig. 14 — Decarburized S.A.E. 4340 modified cast in ceramic mold.

Table 6 — Results of Tensile Tests on Specimens Cut from Initial Castings

| CASTING NO. | TEST TEMPERATURE °F | $F_{TU}$ KSI | $F_{TY}$ KSI | % e |
|---|---|---|---|---|
| 47 | 75 | 217.5<br>219.1<br>204.9<br>205.2 | 171.6<br>181.7<br>171.9<br>172.4 | 13.0<br>13.0<br>12.0<br>8.0 |
| | 900 | 164.0<br>156.0 | 115.8<br>116.8 | 11.0<br>8.0 |
| 48 | 75 | 214.7<br>217.4<br>204.5<br>202.2 | 177.3<br>177.2<br>176.8<br>171.8 | 11.0<br>12.5<br>10.0<br>10.0 |
| | 900 | 148.2<br>148.7 | 113.8<br>113.2 | 10.0<br>10.0 |
| 49 | 75 | 219.0<br>217.0<br>217.6<br>221.9 | 173.3<br>173.0<br>176.9<br>178.9 | 12.0<br>12.0<br>14.0<br>14.0 |
| | 900 | 174.7<br>171.8 | 120.2<br>117.3 | 10.0<br>10.0 |

any significant effect due to vacuum degassing. Gas analyses were conducted on both alloys for hydrogen content and no significant difference was evident. In fact all samples evaluated were found to contain less than 6 ppm regardless of whether or not they were vacuum degassed. However, a significant improvement in microstructure was evident from the metallographic studies made. Why this did not result in an improvement in mechanical properties is unknown, unless it be that the experimental error for the relatively small number of tests is greater than the gain resulting from vacuum degassing.

### *Casting Evaluation*

Once the final processes were established three castings of each part were produced and submitted for evaluation. These castings were subjected to all recognized inspection criteria, radiographic examination and dimensional checks. Castings accepted for evaluation met all the applicable requirements in existing steel casting specifications in addition to those special requirements established for this program.

As originally planned, the evaluation was to be made by cutting specimens from these castings for test. This was done on the initial castings submitted, and the results are shown in Table 6. Since these results were acceptable, work was initiated on the trial production phase. Unfortunately then the castings produced in this latter phase were subjected to full scale testing they failed to meet the design requirements by a considerable amount. This is discussed more completely.

As a result of this it was concluded that the only reliable way to qualify the castings was to conduct a full scale test, and this was incorporated as a requirement for this phase of the program. The results of the castings tested in this manner are shown in Table 7.

These results indicate that it was possible to produce castings whose strength characteristics almost duplicated the results obtained in the alloy development phase.

Table 7 — Static Test Results for Cast A.M. 355 Drag Links

| CASTING NO. | ESTIMATED FAILING LOAD (1) POUNDS | TENSILE SPECIMEN $F_{TU}$ KSI | TENSILE SPECIMEN $F_{TY}$ KSI | RESULTS % e | ACTUAL FAILING LOAD POUNDS | STRESS (2) KSI |
|---|---|---|---|---|---|---|
| 55 | 137, 000 | 201.5<br>202.4<br>200.1<br>201.6 | 169.3<br>170.3<br>160.7<br>157.2 | 12.0<br>12.0<br>7.0<br>14.0 | 101, 825 | 92.5 |
| 60 | 137, 000 | | | | 97, 000 | 88.2 |
| 63 | 137, 000 | | | | 84, 970 | 77.3 |
| 102 | 150, 000 | 241.2 | 214.1 | 11.3 | 154, 430 | 215.4 |

NOTES:
(1) LOAD BASED ON 200, 000 PSI AND NOMINAL AREA.
(2) STRESS CALCULATED AT FAILURE AT THE SECTION WHERE CASTING FAILED.
(3) CASTING 102 WAS PRODUCED AFTER INCORPORATION OF PROCESS CHANGES DICTATED BY CASTINGS 55, 60 AND 63.

## PHASE IV

### *Trial Production*

Initially in this phase ten castings were to be produced representing a minimum of three heats. These castings were to be static tested in a manner simulating the actual design loads on the parts. In addition, sample production parts were to be tested for comparison purposes. As a result of the premature failure of the initial castings submitted for test a significant number of process modifications were required to correct the deficient condition.

Essentially the modifications involved re-location of several risers, change in riser sizes and changes in pouring techniques. All of these changes were aimed at reducing the thermal gradients in the casting during solidification. Once these changes were incorporated and proved to be effective, the series of ten castings were poured and processed.

For the two foundries still participating in the program, the inspection revealed that for one part, nine acceptable castings were produced out of the ten poured, and for the other a total of 13 castings were poured to produce ten acceptable castings. This indicated a scrappage rate which was somewhat higher than expected. A careful review of the defective castings disclosed the fact that with one exception the deficiencies were directly traceable to mechanical errors by one or more persons during the production of the castings. All of these errors occurred during the molding operations.

It was only logical that the severe thickness transitions complex shapes and tolerances required would impose more rigid mold fabrication and handling restrictions than were common practice. In almost every case the error had been noted, but it has been assumed that it would not be important. For example, in fabricating one mold a relatively small section of one core was broken off during handling. Since this was a relatively assessible area it was decided to use this broken core and plan to grind off the small amount of excess material that resulted during the cut off and cleaning operation. This was done, and the casting processed. When this casting was inspected it was found that the small hot spot produced by this excess material resulted in a tear at this location.

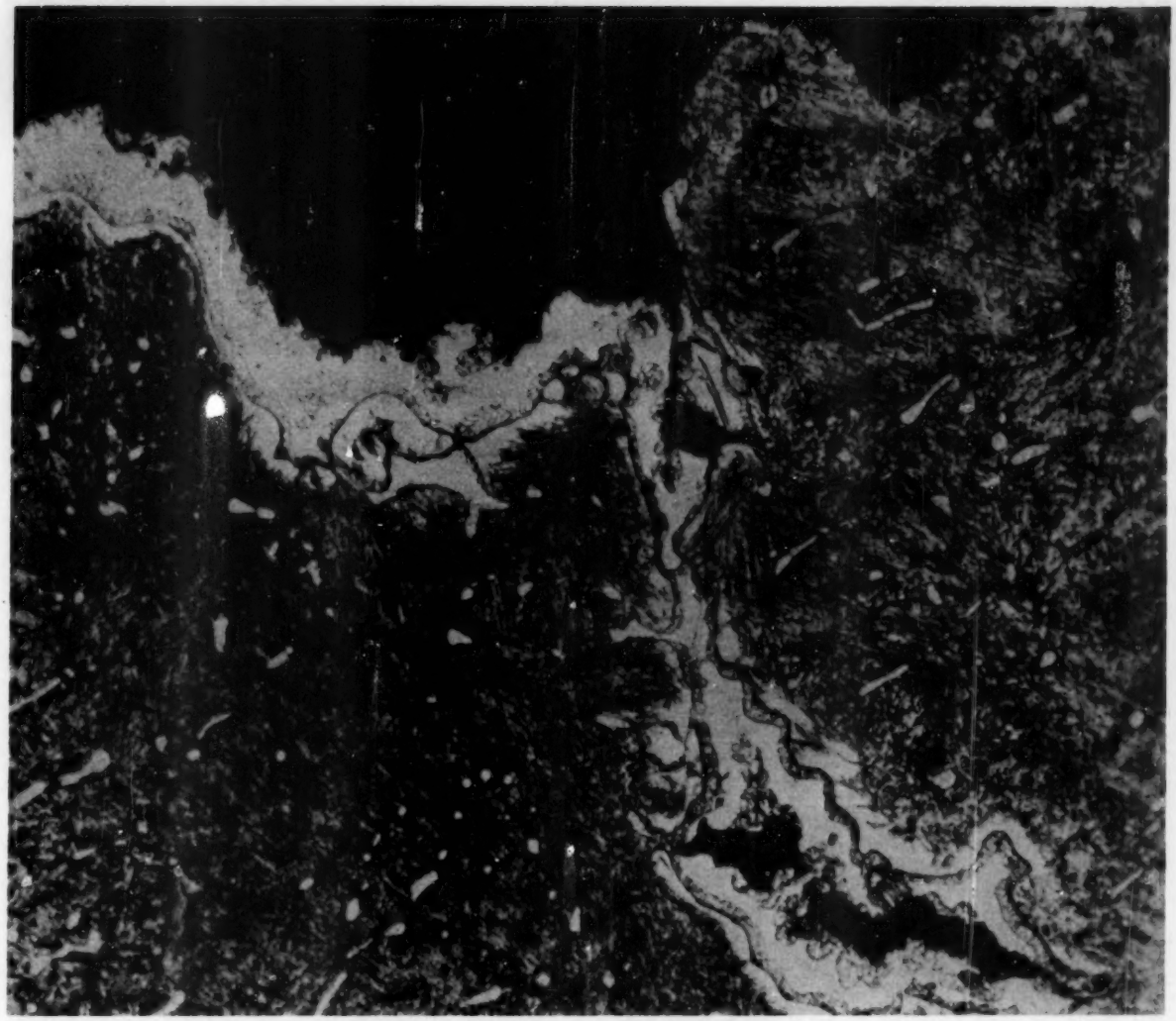


Fig. 15 — Failed section of A.M. 355 drag link casting.

*Casting Evaluation*

The results obtained on the first three drag link castings were significantly lower than expected (Table 7). Examination of the failed castings revealed a severe band in the casting that can be described as a structure consisting primarily of an austenite band with large grain boundary precipitates, as shown in Fig. 15. Mechanical property specimens taken across this area revealed strength levels on the order of 60,000 to 90,000 psi.

A review of the process records revealed that a magnetic particle indication had been present during inspection, however microscopic examination revealed that no crack was present. Furthermore, it was located in a noncritical section and therefore it was assumed that the condition would not seriously affect the strength. Review of the other castings revealed similar bands although not as severe as this one.

A complete step by step review and analysis was made for the castings checking after each step by metallographic means. It was found that this band appeared on the castings from the beginning. Several hypotheses were advanced on the causes—for alloy A.M.-355, hot tears that were refilled since the section was immediately underneath a heavy riser, thermal gradients within the casting and lack of mold collapsibility. All of these were investigated by means of trial pours. The alloy was not at fault, since the same casting poured from other alloys, 4340, 410 and 17-4 also had the band present. Increasing the mold collapsibility within the limits possible produced no significant change. It was finally concluded that the cause must be due to the hot tears and/or in combination with the thermal gradients and mold rigidity.

As a result a complete new process was evolved which reduced the thermal gradients by relocating the sprues, shifting the risers and reducing the riser sizes. This effectively eliminated the problem in that the banding was broken up so that it appeared randomly scattered and to a much less degree than

before. Similar changes in processes were required for the feedback casting to eliminate the same defect.

The castings required for this phase were produced again and re-submitted for evaluation. The results obtained are summarized in Table 8. From these results it is evident that it was possible to obtain reproducible and predictable results for these castings. It is of interest to note the differences in tensile strength as obtained by the different methods. Likewise a difference in strength levels can be noted for the two castings. Additional testing also indicated that similar variations exist from one portion of the casting to the next.

Table 8 — Results of Static Tests on A.M. 355 Castings

| PART | CASTING NO. | TEST TEMP. °F | FAILING LOAD–KIPS | | FAILING STRESS KSI |
|---|---|---|---|---|---|
| | | | PREDICTED(3) | ACTUAL | |
| DRAG LINK | 97 | 80 | 150 | 146 | 192.5 |
| | 102 | 80 | 150 | 154.4 | 215.4 |
| | 92 | 600(1) | 126 | 149.9 | 187.0 |
| | 100 | 900(1) | 119 | 129.6 | 184.0 |
| | 105 | 900(2) | 119 | 137.6 | 196.6 |
| FEEDBACK BEAM | 75 | 80 | 14.1 | 14.4 | 204.0 |
| | 74 | 80 | 12.3 | 12.8 | 208.5 |
| | 71 | 600(1) | 10.9 | 11.6 | 179.5 |
| | 67 | 900(1) | 10.3 | 11.3 | 174.3 |
| | 66 | 900(2) | 10.6 | 11.8 | 176.8 |

NOTES
(1) 1/2 HOUR SOAK AT TEMPERATURE BEFORE TEST
(2) 1 HOUR SOAK AT TEMPERATURE BEFORE TEST
(3) BASED ON NOMINAL AREAS, TYPICAL PROPERTIES FOR DRAG LINK AND 200, 000 PSI $F_{TU}$ FOR FEEDBACK BEAM.

## PHASE V

### *Casting Production*

Using the procedures established in the preceding phase a production run of 30 castings of each part was made by each foundry. Normal production personnel and practices were used throughout this effort. It must be noted that controls proved to be essential in the prior work were incorporated prior to initiating this phase.

The castings represented a minimum of ten heats. For this phase inspection revealed that three of the 30 drag link castings, and five of the 30 feedback beams were unacceptable. This indicates a scrappage rate which is still somewhat high, but as before most of the rejects could be traced to personnel errors rather than process deficiencies.

The greatest single problem during the production phase was the convincing of personnel that the degree of control set up on the process sheets was necessary. It became almost necessary to permit a casting to be processed without rigid controls to prove they were required. This factor was even more difficult to understand when there was no significant cost or time increase necessary to obtain the more rigid controls.

On receipt of the finished castings a rigid examination for conformance to all inspection criteria established during the preceding phases was performed. Rejection was based only on those defects proved by previous work to be detrimental to the casting performance. As a result of the rigid inspection by both the foundries and the aircraft company inspector one factor emerged—the normal human element will result in the tendency of the foundries to accept castings which the aircraft company will reject. Neither of these inspections will be 100 per cent correct, i.e., some castings accepted by the foundries will fail prematurely and some castings rejected by the aircraft company will perform satisfactorily.

### *Inspection Rejects*

The only thing common to both inspections was that gross defects were rejected by both. Therefore, it can readily be established that the primary cause for casting malfunctions are not gross defects. This was borne out by this phase of the program, in that the defects that resulted in low strengths were not detectable by conventional inspection criteria. In most cases they were found only by some form of destructive test. The use of the conventional inspection criteria was no better than the judgment used in interpreting the results. A typical defect observed during this phase is shown in Fig. 16.

Since the program called for these castings to be used without machining a careful study was made of dimensional tolerances and surface finish. Castings were selected at random from the production lot and random dimensions measured as well as determining the surface finish of random areas. The data for a typical dimension and surface are summarized in Fig. 17. The principal result was that the dimensional tolerances possible were much less than those usually requested by the foundries. The reason for this is fairly obvious in that the final part dimension is affected by two factors.

First, there will be the error related to the pattern caused by complexity and/or incorrect shrink rule. Second, there will be the variation from one part to the next as a result of normal process variables. The combined allowance for both of these factors usually leads to the foundry request for relatively wide tolerance ranges. If provision is made for pattern changes in the original design then the final tolerances required for the casting will be relatively small.

In the case of surface finish, care and control of the mold fabrication, handling and composition can produce surface finishes comparable to many machine surfaces as presently used on aircraft parts. In the case of these castings produced by the shell and ceramic process, a satisfactory surface was produced with the exception of those limited areas where mating part requirements imposed extreme finish and tolerance requirements.

Ten castings were selected at random from each production lot for static and fatigue testing. These castings were re-examined carefully to insure that they were representative of the lot insofar as could be determined by nondestructive inspection. The results obtained are presented in Tables 9 and 10. The results obtained were not as consistent as had been expected. There were several hypotheses advanced for this fact. The principal one being that in modifying the processes to eliminate the banding discussed earlier several detail changes, not previously evaluated,

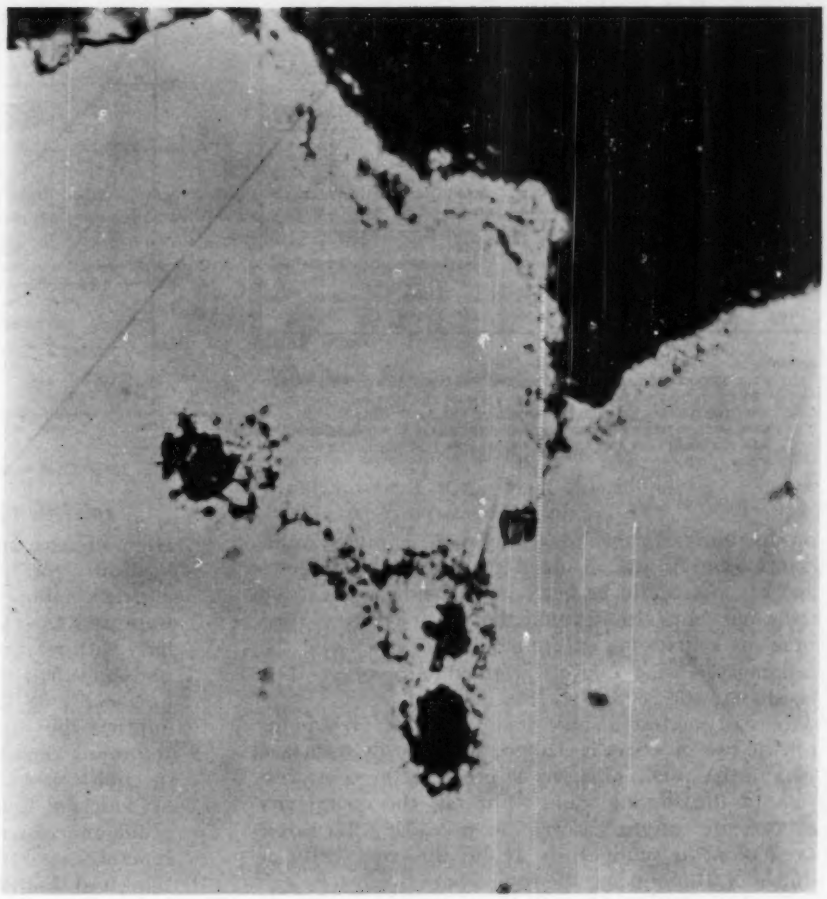


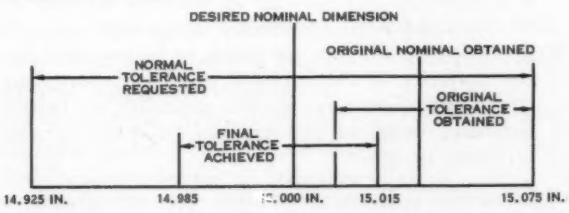
Fig. 16 — Typical defect.

were introduced. Apparently these did have a slight effect on the mechanical properties.

Another item noted was that the strengths developed in the castings were not as high as those developed by test bars during the alloy development phase. This was not entirely unexpected, since it had already been determined that casting strength could not be predicted by cast test bars regardless of whether they were separately or integrally cast. In some random cases due to a fortunate combination of casting dimensions, riser location, sprue location, gate locations, pouring conditions, etc., it would be possible to use cast test bars to predict casting strength but this would be the exception rather than the rule.

As a final step in this program a comparison was made between the casting costs and the cost of the

DIMENSIONAL TOLERANCES




SURFACE FINISHES

| CASTING | AREA | FINISH — RM S | | |
|---|---|---|---|---|
| | | MIN. | MAX | TYPICAL |
| DRAG LINK | LUG | 75 | 90 | 85 |
| | CHANNELS | 80 | 90 | 85 |
| | PIVOT TUBE | 80 | 100 | 95 |
| FEEDBACK BEAM | WEB FLANGES | 60 | 100 | 80 |
| | | 70 | 90 | 85 |

Fig. 17 — Typical dimensional and surface finish data.

Table 9 — Static Tests of Production Lot A.M. 355 Castings

| PART | CASTING NO. | TEST TEMP. °F | FAILING LOAD~KIPS | | FAILING STRESS KSI, (2), (3) |
|---|---|---|---|---|---|
| | | | PREDICTED(1) | ACTUAL | |
| DRAG LINK | A–19 | 80 | 178.0 | 156.5 | 194.6 |
| | A–28 | 80 | 153.0 | 144.0 | 211.0 |
| | A–30 | 800 | 128.0 | 132.9 | 196.0 |
| | A–31 | 900 | 125.0 | 127.3 | 158.3 |
| | A–24 | 1050 | 103.0 | 106.1 | 156.5 |
| FEEDBACK BEAM | 83 | 80 | 13.3 | 17.7 | 224.0 |
| | 84 | 80 | 15.6 | 18.7 | 212.2 |
| | 81 | 600 | 11.7 | 13.4 | 161.6 |
| | 86 | 775 | 12.2 | 17.3 | 210.4 |
| | 80 | 900 | 11.0 | 15.6 | 209.0 |

NOTES:
(1) PREDICTED FAILING LOADS ARE BASED ON AVERAGE $F_{TU}$ OBTAINED FROM COUPONS CUT FROM THE CASTINGS.
(2) APPARENT FAILING STRESS FOR THE DRAG LINK BASED ON PURE AXIAL LOAD AT THE FAILED SECTION.
(3) APPARENT FAILING STRESS FOR THE FEEDBACK BEAM BASED ON BENDING PLUS TENSION AT THE FAILED SECTION. THE STRESSES SHOWN ARE APPROXIMATE SINCE PRECISE DEFLECTION DATA WAS NOT AVAILABLE.

Table 10 — Fatigue Tests of A.M. 355 Castings

| PART | CASTING NO. | TEST TEMP. °F (1) | TEST LOADS KIPS | | CYCLES TO FAILURE | |
|---|---|---|---|---|---|---|
| | | | HIGH | LOW | ACTUAL | PREDICTED (2) |
| DRAG LINK | A–13 | 80 | 100.0 | 10.0 | 4103 | 6500 |
| | A–21 | 80 | 100.0 | 10.0 | 6209 | 5300 |
| | A–23 | 600 | 100.0 | 10.0 | 1197 | 2250 |
| | A–11 | 900 | 85.0 | 10.0 | 2246 | 5300 |
| | A–27 | 900 | 85.0 | 10.0 | 5750 | 4200 |
| FEEDBACK BEAM | 78 | 80 | 12.0 | 1.84 | 2787 | 1180 |
| | 104 | 80 | 12.0 | 1.84 | 5900 | 2000 |
| | 82 | 600 | 12.0 | 1.84 | 1672 | 1180 |
| | 115 | 900 | 10.93 | 1.84 | 3551 | 2180 |
| | | 900 | 10.93 | 1.84 | 1446 | 1560 |

NOTES:
(1) HELD 1 HOUR AT TEMPERATURE BEFORE TESTING. TEST TIMES VARIED FROM 2 1/2 TO 6 1/2 HOURS.
(2) PREDICTED CYCLES BASED ON CURVES OF FIGURE 11 AND $K_T = 2.0$. STRESSES IN THE FAILED SECTIONS WERE BASED ON AXIAL LOADS FOR THE DRAG LINK AND AXIAL PLUS BENDING FOR THE FEEDBACK BEAM.

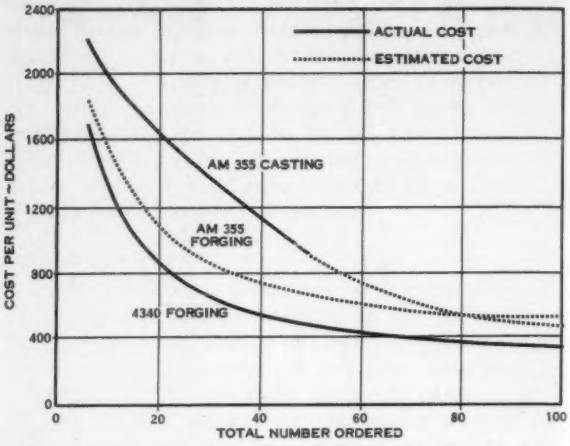



Fig. 18 — Cost comparison for forged vs. cast drag link.

production parts, machined forgings. To obtain valid comparisons it was assumed that both the forgings and castings would be heat treated and machined by the same shops. In addition the scrappage rates used were the actual ones encountered in this phase of the program and the forging production program. The results are shown in Figs. 18 and 19.

It is of interest to note that the break even point for the two processes is changed significantly from one part to the other. This can be predicted by consideration of the forging alone. That is, the complexity or difficulty of the casting did not affect the break even point as significantly as did the complexity or difficulty of producing the forged part.

## CONCLUSIONS

On the basis of the results and data obtained during the conduct of this program, the following conclusions were reached:

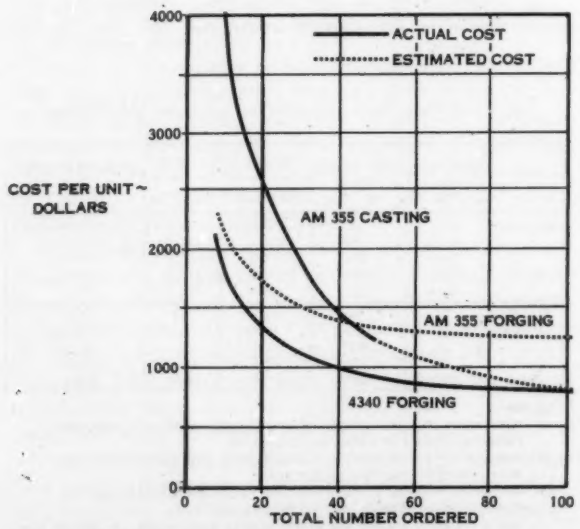



Fig. 19 — Cost comparison for forged vs. cast feedback beam.

1. High strength steel castings can be produced economically for use as highly loaded aircraft and missile components.
2. Improved knowledge of foundry processes, particularly with relation to casting alloys, mold compositions and their interaction are going to be required to avoid long and costly qualification trials and improve the competitive position of the castings.
3. Improved knowledge and understanding of the inspection criteria and their limitations is required of both the foundries and aircraft companies to avoid unnecessary and costly rejections and/or acceptances.
4. Improved knowledge and understanding by the foundry and aircraft industry of each other's design limitations will be required to solve in an economical manner many of the problems encountered.
5. There is no consistent relationship between the mechanical properties of a casting and the various defects revealed by nondestructive inspection, such as x-ray or magnetic particle. The grading of castings into classes by the number of defects has no relation to the final casting performance.
6. A highly loaded airframe component that is produced as a casting is an engineered product which will result in problems, delays and failures unless it is procured and handled in the same manner as all other expensive engineered components.
7. Quantitative answers are going to be required for the majority of the foundry process variables on which only qualitative answers are available, before extensive usage of castings can be made by the airframe designer.

In summary, it was determined that the objectives of this program were achieved only as a result of mutual understanding and cooperation by both the airframe designer and the foundry engineer. The solutions reached were generally the result of design compromises made by both rather than one alone. By application of this general approach both industries can obtain mutual benefits, the final result being an improved efficiency in the military weapons systems required for national defense.

# DISCUSSIONS

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61. R. Kottke is with Archer-Daniels-Midland Co., Cleveland.
62. O. J. Myers is Vice-Pres., Reichhold Chemicals, Inc., White Plaines, N.Y.
63. P. J. Neff is Asst. Mgr., Mfg. Rsch. Lab., American Steel Foundries, East Chicago, Ind.
64. C. M. Adams, Jr., is Assoc. Prof. Met., Massachusetts Institute of Technology, Cambridge, Mass.
65. E. M. Broad is Chief Engr., Hitchiner Mfg. Co., Milford, N.H.
66. C. Stanley is with Aerojet General Corp., Sacramento, Calif.
67. H. Gould is with Washington Iron Works, Seattle.
68. W. Giff is with Atlas Foundry, Cleveland.
69. J. T. Berry is with Armour Research Foundation of Illinois Institute of Technology, Chicago.
70. A. J. Kiesler is Rsch. Assoc., Met. & Ceramics Lab., General Electric Co., Schenectady, N.Y.
71. W. Koppi is Met. Engr., International Nickel Co., New York.
72. H. W. Lownie, Jr. is Div. Chief and A. J. Stone is Asst. Div. Chief, Battelle Memorial Institute, Columbus, Ohio.

## Brass and Bronze Division

**Weldable As-Cast Nickel Aluminum Bronze Composition and Microstructure,** by M. L. FOSTER and S. GOLDSPIEL *(paper printed beginning on page 80 of this volume).*

J. S. VANICK[1] *(written discussion):* The authors could be right about the ability for weldable aluminum bronze to tolerate 0.10 per cent lead and 0.16 per cent Si, and they could also be very wrong.

The study seems to indicate that certain microstructural components resist cracking while others accentuate cracking. The process for controlling the desired structure is not spelled out, and until it is, the hazard remains.

The movement of lead in the structure is not defined. Lead is rated as insoluble in the solid in most copper-base compositions, but it might be found to possess a limited greater solubility in one structural component as compared to another. The fundamental work of determining this solubility and its influence on hot shortness, for example, remains to be done.

Meanwhile the circumstantial evidence of the damage to weldability of lead in copper-base alloys remains. The evidence is supplied for nickel-base alloys, and Kihlgren's early work on the lead damage to cupro-nickel ductility at elevated temperatures has been repeatedly confirmed. Nickel-base alloys will crack along the line of a pencil mark, if lead is a constituent of the scribe. At such a surface the lead content may be high, in terms of per cent lead. For specification purposes, the base metal may be restricted to 0.01 per cent, and in many cases as little as 0.001 per cent, with the hope that the contamination will be less. In these alloys, lead is regarded as practically insoluble, and minute percentages can spread themselves over the grain boundary surfaces leading to hot cracking. Fine grained structures would possess a greater grain boundary surface and conceivably accommodate more lead, with less damage to weldability.

Until lead can be fixed by establishing its solubility in the several structural components or rendered harmless by agglomeration, and the methods for controlling these structures established, it would continue to be desirable to avoid the risk, by holding lead contents down to relatively harmless amounts.

AUTHORS' REPLY: Any comment from Mr. J. S. Vanick is a compliment, since it shows interest in a paper by a man who holds a high reputation among metallurgists and foundrymen throughout the country for many years. However, Mr. Vanick in the current case appears to over-emphasize an aspect of the authors' paper which was intended and admitted to be an incompletely explored, though interesting and important, side light. The main point of the basic paper is that control of the major alloying element of the alloy type, namely Al (which for the particular range of section thicknesses studied was shown to be from 9 to 10 per cent, when Fe, Ni and Mn are held at 4, 5 and 0.5 per cent, respectively), is most essential in assuring as-cast structures with consistent weldability. The influence of unfavorable Al content on microstructures in the as-cast condition is such as to produce an unweldable material, regardless of whether or not Pb is present. In retrospect, this conclusion explains some of the data not clearly understood during preliminary work, and based not only on the authors' own data but also those of a prominent industrial firm, which was kind enough to furnish the Material Laboratory with some samples (G through K, listed in Table 1 of the paper), as discussed in the resume of the preliminary portion of the work reported in the paper.

The mention of added experiments with wedges cast to optimum Al (with Fe, Ni and Mn at 4, 5 and ½ per cent, respectively) was to highlight the importance of good Al control. For this composition the wedges containing up to 0.10 per cent Pb and 0.18 per cent Si do not appear to affect welding adversely, as judged merely from the macroexamination of the butt-welded wedge castings. The planned work, in this connection, could, unfortunately, not be carried out to completion, because the wedge castings with Pb and Si additions were excessively drossy, principally due to the use of a new flux in the corresponding heats. As a result, even though the wedge plate welds were successfully completed, hot tests on cut bars were actually considered warranted, because the presence of dross in the base metal would probably mask the true tensile properties of the welded specimens. However, the work will be repeated with clean castings, at which time the specific effect of Pb and Si will be checked.

Pending confirmation of the results and better resolution of the mechanisms involved the authors have no intention of suggesting that Pb, Si or any other element, which traditionally has been kept out from related alloy types as harmful impurity, be tolerated except at levels as low as possible in economic utilization of ingot and remelt materials in casting production. While the work required was interrupted due to lack of funding during the past year, it is scheduled to be resumed in July of 1961. It is hoped that one of the results of this work will be a determination of the level of Pb and Si, which may be tolerated without significantly adverse effects on weldability or hot bending of the alloy type.

**Aluminum Bronze Hard Spots,** by N. A. BIRCH *(paper printed beginning on page 457 of this volume).*

R. A. FLINN[2] *(written discussion):* The author is to be congratulated for his careful definitive observations in a difficult problem. The observations that the hard spots are always round lead us to suspect that they may originate from the separation of a liquid, iron-rich phase in the liquid bronze. We have found a similar iron-rich phase separates from a homogeneous liquid magnesium, which was equilibrated with liquid cast iron. We note that carbon (which was also in solution in the liquid magnesium) is found in the iron-rich phase as iron carbide. The separation of these liquid iron pellets in the case of the aluminum bronze may be caused also by foreign elements in addition to iron, and the spots will be especially noticeable if the foreign elements harden the iron.

J. F. KLEMENT[3] *(written discussion):* Needless to say, the paper on hard spots of aluminum bronze is extremely interesting. Although this discussor's company probably manufactures more aluminum bronze than anyone in the country, we have not been confronted with the problem of hard spots in aluminum bronze as indicated by this group in discussion.

Listening to the various comments, it leads me to believe that the impurities mentioned are the primary source of these hard

spots, inasmuch as aluminum bronze alloys have an iron-rich phase that exists in the melt which upon cooling dissolves in a solid solution, and then again is thrown out of solution as cooling progresses. When the fact was mentioned by one of the foundries pouring aluminum bronze that the hard spots existed even after metal had gone through a strainer core, it is pretty evident that these hard spots must be liquid otherwise they could not go through a small strainer core, and then reform into sizes that are larger than the opening in the strainer core.

Inasmuch as iron exists in the melt, it is my suspicion that certain impurities have a greater attraction for the iron than the aluminum bronze material, and consequently interfere with the normal solidification pattern of this alloy, as is dictated by the constitutional diagram. Each foundry has their own type of impurity, and I do not believe the problem is limited to one element or another. It appears from this assumption that there must be at least three or four different elements that can cause this condition, so it appears the best solution to the problem is to guard against impurities in this type of material.

## Die Casting and Permanent Mold Division

**Maximizing Soundness in Aluminum Die Castings,** by L. M. ELIJAH *(paper printed beginning on page 328 of this volume).*

F. B. DIANA[4] *(written discussion):* We have on numerous occasions observed lack of fluidity in secondary smelters 380 Alloy ingots procured from various smelters (for die-castings). Will you kindly explain why two different lots of ingots from the same smelter vary greatly in fluidity for identical castings under identical operating conditions. The smelter's analyses are according to specifications.

D. L. COLWELL:[5] There are numerous factors in an aluminum die casting process which may affect the fluidity. Among them we might mention:

1. In SC84A alloy the silicon is specified as 7.5 to 9.5 per cent. This wide range opens up the possibility of variation within the specification of silicon content. Everything else being equal, a composition containing 7.6 per cent silicon will have different fluidity characteristics from one containing 9.4 per cent silicon both of which are within the specification. It has been our practice to maintain a silicon content almost on the button, and in a repetitive process such as the die casting process we would recommend that variables in metal composition be minimized by using ingot from a minimum number of sources with a much narrower specification.
2. Fluidity can be affected by plain dirty metal. Oxides and other nonmetals may be found in poorly cleaned ingot, or they may be introduced in the casting process by handling the molten metal with so much agitation that froth is developed or by the melting of oily and contaminated scrap. If central melting is practiced, and the metal given a proper fluxing and cleaning in the central furnace, the effect of dirty metal on fluidity can be minimized.
3. Temperature of course affects apparent fluidity. Metal temperatures should be maintained constant by hot charging from a separate melting source, and die temperatures should remain constant not only by keeping a uniform cycle but by controlling cooling water as well. The fact that cooling water temperatures vary during the day or with different weather conditions is a point often overlooked and this can affect fluidity.
4. Certain minor elements sometimes added either deliberately or accidentally, such as sodium and calcium, can affect fluidity by increasing the surface tension of the metal. These additions are not recommended for SC84A alloy. Variation in the magnesium content may also have a similar effect.

There may be other items which affect fluidity, but these serve to illustrate the necessity for uniformity in conditions of all elements in the repetitive die casting process.

MEMBER: How does high iron affect mechanical properties, and are there any additional elements to compensate for high iron?

L. M. ELIJAH: In answer to the first question – As a binary alloy iron forms $Al_3Fe$, however commercial alloys contain many other elements for better mechanical properties and castability reasons. It is therefore present as a complex compound. As the amount of iron increases (this point being lower in sand castings than in die castings) the iron-rich constituent commences to form large needles, which embrittles the casting. This is indicated by lower elongation values.

As to the second question – This embrittling effect can be neutralized in the standard aluminum alloys in these ways:

By increasing the cooling rate. Chilling tends to break up the large, iron-rich needles. For this reason, alloys containing one per cent Fe are used in die casting, the higher iron content decreasing die soldering.

By adding manganese, the coarse brittle needles are converted to the "Chinese Script" form. However, this method is only efficacious within limits.

Mainly only the effect of chilling is utilized for die castings. However, Beryllium can also be used to break up the long brittle needles into tiny globules as is done in sand and permanent mold castings. This is a relatively expensive process and hence is restricted. In general, the effect of beryllium on the iron-rich constituents is similar to the modifying effect of sodium on the silicon rich eutectic constituents.

**Aluminum Furnace Refractories,** by C. H. SCHWEINSBERG and J. L. DOLPH *(paper printed beginning on page 377 of this volume).*

F. L. ARNOLD:[6] The author stated that the melting temperature of the salt mixture used for impregnating furnace refractories is 1463 F. Was this mixture chosen deliberately to have a melting point somewhat above the normal aluminum molding temperature, and would harm be done if the metal temperature rose above 1463 F?

AUTHORS' REPLY: The salt mixture should have a melting point above the normal aluminum furnace holding temperature so the salt will be solid in the pores and joints during the operation of the furnace. If this were not the case, the salt mixture would run out and float to the top of the bath. Due to the cryolite, the salt mixture reacts with the brick to form a more refractory slag so a metal temperature slightly in excess of 1463 F would probably not harm the lining.

More complete information on the salt impregnation procedure is available in a paper by W. Helling and E. Kistermann in the German publication *Aluminium*, p. 513 (Aug. 1957).

**Fundamental Mechanical and Thermal Aspects of Die Casting,** by A. B. DRAPER *(paper printed beginning on page 832 of this volume).*

J. GRAHAM:[7] Is more work being done with the glass die?

A. B. DRAPER: Yes, more work is being done. Several companies are working with the die – one company working with glycerine and another company is using cerro-alloy. The author is also continuing his work with this die.

**Tilt Pouring Permanent Mold Castings,** by G. W. STAHL *(paper printed beginning on page 476 of this volume).*

J. G. MEZOFF:[8] If you were to build a new mold for this part, would you eliminate the side runners? If you would, why?

A. B. DRAPER: We believe that it would be possible to produce the subject casting without side runners, although part of it would probably show more flow lines. Savings in cleaning time would be substantial without side runners, and this would justify a trial of the procedure.

L. A. MCGILL:[9] Is metal fed to the casting as it is poured into the receiving reservoirs, or is metal fed to the casting from the reservoirs?

G. W. STAHL: Metal is fed to the casting from the reservoirs, and not from the hand ladle.

L. A. MCGILL: Is turbulence pock marking observed in the casting and also in the sprues?

G. W. STAHL: We have no interest in sprues, and consequently observation has not been made in this area.

**Porosity in Zinc Die Castings,** by G. P. LEWIS, D. A. CRAW and R. C. BELL *(paper printed beginning on page 537 of this volume).*

J. O. EDWARDS[10] *(written discussion):* I would like to endorse the remarks made by others on the quality of the paper. I think

that it emphasizes the importance of die design, in particular the gating and venting on the quality of the casting, and it would be interesting to extend the study to cold chamber die casting where there are substantial differences in say gating design.

I would like to comment on the influence of entrapped gas on shrinkage. If two interconnected areas of the casting contain high and low gas contents, respectively, on injection, as these areas freeze and contract, the pressure drop in the low gas area is greater than in the high gas area. Thus, metal is driven from the high gas area to balance the pressure, and the high gas area feeds the low gas area. Thus, the high gas content also concentrates the shrinkage in this area.

D. L. COLWELL[5] *(written discussion):* The authors are to be highly congratulated, not only on bringing a zinc-base die casting alloy paper to the Die Casting and Permanent Mold Division, but for presenting an entirely new approach to the question of porosity in die castings. One of the chief merits of the paper is that it suggests so many possibilities for additional work on such factors as the effects of temperature and others.

For example, the casting cross-sections shown were from castings made supposedly under the same conditions but at widely separated intervals of time. The gross difference between the equiaxed and the columnar structures indicated the differences in cooling conditions, which may have been due to an unlooked for variation in the temperature of the cooling water or some similar factor and such variations might have affected the results.

It is also worthy of note that the analyses of the gases in the porosity showed such a small effect of the lubrication. We know from experience that gross porosity in an aluminum die casting often contains so much lubricant that it can be smelt and that a section through such an area will often develop droplets of oil. Perhaps this points to the fact that an aluminum die casting made in a cold chamber machine picked up this oil from the plunger lubricant rather than the die lubricant because the zinc die casting was made in a hot chamber machine. Anyone who has seen the careless application of plunger grease in a plunger well can well understand how such porosity would be possible.

It is to be hoped that the authors will be able to make further contributions along the lines of this excellent paper.

AUTHORS' REPLY: We would like to thank Mr. Edwards for his comments on the paper. His observations regarding the effect of a localized high gas content on shrinkage porosity in a portion of the casting are interesting, and we would agree with his reasoning. This effect would also help to explain the existence, in some sections of the casting, of unfed solidification shrinkage in excess of the theoretical solidification shrinkage for the alloy.
MEMBER: Has much work been done to assess the value of vacuum in die casting?

AUTHORS' REPLY: We have thus far done only a limited amount of work in this field. In a single test it was found that the use of 21 in. of vacuum resulted in the reduction of the gas content of the casting by about 25 per cent or from 4.95 to 3.68 cc per 100 g. The volume per cent voids decreased from 1.69 to 1.44. It should be pointed out, however, that the casting examined was not well suited for an evaluation of the possible beneficial effects of vacuum since the casting produced without the use of vacuum was of good quality and had a low porosity level.
MEMBER: Would you comment on the fact that die lubricant even in excess quantities had virtually no effect on the overall porosity of the castings?

AUTHORS' REPLY: In aluminum die castings, lubricants appeared to have a more significant effect, and porosity has been observed which was definitely due to lubricant. This could possibly be attributed at least in part to lubricant applied to the plunger tip and shot sleeve. We would also emphasize that while we observed no increase in overall porosity due to the use of excessive lubricant, we did observe increases in gas content in some sections of the castings and it is likely that excessive lubricant could, in some cases, have a detrimental effect.
MEMBER: What are the effects of gate dimensions and overflows using this technique?

AUTHORS' REPLY: We have not yet had an opportunity to use the technique described to make a systematic study of the relationship between these factors and the porosity in the casting.

**Inclusions in Aluminum Alloy Castings,** by J. M. FOX *(paper printed beginning on page 737 of this volume).*

W. S. BLAZEL:[11] Please comment on the formations of oxides that show up on a radiograph as a "more dense inclusion." This inclusion is not from sludge, for it has occurred in top metal. Its color is black. Could it be a complex oxide?

W. E. SICHA:[12] Aluminum oxide actually has a higher density than aluminum. However, the aluminum oxide that may be present in aluminum alloy castings usually is associated with gas, and this spongy material is lower in density radiographically than the surrounding metal. The inclusion that you described is a type that I have not encountered. It seems unlikely that it is a complex oxide, and the best method to identify it would be by x-ray diffraction.

J. O. EDWARDS:[10] With reference to the crystals of spinel, is there any easy method of rapidly identifying spinel crystals from sand inclusions? It appears essential that there should be differentiation, since the remedial measures required would obviously be different in each case.

W. E. SICHA: The more rounded shape, lighter color and proximity to the casting surface of individual sand grains usually will distinguish them from spinel inclusions. Although requiring more time, x-ray diffraction would afford positive identification.

D. L. COLWELL:[5] Would you give us your opinion of charging cold ingot with a die casting holding furnace at 1175 - 1200 F? In an oxide identification by x-ray diffraction there are usually a larger number of different oxides. Do these differences have any significance, or are they all harmful?

W. E. SICHA: In my experience, charging cold ingot into a holding furnace has not contributed to the occurrence of inclusions in castings, but it might have this effect in die casting operations, and it will interfere with proper temperature control. Identification of the specific oxide inclusion can be helpful in selecting the proper corrective measure, and exclusion of all oxides is desirable.

**Iron — The Problematic Factor in the Quality of Aluminum Alloy Die Castings,** by W. BONSACK *(paper printed beginning on page 712 of this volume).*

J. O. EDWARDS:[10] With regard to the beneficial effect of increasing the temperature to reduce the iron hard spot problem, in the case of iron melting equipment or gooseneck machines, would not this increase of casting temperature compound the felony by increasing the iron content of the alloy? Or is the recommended increase in pouring temperature not sufficient to substantially increase the rate of iron solution?

W. BONSACK: It is undoubtedly true that the higher pouring temperatures advocated in this discussion would increase the iron content of aluminum alloys processed in the gooseneck machines. However, I intended to confine my remarks only to the high pressure cold chamber diecasting process. In it iron pots are seldom used, and the ladles or ladling devices do not add to the iron content of the alloy even at much higher temperatures. The demand for high quality die castings has forced most producers into use of the cold chamber process, and this discussion is directed to them.

W. S. BLAZEL:[11] Has the sludge in the bottom of an aluminum melt been known to have a grain refining effect when stirred and dispersed throughout the melt?

W. BONSACK: This appears to be a misapprehension, because the sludge particles are too large to form nuclei for aluminum grain refinement. However, any stirring action which will disperse oxides and other nonmetallics or even small sludge particles, may cause a "pseudo" grain refinement. This type, however, will not contribute to the properties of the alloy and will certainly detract from the quality of the casting, and particularly machineability.

## Ductile Iron Division

**Heavy Ductile Iron Castings Composition Effect on Graphite Structure,** by I. KARSAY and R. D. SCHELLENG *(paper printed beginning on page 672 of this volume).*

H. HENDERSON:[13] How much cerium was added, and in what form?

AUTHORS' REPLY: The cerium addition to melt 26 was made as 0.032 per cent mischmetal added with the ferrosilicon inocula-

tion. This is a rather large addition, as pointed out in the paper. The structure obtained is an indication that an increase in the amount of cerium used to offset the effect of interfering elements in heavy castings is not advisable.

MEMBER: Could the reason why many midwest foundries get good results from additives containing cerium be that the castings are between 2 to 3 in. thick?

AUTHORS' REPLY: In this paper we were concerned with castings equivalent to a 10 in. sand cast cube. In the lighter 3 in. castings mentioned, cerium would be beneficial if interfering elements were present.

D. MATTER:[14] With lower magnesium content would you have a high cerium content and as much chunk graphite?

AUTHORS' REPLY: The magnesium content of the heat to which mischmetal was added was high. However, high magnesium contents normally produce sharp branching flakes of the type shown in Fig. 6 of the paper. The abnormal graphite formed in the cerium-bearing heat was far different, and its presence is not attributed to the high magnesium content.

**Nickel Alloyed Austenitic Ductile Iron Graphite Structures,** by I. KARSAY and R. D. SCHELLENG *(paper printed beginning on page 725 of this volume).*

MEMBER: What is the effect of holding time in the ladle on chunk graphite?

AUTHORS' REPLY: There is no effect except possibly in Mg less than 0.06 per cent it would be necessary to decrease the saturation ratio per formula.

H. WILDER:[15] What about physical properties of metal containing chunk graphite?

AUTHORS' REPLY: There is mainly extreme loss of ductility. Also there is approximately a 20 per cent loss in tensile strength.

**Ductile Iron Graphite Form Classification,** by C. K. DONOHO *(paper printed beginning on page 297 of this volume).*

A. H. RAUCH[16] *(written discussion):* There is certainly a need for a universal terminology covering the different forms of graphite that are found in ductile iron and an agreement on the causes of these various forms, as well as their effects on properties. We believe that Type V of Fig. 3 should be included in any classification system. We usually find this type associated with the graphite-rich layer resulting from the flotation of carbon.

The inference can be drawn from the last paragraph on page 3 of your paper that the higher than normal ductility for the sample shown in Fig. 6 is due to the small nodule diameter. This may be true, but we do not believe that one example justifies such an implication, if it is intended.

It has been our practice to determine the number of nodules per sq mm rather than rate the nodules by size. However, there is an approximate relationship between size and nodule count. We understand that each nodule represents a nucleation center or eutectic cell during solidification. We have found lower nodule counts, and thus larger graphite nodules, in heavier section castings than in light section castings. This would be expected on the basis that slower cooling rates usually result in fewer centers of nucleation. There are other factors, such as degree of superheating and degree of undercooling, which could affect nucleation. We wonder if something might not be said about this relationship of nodule size to nucleation, if you agree that this relationship does exist.

There is an error in Table 1 under "As-Cast Tensile Properties." Both "T.S. x 10-3" and "Y.S. x 10-3" should be "T.S. x $10^3$" and "Y.S. x $10^3$."

We take exception to the statement on page 5 that normal as-cast ductile iron with all Type I graphite will have the average properties shown. This may be true with the particular composition range employed in your foundry. We do not believe that composition can be ignored or minimized in any discussion which deals with properties. We think it would be desirable to show the composition for each of the test bars listed in Table 1, as well as the composition target range for your iron.

We wonder if some readers might not take exception to the statement "that 80-60-03 requirements can be met with up to 40 per cent vermicular, or Type III, graphite." This is true, of course, when the composition is such that the properties of the iron would have had the "average" properties that you show. Unfortunately, however, specifications are based upon minimum values, not averages, and the discerning reader might well inquire what properties would be obtained if the properties without vermicular graphite were near the minimum of the specification. We believe the point — that vermicular graphite is not as damaging as one would think — could be made without reference to the specification.

We have not encountered the graphite form designated as "L" in German classification or Type II in your paper, at least not quite as irregular in shape as shown on the charts. We quite often find imperfectly shaped nodules and, in fact, the perfect spheroids, as shown as *K* or *I*, are observed less frequently than the slightly imperfectly shaped nodules. We accept these im-

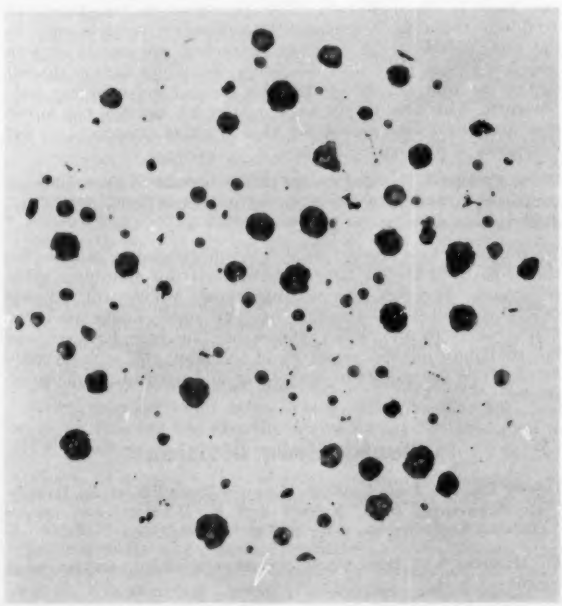


Fig. 1 — Nearly perfect spheroid, 0.072 per cent Mg.

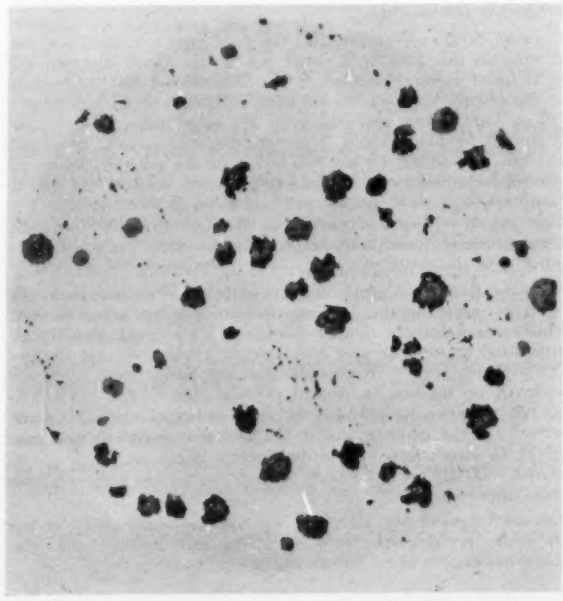


Fig. 2 — Imperfectly formed nodules considered typical, 0.055 per cent Mg.

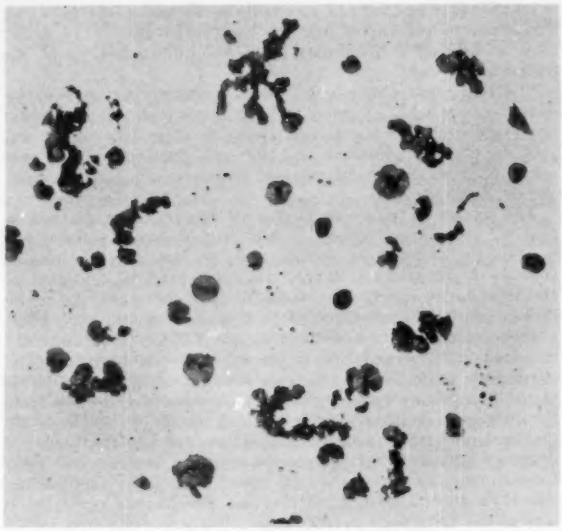


Fig. 3 — Type M graphite, 0.031 per cent Mg.

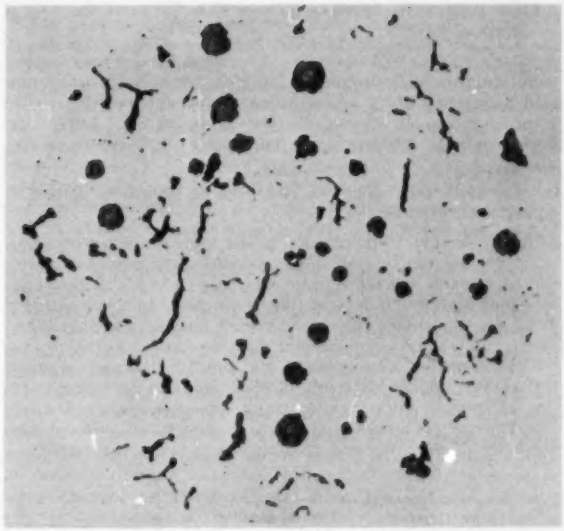


Fig. 4 — Type P or IV graphite, 0.029 per cent Mg.

perfectly shaped nodules as normal and, as far as we know, these have no adverse effect on properties, at least we have not been able to relate the degree of spheroidicity to properties. We were wondering whether or not *L* and "II" are not somewhat exaggerated. We are attaching examples of perfect spheroids, *K* or *I*, shown in Fig. 1, and an example of imperfect spheroids in Fig. 2. We think it would be well to inform users and producers that the nodules will not always be perfect spheroids, as shown in Types *K* or *I*, but the imperfect spheroids do occur and do not adversely affect properties. We do not know what causes the nodules to be perfect sometimes and imperfect at other times.

We also noted that you omitted Type *M* in your classification. We have observed Type *M* and have designated it as "crab" graphite in the past. However, more recently, we have been designating this as a form of vermicular graphite, and have been combining this with Type P or III. Our Fig. 3 shows what we think corresponds to Type M, and Fig. 4 shows Type P or III.

We wonder, too, if some proposal should not be made on names or terminology for these abnormal graphite types, e.g., Type P or III has been called vermicular, wormy, stubby and quasi-flake. Perhaps a name should be suggested for this as well as the other abnormal types. We notice you do use the terms vermicular and spikey, and were wondering whether your usage constitutes a tacit recommendation for adoption of these terms.

We noticed, too, another error on page 4, last paragraph, second sentence — the word "qualities" should read quantities.

AUTHOR'S REPLY: Thank you for your comments on our discussion of graphite classification for ductile irons.

In general, your comments are quite pertinent, and we do not find much to argue about. However, we must disagree with you regarding our captions for Table 1, where we use T.S. x $10^{-3}$ and Y.S. x $10^{-3}$. Of course, we both know that the first figure 81.6 means 81,600 psi. This could be expressed in the Table as 81.6 x $10^{3}$ psi. However, in the caption T.S. x $10^{-3}$ expresses exactly what this figure, 81.6, is; that is, it is tensile strength in psi multiplied by $10^{-3}$. We have been through this many times, and I believe a little reflection on your part will show that we are mathematically correct.

With regard to the assumed average properties of 90-65-12 for as-cast ductile iron, we checked this rather carefully for our as-cast ductile iron for fittings, and these are reasonably average keel block properties when the graphite is predominantly spherulitic. Of course, we get some lower values when the graphite form is not optimum, but we also get some better values with up to 22 per cent elongation, although of course, at somewhat lower strength levels.

We agree that composition cannot be ignored in dealing with properties. However, we did not wish to complicate our discussion of structure with factors that are generally quite well understood. After all, we are not attempting to give a comprehensive text book on ductile iron in this paper.

Glad you caught our error on page 4 which should be quantities rather than qualities.

Again, many thanks for your constructive comments.

C. F. WALTON[17] *(written discussion):* I have just read over your excellent report on the classification of graphite forms and several thoughts came to my mind; I believe the metric system should be used as a basis since it will have international appeal. Anyone using a microscope certainly has to be familiar with the metric system, and most microscope stage traversing devices and micrometers are metric. I believe any type of graphite likely to occur even occasionally in commercial production should be included so that it can be properly classified.

I believe that the types of graphite should be identified by letters other than those *A* through *E*, and numbered for size so as to be consistent with the present flake system. The letter-number significance is already well established, and a single combined specification for all graphite types would seem logical. In view of these comments, should not the malleable types also be included in the same system? If there is a different shape that occurs in the graphitic steels, it might be well to include this too, for we do not know where the technology of our industry will take us in the ensuing years.

A. BOYLES[18] *(written discussion):* Your paper on the classification of graphite in ductile iron is good. I have a copy of Walton's comments, and agree with him on the matter of using the metric system and on the use of letters rather than numerals for the graphite types. The particular letters used are not important so long as they are phonetically distinct. The "M" and "N" in the German chart, for example, could easily be confused over the telephone.

As to your Fig. 3, we find Types I, II, III and IV right along but have never encountered Type V. It has been shown, however, in various publications, and should certainly be included in the chart. On thin-walled castings made in metal molds we sometimes get nodules finer than size 6 in your chart. The following scale is suggested:

Stating the diameter in millimeters gives a clearer mental picture than stating the area in sq mm as in the German chart.

| Size | Dia. mm |
|---|---|
| 1 | 32 |
| 2 | 16 |
| 3 | 8 |
| 4 | 4 |
| 5 | 2 |
| 6 | 1 |
| 7 | ½ |

Aside from these minor criticisms your paper strikes me as an excellent job.

AUTHOR'S REPLY: We appreciate the comments on our admittedly tentative discussion of a rather involved subject. We are glad to see that there is no serious disagreement with the five graphite forms we selected for our proposed chart. Both Mr. Boyles and Mr. Walton agree that our Type V should be included.

D. MATTER:[14] How does A.S.T.M. feel in regards to letters or numbers for classifications?

AUTHOR'S REPLY: With regards to the use of Roman numerals vs. letters for the type designations, this is an editorial matter and presumably will have to be arbitrated in A.S.T.M. We have no strong feelings either way. We are inclined to agree with the discussers who prefer the metric system for size, and that diameter is preferable to and simpler than the area designation. Mr. Boyles' suggested classifications by mm of diameter is quite logical. We wonder, however, if much use will be made of his size 7, which is finer than anything we see, normally. We feel that extremes of graphite size can be designated as finer than size 6, or, at the other end of the scale, as larger than size 1.

**Post Inoculation of Ductile Iron,** by W. C. JEFFERY and H. D. BRADSHAW *(paper printed beginning on page 747 of this volume).*

K. D. MILLIS:[19] This is not pertinent to this paper, but the several slides make it appropriate to refer to the paper by C. K. Donoho on the classification of graphite form. All considerations covered in relation to varying forms included only tensile strength, yield strength and elongation. Mr. Jeffery's slides point out that a far more sensitive property is impact. The curves show what metallographic variations effect elongation on a 2 to one ratio, while impact is effected on a 4 to one ratio. We should keep this in mind when presenting any graphite classification.

**Magnesium Analysis as Reliability Criterion of Ductile Iron Quality,** by AFS DUCTILE IRON DIVISION RESEARCH COMMITTEE (12-K) *(paper printed beginning on page 402 of this volume)*

F. SEFING:[20] What Mg level is to be aimed at for the proposed government sample?

AUTHORS' REPLY: Approximately 0.040 per cent Mg.

F. SEFING: If standards had been available, the industry as a whole might be more able to attain check analysis.

K. D. MILLIS:[19] It is also important to determine what part of the Mg present (in any chemical form) is to be effective as graphite shape control.

MEMBER: Microscopic examination of the metal is still a more positive measure of proper ductile iron rather than Mg content.

## Fundamental Papers Committee

**Vacuum Cast High Strength Steel Sulfur and Carbon Effect on Tensile Properties,** by J. B. DABNEY, M. C. FLEMINGS and H. F. TAYLOR *(paper printed beginning on page 778 of this volume).*

G. HAILEY:[30] Is the relationship to carbon and sulfur the same in air melt heats as it is in vacuum melting?

AUTHOR'S REPLY: Yes.

## Gray Iron Division

**Chilling Behavior of a Roll-Type Cast Iron,** by K. E. PINNOW and R. W. LINDSAY *(paper printed beginning on page 23 of this volume).*

E. C. MATHIS[21] *(written discussion):* The influence of various super heating and pouring temperatures on chill, as demonstrated in the experimental project covered by this paper, conforms to generally accepted beliefs. The conclusion that pig iron brings about a deeper chill than a straight scrap melt is, however, contrary to long established experience.

First of all, let us point out that nonspecification pig iron was used in the pig iron charge. The sulfur content of 0.058 is considerably in excess of standard. Furthermore, the iron had a phosphorus content of 0.62 per cent; whereas, iron normally used in the chill roll industry has a phosphorus content of less than 0.10 per cent.

The important point to be made, however, is that the two tests, i.e., one with all scrap and the other with pig iron, were not made under identical conditions. Graphite inoculation was used in the scrap mixture and was not used in the pig iron mixture. It is an incontrovertible fact that inoculation affects the depth of the chill.

The second inconsistency is pointed out by the wide fluctuation in phosphorus content. There should have been no loss or gain in phosphorus during melting; yet, the phosphorus content of the all-scrap charge was 0.446 per cent and that of the final melts varied from 0.402 to 0.456 per cent. Assuming 20 per cent pig iron was used in the pig iron mix, the phosphorus as charged would have been 0.479 per cent; whereas, the final melts varied from 0.416 to 0.478 per cent. This inconsistency could only have been brought about by oxidation of the phosphorus and its removal by a basic slag derived from the magnesia crucible. Thus, the melting conditions were not identical and the conclusions, therefore, are not tenable.

AUTHORS' REPLY: The pig iron used in the investigation came from production about a decade ago. There is no question but that this iron has high sulfur and phosphorus contents. It should not have been described as a low-sulfur pig iron in the section on "Materials and Experimental Procedure."

As we had mentioned, our results on the use of pig iron in the charge were contrary to the results of previous investigations by others, which related to the amount of graphite in the charge. Mr. Mathis also points out that our results from the use of pig iron are contrary to long-established experience. His point concerning the possibility of inoculation by the addition of graphite to the roll-iron charges, and not to the charges of roll-iron plus pig is quite pertinent. We recognized this to be controversial at the time the investigation was planned. It can only be restated that we added graphite to any heat as soon as possible. In no case in which graphite was added to a heat were test bars poured any sooner than 28 min after the addition; and in the case of the heats superheated to 2900 F there was a time interval of 50 min after the graphite addition before the first test bars were poured. Still, it can not be denied that there might have been an inoculating effect, and based upon previous investigation[1] the results are not inconsistent with the behavior of graphite inoculation.

It is agreed that there are variations in the phosphorus contents which are somewhat unexpected. There are a few corrections which should be made in Table 2: for chill test numbers 2A, 2B the phosphorus content should be reported as 0.436 and the sulfur content as 0.120; for chill test numbers 6C, 6D the phosphorus content should be reported as 0.446. Nevertheless, as Mr. Mathis has pointed out, there is still a spread in P content in the all-scrap heats from 0.402 to 0.456. We can not offer an explanation for this behavior, and we are seeking to have P and S checks made on our samples for analysis. We do maintain that our melting procedure in terms of charging and additions was essentially the same from melt to melt; of course, temperature was varied deliberately and the increase of superheating temperature of necessity prolonged the melts. It might be stated that a second set of heats was made and reported upon in the original investigation.[2] The chilling behaviors of these heats were quite similar to the heats reported in this paper. In the unreported heats, P varied from 0.434 to 0.450 and S varied from 0.094 to 0.102. The objections raised by Mr. Mathis would not apply to these unreported heats.

In closing, we wish to caution against making any changes in present practices involving pig-and-scrap charges based on the results in this paper. It is strongly emphasized that our results on charges of roll-iron plus pig are experimental, and would bear checking in the light of experience and other investigations. Selected pig iron and selected scrap are recognized ingredients of the charge for controlled melting operations, and no change in the charge mix, based upon experiments, should be made without adequate confirmation of the experimental results.

### REFERENCES

1. "Some Factors Influencing the Graphitizing Behavior of Cast Iron," S. C. Massari and R. W. Lindsay, AFA TRANSACTIONS, *vol. 49*, Figs. 11 & 12, p. 977 (1941).
2. "Some Factors Affecting the Chill Behavior of a Roll-Type Cast Iron," M. S. Thesis, K. E. Pinnow, The Pennsylvania State University, University Park (June 1955).

**New Process for Foundry Iron,** by T. E. BAN, B. W. WORTHINGTON and C. D. THOMPSON *(paper printed beginning on page 163 of this volume).*

H. W. LOWNIE, JR. and A. J. STONE[72] *(written discussion):* The authors have presented a useful and informative description of one of the ways that foundries might produce their own cast iron directly from ore. In our paper on this subject given to the AFS Annual Meeting in 1959,* we described a number of such processes, including the D-LM Process, in as much detail as was available at that time from public sources. It was interesting to us, therefore, to compare the authors' results and estimates with those we reported in 1959. The comparison is about as the table.

We have a question on one minor point. No statement is made in Table 1 about the manganese content of the iron ore used in the pilot plant. The pig iron contains 0.72 per cent of manganese. Was there sufficient manganese in the ore to do this? We raise the question because in our own work along similar lines we found that the cost of adding manganese to get this level in the iron approached about $1 per ton of pig iron. The authors seem to have made no allowance in their Table 2.

COMPARISON OF RESULTS

| | Per Net Ton of Product Pig Iron | |
|---|---|---|
| | 1959 Lownie-Stone | 1961 Ban-Worthington-Thompson |
| Iron ore, lb | 3600 (58% Fe) | 3830 (47.6% Fe) |
| Coal, lb | 2000 | 1803 |
| Electric energy, kwh | 1000 | 1040 |
| Electrodes, lb | 13 | 10 |
| Reduction on sintering machine, % | 60 | 50 to 60 |
| Raw materials cost, dollars | 30.90 | 23.92 to 27.55 |
| Energy cost, dollars | 7.00 to 7.50 | 3.80 to 8.80 |
| Operating cost,[1] dollars | | |
| 100 tons/day | 23.10 | 23.00 |
| 400 tons/day | 8.90 | 10.50 |
| Total cost, dollars | | |
| 100 tons/day | 61.50 | 50.72 to 59.35 |
| 400 tons/day | 46.80 | 38.27 to 46.85 |

1. Including labor, overhead, amortization, insurance, etc. Operating 350 days/year.

*"Foundries Can Produce Their Own Cast Iron Directly From Ore," H. W. Lownie, Jr., and A. J. Stone, AFS TRANSACTIONS, *vol. 67*, p. 1 (1959).

Finally, we believe that the authors have been too restrained in expressing the potential value of direct-iron processes to large iron foundries.

AUTHORS' REPLIES: We appreciate the comments by Messrs. Lownie and Stone. The presentation on requirements and economics of production shows a reasonable agreement despite the differences in raw materials. Our presentation concerned the use of relatively inferior grades of raw materials and it is interesting to observe the illustration of comparable iron production costs in the D-LM system.

Ordinarily in making grades of foundry iron we adjust our flowsheet and operating conditions so that approximately 3-5 per cent of the sulfur units are recovered in the metal, about 60-80 per cent of the phosphorus and 30-50 per cent of the manganese. In the case illustrated in our paper, the iron ore inherently contained about one per cent manganese, and for this reason we did not include an addition. With the advent of ductile iron, the presence of manganese is of some concern and perhaps Mr. Worthington would want to amplify this point.

We favorable accept the remark of displaying restraint in illustrating the potential value of our process for large iron foundries. For a number of years we recognized the value of the process for these industries, we have, however, preferred to thoroughly appraise the value by relatively lengthy studies before we expressed our confidence. Again thank you for your comments. — T. E. BAN

The interesting comments by Messrs. Lownie and Stone are appreciated. These men did much of the early work on the process and they should know both the problems encountered and the distinct economic advantages possible.

Mr. Lownie's question about the manganese in the pig iron has been largely answered by Mr. Ban, but a few comments on this point may be timely. The ore used in the early work carried only about 0.1 per cent manganese, and it was necessary to add ferromanganese in order to produce a product with 0.6 per cent manganese. This could have been accomplished equally well by using a small amount of manganiferous ore in the burden.

At the time of the earlier work on consideration was entertained in reference to ductile iron and normal pig iron for cast iron pipe was set as the target grade of iron to be produced. In the production of ductile iron the presence of manganese in excess of 0.2 per cent lengthens the annealing time and is therefore undesirable.

Finally, I concur with Mr. Lownie in the restraint of the authors in expressing the potential value of the Dwight-Lloyd McWane Process to large iron foundries. In the paper no data was presented other than that obtained "on the average" from an entire pilot plant campaign. No projections were made other than between the efficiencies of a pilot size and a commercial size electric furnace. It is obvious that conditions near start-up and shut-down of a pilot campaign are neither normal nor average. If the first and last seven taps (out of 85 to 90) are eliminated from the power evaluation the commercial power consumption is reduced to the range of 850 to 906 kwh per net ton of pig. — B. W. WORTHINGTON

R. C. SHNAY:[22] What constitutes the St. Lawrence River area?

AUTHORS' REPLY: The St. Lawrence River area was chosen as a source of favorable raw materials costs as cited by publications. The $9.59 cost for ore market price in Canada was given in the April 1960 Skillings' Mining Review. The 4 mil power is a stated rate at St. Jean in the Montreal vicinity. Coal is considered to be imported from the Lake Erie area.

R. C. SHNAY: Have you had any experience with ores high in manganese and titanium?

AUTHORS' REPLY: We have had pilot experience with titaniferous ores and observed that these can cause slag problems. The Bremanger plant in Norway has smelted titaniferous varieties of iron ore for a great number of years by using smothered arc electric furnace. The people of this plant are very skilled operators and have proven in practice the capability of using titaniferous ores as a furnace burden. Slag and temperature controls are very important for smelting this ore in avoiding excessive formation of titanium carbide.

R. C. SHNAY: Have you as yet any commercial experience?

AUTHORS' REPLY: At the present time we have not made a public announcement of a D-LM turnkey plant award. We are in the process, however, of evaluating the process for several clients who do not desire to have their plans made public at this time.

R. C. SHNAY: What is the difference between your process and Strategic-Udy?

AUTHORS' REPLY: Essentially the D-LM Process uses a Dwight-Lloyd machine for pre-reduction of pelletized composite charges. This charge is then fed into conventional smothered arc electric furnaces. The Strat-Udy Process uses a rotary kiln that produces a charge that is not agglomerated. This charge is fed into an open bath type electric furnace.

R. MELMOTH:[23] Can you tell me the reason for the sharp break in the cost and assets curve at about 70 per cent?

AUTHORS' REPLY: That is a good observation on your part. In designating economic information as a per cent of design capacity we considered a point of 70-75 per cent capacity to be critical for considerations of lay off periods. One part of the curve illustrates economics at reduced operations using the full compliment of staff. Ordinarily this would exist for short periods of time rather than resorting to reduced personnel. The other part of the curve represents reduced operations wherein full complement was modified to meet reduced operation schedules. Actually this should probably be plotted more properly as two distinct curves without a junction line.

M. H. HORTON:[24] Can this unit be shut off daily or weekly without losing control or getting other adverse effects? The reason

for this question is that many foundries operate on a one or two shift basis producing 400 to 1000 tons of melt per day. The third shift is used for maintenance. If the unit could not be shut down one would be forced to produce pig iron or operate continuously. Our studies in this type operation indicate it is impractical to operate continuously and we do not wish to produce pig iron for the open market.

AUTHORS' REPLY: In comparison with usual blast furnace-coke oven systems of making iron the D-LM Process can be considered interruptable without serious loss of control. The large smothered arc furnaces in the Norwegian iron and steel industries have operated for a number of years at 50 per cent of full loads. There are economic factors of operating, for instance, on a five day week basis, and perhaps fixed costs is one of the more important items.

In our pilot plant operations we have banked our furnace for periods as long as 8 hr without serious loss of process control. This capability is probably related to the high rates of smelting within the system by using the hot charge. This allows the crucible to be re-established within minimum periods of time.

The capability of running intermittently should be analyzed for your particular situation where such items as tolerance of metal grade, function of fixed charges and apparatus maintenance are given thorough analysis.

I believe you remarked that costs of slag disposal could appear excessive for your case. Our economics illustrate sufficient charges for this item.

**Engine Castings Development by Experimental Stress Analysis,** by L. A. GROTTO *(paper printed beginning on page 636 of this volume).*

D. KRAUSE:[25] What is the relationship of hardness and fatigue life as influenced by cooling rate?

L. A. GROTTO: They are related, and research work is presently being carried on to determine these factors.

M. H. HORTON:[24] What effect does mold cooling rate have on residual stresses?

L. A. GROTTO: Indications are a definite relationship exists here. Work is continuing in this field.

**Calcium Carbide Use in Acid Cupolas,** by R. SCHULZE *(paper printed beginning on page 307 of this volume).*

R. CLARK:[26] I was pleased to present this paper for the author Mr. Schulze.

Our Company has been interested in this German development. Although calcium carbide has been used in cupolas for some years the eutectic composition having a lower melting point had not been tried in this country. A few months ago we made up a quantity of this material to determine whether it would work as well in the United States, where better coke and larger cupolas are the rule, as it apparently does in Germany.

Experience to date indicates that under normal conditions an addition of one per cent will increase the spout temperature about 50 F, with 2 per cent giving a 100 F increase assuming the coke ratios remain unchanged. These increases will vary a good deal depending on conditions in the individual foundry. Where rather cold metal is being produced the carbide has more influence on iron temperature than in a foundry already melting well above 2800 F. We have seen spout temperatures of 2900 F or slightly above in normal cold blast cupolas, but believe this is about the limit to be expected.

In cases where the amount of coke between charges is reduced, iron temperatures have been satisfactory and melting capacity increased. In spite of the fact the calcium carbide is compensated for by a reduction in limestone sulfur contents have normally been reduced slightly.

We believe that the future of this material will depend on the application and the individual foundryman's experience with it. One probable use would be in melting ductile iron in acid cupolas, depending on the carbide to produce metal high enough in temperature to permit efficient desulfurization in the ladle. We still have a good deal to learn about this material and its possible advantages.

D. D. PROFANT:[27] Why not use 80 per cent carbide and let it react with the stone?

R. CLARK: This appears to be a question of proximity. While the patents mention a mechanical mixture of 80 per cent carbide with lime, Dr. Schulze tells us this scheme has not worked well. As mentioned in the paper, 80 per cent carbide melts at a temperature higher than exists in the operating cupola and is not readily soluable in acid slags. The pieces are rather coarse and would not normally be in close enough contact with stone for them to flux each other.

H. S. SCHAUFUS:[28] I have two questions — (1) What about cost? (2) Has it been used in hot blast?

R. CLARK: (1) Calcium carbide of any grade is more expensive fuel than coke, and its use must be justified by metallurgical or other advantages. As far as lining attack is concerned, the Germans report that attack on gannister lining is somewhat deeper directly above the tuyeres when carbide is used, but does not extend as high in the cupola. In our experience we have not encountered any unusual refractory damage. It will take more experience with longer campaigns to definitely establish this fact.

(2) The eutectic carbide works satisfactorily in both hot blast and unlined cupolas. In one case a 1½ per cent addition in a hot blast unlined water cooled cupola permitted a reduction in coke usage by about 25 per cent without reduction in iron temperature and with a proportional increase in melting rate.

J. OLIVIERI:[29] What about economics?

R. CLARK: As just stated, carbide is a more expensive source of heat than coke. The economics depends on its use to obtain higher temperature metal together with increased carbon absorption which in many cases might permit a more economical metal charge. It can also be used to increase melting rate, and may benefit the properties and castability of the iron.

**Inoculation Influence on Structure of Gray Irons at Various Carbon Equivalents,** by H. D. MERCHANT, L. I. TORIELLO and J. F. WALLACE *(paper printed beginning on page 241 of this volume).*

I. C. H. HUGHES[31] *(written discussion):* This is a useful contribution to the literature, and we are pleased to see that the results are substantially in agreement with many of those which have been reported by J. V. Dawson in his paper to the March 1961 issue of the *B.C.I.R.A. Journal.* It would appear that the two investigations have been proceeding along somewhat similar lines. Although Mr. Dawson did not find that the degree of nucleation in uninoculated irons increased with carbon equivalent up to 4 per cent and then fell off again, we believe that this observation is not uncharacteristic when the charges used contain carbonaceous material.

We are surprised to see that the effectiveness of inoculation, after decreasing with carbon equivalent to the eutectic, increases again in hyper-eutectic irons. This, if substantiated, could be an important result.

It is our experience that the variation in eutectic cell number from edge to the center of a bar increases in inoculated irons. This is almost certainly associated with the undercooling at the surface, causing a closer approach to the maximum possible number of eutectic cells for a given degree of nucleation.

AUTHORS' REPLY: The authors appreciate the kind words of Mr. Hughes concerning this paper. The similar results of Mr. J. V. Dawson's in the March 1961 issue of the *B.C.I.R.A. Journal* were also noted.

This latter paper, however, was concerned primarily with the effect of various additions on the inoculating effect of the irons rather than influence of carbon equivalence.

**Solidification Progress in Inoculated and Uninoculated Gray Irons,** by H. D. MERCHANT and J. F. WALLACE *(paper printed beginning on page 249 of this volume).*

I. C. H. HUGHES[31] *(written discussion):* The object of this work was to determine, if possible, by a study of the solidification sequence of castings, the reason why inoculated castings expand in the mold to a greater extent than uninoculated castings. Interesting experiments were carried out using a similar technique of quenching and examining to that employed by investigators at the British Cast Iron Research Assoc. We have ourselves carried out similar experiments and in many respects we have obtained similar results. There are, however, some details of the experimental investigation upon which we should like to comment.

In spite of the fact that inoculated and uninoculated irons had the same carbon equivalent content, their silicon contents were different, with the result that they contained appreciably different eutectic graphite contents. This has caused the uninoculated iron of higher eutectic graphite content to have the longer eutectic arrest. This iron also had a lower liquidus temperature, which must be ascribed to its different composition, since our experience is that inoculation does not affect the liquidus temperature.

We believe that inferences on the formation of an air gap, based upon interface couple readings, are likely to be erroneous. Such couples are subject to error, and readings from them would need to be repeated before confidence could be placed upon them. In any case, we are not aware of any experimental evidence for the formation of an air gap in sand-cast gray irons. From the results of this experiment, such an air gap would presumably have been present early in the uninoculated iron. We think this is unlikely to be the case.

Except in rigid molds, both inoculated and uninoculated castings are generally larger than the mold cavity in which they were made. It is difficult to see how an explanation of the difference between inoculated and uninoculated irons, based upon air gap formation, can be in agreement with these general observations.

The differences in the extent of a solid skin and of a mushy zone between inoculated and uninoculated irons are complex, and largely disappear towards the end of solidification. Our experimental measurements of casting expansion suggest that the difference between inoculated and uninoculated irons is clearly shown during expansion measurements during the later periods of solidification. This is difficult to reconcile with the kind of observations which have been made by the authors and which we ourselves have been able to make. This holds true even though our detailed observations of the progress of eutectic solidification are not wholly in agreement with those of the authors.

It is surprising that no reference has been made to the difference in eutectic cell numbers in the two irons, which are clearly shown in Fig. 2. Our view is that this difference in cell number is of importance and that more attention should be paid to the mechanics of expansion occurring when the mushy zone contains high or low eutectic cell numbers. This we believe to be the key to the difference in expansion, and we suggest that in spite of the authors' last conclusion the observations which have been made do not really assist in explaining the greater mold wall movement that occurs in inoculated irons poured in green sand molds.

AUTHORS' REPLY: The authors' reply to the discussion of Mr. Hughes will be general in nature since his comments are of this type. We are interested in the mechanism that results in the greater mold wall expansion found in inoculated irons compared to uninoculated irons poured in green sand molds. By identifying these mechanisms a start can be made for approaching the control of this rather complex phenomenon. The authors have recognized the role played by the smaller cell size of inoculated irons in increasing mold wall movement. This subject is discussed in detail in the last AFS progress report on the effect of inoculation on the risering of gray iron referred to as reference one in this present paper. The question is whether this finer cell size is the only significant factor causing the greater expansion, as inferred by Mr. Hughes, or whether other mechanisms are active. We believe that other mechanisms are active and have attempted to show what these are in this paper.

Eutectic solidification in gray iron takes place by nucleation and growth of spheroidal shaped eutectic cells, which continue to grow until they impinge upon each other. Graphite is precipitated during eutectic solidification, and due to its low density the solidified eutectic has larger volume than that of the liquid from which it forms. This situation creates internal pressure acting on the casting walls. A part of this pressure is relieved through the riser in form of exudation, while the rest of it is relieved by movement of the casting wall. When the casting wall is quite rigid, as in dry sand mold, the exudation in the riser becomes clearly visible.

The magnitude of the internal pressure becomes particularly large during the last stages of eutectic solidification near the casting wall. At this stage, the eutectic cells have impinged upon each other and have some liquid entrapped between them. For obvious reasons, graphitization of the entrapped liquid tends to move the cells apart and the pressure, at least a part of it, is relieved by the movement of the casting walls. This process of liquid entrapment and pressure exertion occurs over a wide region of the casting and continuously along the advancing solid-liquid interface. Soon after the solid skin is formed on the surface of the casting, internal pressure may be relieved by plastic movement of the skin; however, as the solid-liquid interface gets further from the casting surface so that the skin becomes quite thick, pressure relief by movement of the skin becomes increasingly difficult.

Inoculation has been shown to increase the number of eutectic cells; in addition, the manner of freezing becomes mushier and there is some delay in formation of completely solidified skin. From simple geometrical considerations, it can be observed that finer eutectic cells have larger surface area and, hence, they will exert larger total force for a given pressure due to entrapment and precipitation of graphite; also the mushier manner of freezing will widen the region in which simultaneous liquid entrapment occurs. These two factors combine to increase the magnitude of internal pressure that may be exerted on the casting wall. In addition, delay in formation of the solid skin will decrease the resistance of the casting wall to withstand internal pressure. Thus, it may be concluded that increase in internal pressure due to graphite precipitation as well as decrease in resistance of the casting wall to withstand this internal pressure contribute towards increase in mold wall movement in inoculated casting.

The authors believe that a real difference exists in the extent of the solid skin and mushy zone between inoculated and uninoculated irons, contrary to Mr. Hughes' discussion, although more data on this subject would certainly be desirable.

The problem of the air gap is a touchy one and we have interpreted our data to make the point that one may exist. However, the experimentation is difficult and we do not have much data — only three sets. Mr. Hughes makes several good points that cast doubt upon the existence of this air gap. However, more complete experimentation will provide the ultimate answer.

## Heat Transfer Committee

**Factors Affecting Casting Soundness in Certain Alloys With Long and Short Freezing Range,** by J. T. BERRY and T. WATMOUGH *(paper printed beginning on page 11 of this volume).*

R. C. SHNAY:[22] Would you expect the same advantages in long freezing range alloys by applying the pressure as centrifugal force as in a centrifugal casting?

AUTHORS' REPLY: It would be expected that a similar advantage would be had in centrifugal casting in long freezing range alloys. However, it should be noted that the inner layers of such castings might have to be "sacrificed" as a source of liquid metal which would be centrifuged toward the outer layers. Further, careful attention might have to be given to the segregational problems which are known to occur in centrifugal casting.

G. VINGAS:[32] You have shown that castings could be produced that are smaller than the pattern dimension. Would this be due 100 per cent to expansion of sands?

AUTHORS' REPLY: Whether the casting surfaces, such as those shown in Fig. 3 for flat plates, are convex or concave depends upon both mold and metal contraction or expansion characteristics. We would refer the discussor to the table included in the second paragraph of our paper regarding these effects. In addition reference 5 discusses this topic most lucidly.

## Industrial Engineering and Cost Committee

**Establishing Work Standards by Sampling,** introduction to motion picture by E. C. KEACHIE *(paper printed on page 664 of this volume).*

A. T. KIRK:[34] How do you achieve "correct" rating of performance?

REPLY: Rating during work sampling is a questionable practice. Performance "rating" of specific motions is done in stopwatch timestudy and requires careful definition of the job. Overall "performance rating" of employees is done periodically by super-

visors. Both are subjective techniques, and "correct" rating in either case is a specialty in itself, as it would be if done in work sampling.

P. W. FORD:[35] What is the practical maximum number of workers that an observer can include in a project?

REPLY: This depends so much on the nature of the work that it would be misleading to estimate in general. However, the relative effectiveness of the technique probably would be fairly constant so that the number of workers to be observed would not make or break the project.

H. F. DIETRICH:[36] If a work incentive is based on this system, what happens to the wage system when work increases or decreases, as in normal business?

REPLY: Even if incentives were based on work sampling, and work loads fluctuated, the condition would not be peculiar to a single method of work measurement, or even incentives. Proper controls over work assignment and output are still necessary. As in other normal businesses, work increases or decreases may be met by well tried means, and external conditions. For example, the adjustment of sales efforts, delivery dates and flexible inventory controls may be used in addition to the regulation of the labor force. Actual hirings or layoffs may be tempered by working overtime or, conversely, by not replacing those who leave.

## Light Metals Division

**Light Alloy Grain Size Control and Supercooling Measurements,** by V. B. KURFMAN *(paper printed beginning on page 234 of this volume).*

J. F. WALLACE[37] *(written discussion):* The author is to be complimented on a most interesting and fundamental presentation on grain refinement. The correlation obtained between degree of supercooling and grain diameter is excellent and consistent, showing the accuracy of the theory. The effect of solute content on the degree of supercooling of the grain coarsened magnesium alloys also indicates the effectiveness of constitutional supercooling in restricting grain growth. It is rather surprising that zinc does not show more supercooling than aluminum in this latter case because of the greater liquidus-solidus spread. The effect of copper, silicon and magnesium solutes on supercooling in aluminum alloys is rather puzzling. Would the author offer an explanation.

The suggested method of determining the magnesium content of Al-Mg alloys by thermocouple measurements of the liquidus and solidus arrests as recommended by the author has been found to be effective for foundry use.

V. B. KURFMAN: It was gratifying to learn that Professor Wallace has found the methods described in the paper useful. In reply to the comment on the lack of a quantitative correlation between observed supercooling and the degree of spread between liquidus and solidus lines on the equilibrium phase diagram: the results may be considered to show that the supercooling which occurs in practice is not simply proportional to that theoretically available. One reason may be that there is appreciable convection mixing and thermal agitation, which will surely prevent the development of as much supercooling as might at first be expected.

W. BLAZEL:[11] With a 9 per cent aluminum magnesium alloy is there a noticeable difference in supercooling between one melt which was superheated (grain refining procedure) and one melt which was not superheated?

V. B. KURFMAN: As shown in Fig. 2, there is a certain grain size below which no supercooling is observed. It does not matter what procedure is used to achieve the grain refinement. Where small grain size differences exist, it is usually possible to detect these differences by a more careful examination of the cooling curve for the configuration shown in Fig. 1b.

B. F. DARIN:[38] In samples taken after various times of chlorination is not it possible that the more you chlorinate the more nuclei you remove by flotation?

V. B. KURFMAN: Yes. The reason for chlorination was to grain coarsen the melt to increase the maximum observed supercooling.

**Hypereutectic Aluminum-Silicon Casting Alloys Phosphorus Refinement,** by F. L. ARNOLD and J. S. PRESTLEY *(paper printed beginning on page 129 of this volume).*

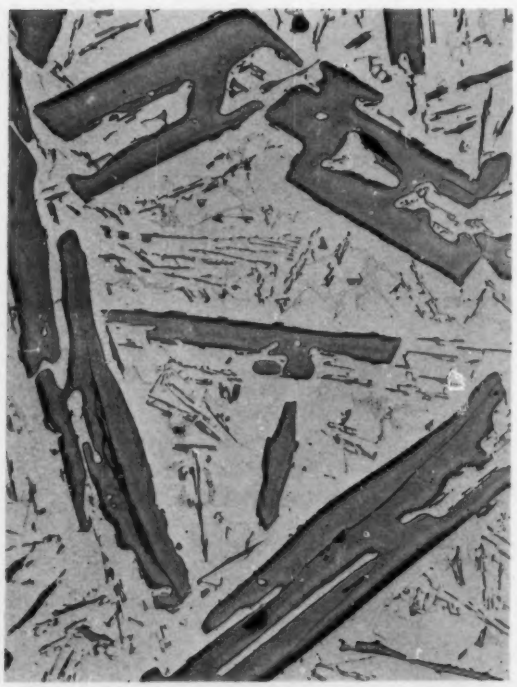


Fig. 1 — Unmodified slow cooled ¾-in. plate. Note massive primary and coarse eutectic phases. Unetched. 100 ×.

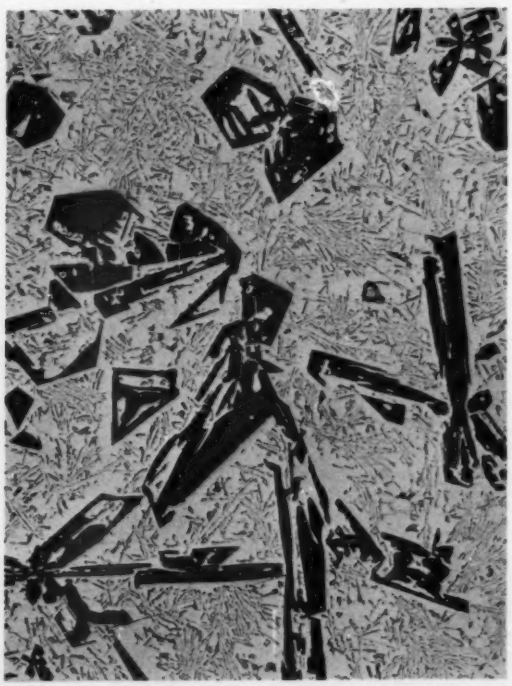


Fig. 2 — Unmodified ¾-in. plate cast against copper chill. Note refinement. Unetched. 100 ×.

| Alloy | Holding Time After Modification, Hr | Min | U.T.S., psi |
|---|---|---|---|
| 20.53 Si, 0.86 Mg, 0.58 Mn | 1 | 45 | 44,200 |
| 0.85 Cu, 0.015 P | 2 | 13 | 42,200 |
| | 3 | 23 | 40,000 |
| | 4 | 8 | 44,600 |
| 20.80 Si, 0.86 Mg, 0.56 Mn | 1 | 40 | 46,000 |
| 2.49 Cu, 0.018 P | 3 | 34 | 48,500 |

Y. P. TELANG[39] *(written discussion):* The authors should be congratulated for their interesting work; particularly noteworthy is the introduction of the electron microprobe x-ray technique to study the phase which nucleates the primary silicon in hypereutectic Al-Si, phosphorus modified melts. Lund has centrifuged phosphorus modified melts and has positively identified the nucleating phase as AlP using x-ray diffraction.

The observation that "properly inoculated metal could be held at pouring temperature for a least 1½ hr with no great change in refinement" is significant. We concur in this finding. We have examined the ultimate tensile strength levels and the

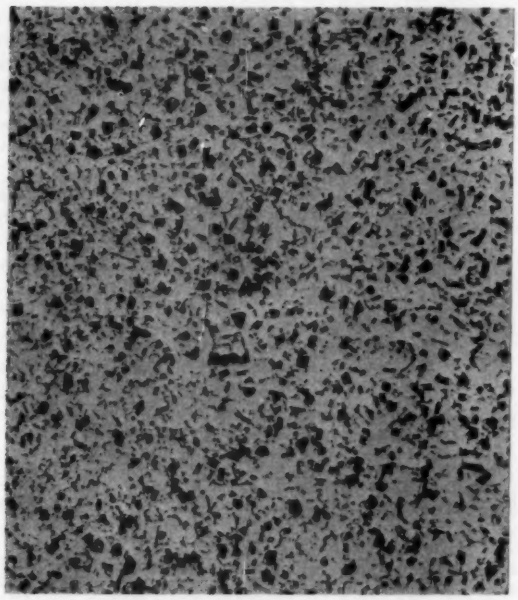


Fig. 3 — Modified with 0.03 per cent P as Cu-P alloy, chill cast. Note refinement of primary and eutectic phases. 100 ×.

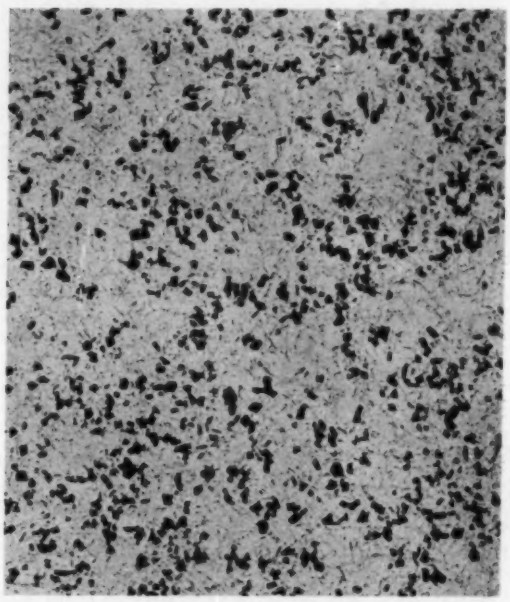


Fig. 5 — Same heat. ½-in. permanent mold casting section. 100 ×.

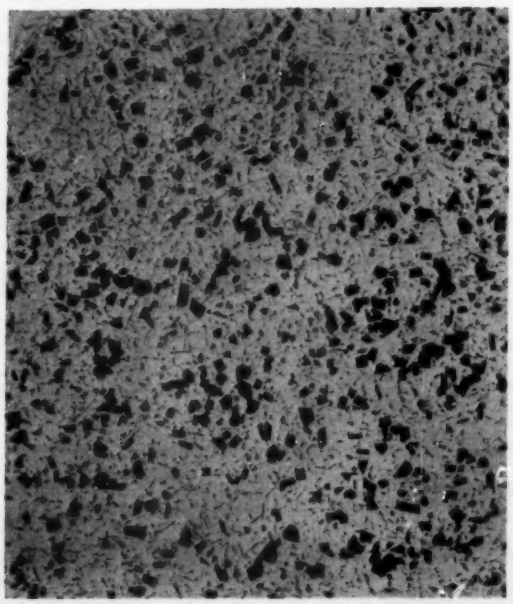


Fig. 4 — Same heat. One in. round permanent mold casting. 100 ×.

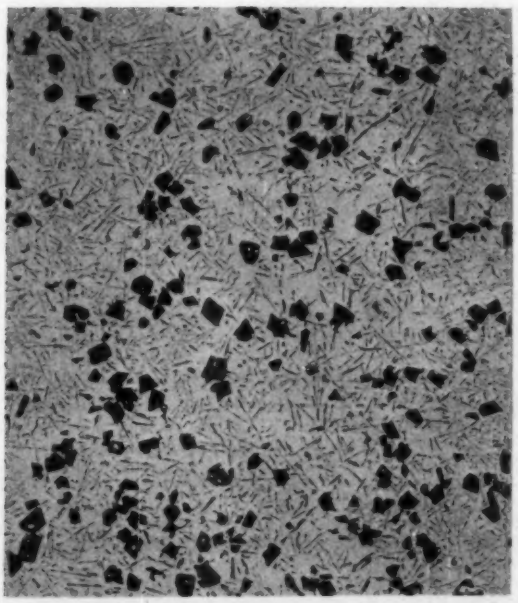


Fig. 6 — Modified with 0.03 per cent P, chill cast. No Mg added. Note primary Si and eutectic size. 100 ×.

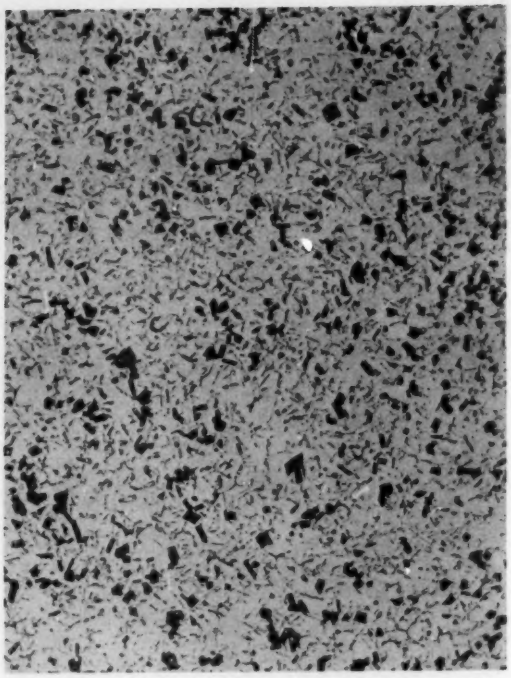


Fig. 7 — Same heat, poured after 0.80 per cent Mg addition. Note refinement. 100 ×.

microstructures of 0.03 per cent phosphorus modified Al-21 per cent Si alloy after holding the melt in the range of 1380-1420 F for up to 4½ hr and have not noticed significant coarsening of the microstructure or deterioration in the strength. The tensile test specimens were machined from one in. permanent mold, $T_7$ heat treated one in. round bar castings.

The microstructures in Figs. 3a, 3c and 4a, 4c clearly indicate the substantial refinement of the eutectic silicon constituent by the phosphorus addition, although the refinement of the eutectic by phosphorus is disputed by Onisch Modl. The order of magnitude of the refinement is not as high as it is in case of sodium modified hypereutectic Al-Si alloys due to the difference in the two refinement mechanisms. But the ultimate tensile strength of the finer eutectic in sodium modified melts cannot be effectively realized due to the dendritic and/or spheroidized pattern of primary silicon in the sodium modified alloys. The investigation by the authors does not reflect the real order of magnitude of the refinement of the primary silicon particles with phosphorus, because the unmodified basis specimens show marked inherent refinement due to the pronounced undercooling of the castings. The microstructures in Figs. 1 and 2 reveal the massive primary silicon size in the unmodified slow cooled and chill cast specimens, and those in Figs. 3 to 5 show the refinement in the 0.03 per cent phosphorus modified chill cast and permanent mold specimens, the latter approaching the ultimate tensile strength levels of 50,000 psi.

We are interested in the authors' technique of melt preparation which permitted them to chlorinate the melt at 1400 F, for 5 min and hold it for 5 to 10 min prior to pouring at 1400 F. We find that while chlorinating a 15 lb melt around 1450 F the temperature initially rises and then drops rapidly without permitting adequate holding time prior to pouring at 1380 F to 1400 F range.

The authors have indicated that "elements like copper, magnesium, manganese, iron and titanium . . . have no effect on primary silicon so far as amenability to phosphorus refinement is concerned." Our experience in this area has been that metallic magnesium additions to the phosphorus modified melt prior to pouring result in a substantial added refinement of the primary and the eutectic silicon particles over a wide solidification range as shown in Figs. 6 and 7.

AUTHORS' REPLY: We are, of course, pleased to hear of the close agreement of Mr. Telang's work and our own concerning the effect of holding time on refinement. Especially so, since his conclusions were based on tensile and ours on metallographic data.

Our statement concerning the effect of alloying elements on refinement was to the effect that the phosphorus refinement process could be effectively used on a hypereutectic aluminum-silicon alloy regardless of other elements in the alloy's composition. Mr. Telang's finding that magnesium additions further improve refinement is very interesting and, to our knowledge, has not been reported in the literature.

With reference to sodium modification, Mr. Telang states that the high ultimate tensile strength which might be expected of modified metal is not realized due to the dendritic and/or spheroidized pattern of the primary silicon. We have also observed this phenomenon but feel that the lower strength is more attributable to the fact that sodium just does not appreciably reduce the size of the primary crystals.

We feel that our investigation does reflect the real order of magnitude of the refinement for a 16 per cent silicon alloy. The contrast between the size of the unrefined and the refined silicon particles does not appear as great in our microstructures as in Mr. Telang's primarily because our alloy contains only about half the primary silicon as the one he has worked with and therefore produces much smaller primary crystals in the unrefined condition.

J. F. WALLACE:[37] What is the reason for the relationship between the volume of nuclei and the volume of silicon particles? This relationship is not customary in other inoculated metals.

AUTHORS' REPLY: We did not intend to imply a relationship between the size of the nucleus and the size of the primary silicon crystals. We realize that the degree of refinement would be expected to depend on the number rather than the size of the nuclei. However, we did obtain some indication that the size of the nucleus became larger the greater the degree of refinement. We believe that if this relationship exists, it is probably because higher phosphorus contents produce larger, as well as more numerous, aluminum phosphide crystals.

**Initial Bubble Test for Determination of Hydrogen Content in Aluminum Alloys,** by D. J. NEIL and A. C. BURR *(paper printed beginning on page 272 of this volume).*

S. LIPSON:[40] Would the authors be in a position to specify the range of gas contents which are tolerable for producing premium quality castings, and if so would the instrument be discriminatory in repeating heats which contain excessive gas and those in the acceptable range?

AUTHORS' REPLY: We are not in a position to specify the gas content necessary to produce a precision quality casting in a particular alloy; however, work is in progress which we hope will provide such information in a general way. We are of the opinion that a given heat will produce results which are quite dependent upon mold design, and consequently it may be necessary to specify the gas level necessary to produce a high quality casting for a particular alloy cast into a particular mold.

It is impossible to provide a categorical answer to the second part of the question until both the optimum gas levels and the necessary control limits are known. In general, we are of the opinion that optimum gas levels for production of quality cast-to-shape products are not likely to be as low as has been the goal for ingot to be used for fabrication but the section thickness and, consequently the freezing rate will be a determining factor. As stated in the paper, experience indicates that the results of careful measurements with this tester are likely to be within $\pm$ 0.05 ml $H_2$/100 g of the correct value. The application of this instrument to hydrogen content measurement of casting alloy heats, in cases of higher optimum gas levels, would lead to lower relative errors. The absolute error may be rather high, but we feel that changes in gas content can certainly be followed and multiple observations will improve the accuracy.

I would like to take this opportunity to point out that the abscissa units in Figs. 2, 3, 4 and 5 of the paper should be $\sqrt{\text{pressure (mm Hg)}}^{1/2}$.

**Meeting the Challenge of High Reliability Castings,** by W. BABINGTON, E. E. MILLER and W. R. CLOTHEROW *(paper printed beginning on page 695 of this volume).*

K. E. NELSON:[41] The team effort of the authors' companies is an excellent example of what can be done by coordination of

efforts between engineering design and manufacturing to assist the foundry in its efforts to expeditiously provide quality castings. Their realistic approach to quality and strength requirements is also exemplary. We wish to congratulate them on this presentation.

Most foundrymen, I am sure, are pleased that these companies have recognized that the procurement of high quality castings requires dealing with foundries having the capabilities to procure quality parts from producers of quality. It is a healthy sign to see that the authors' companies are advocating sound procurement practices.

This discussor's company is pleased to have passed the scrutiny of the authors' companies, and are producing castings meeting their requirements. We intend to improve our position by developing production, inspection and delivery practices which are better than those now followed. It is only by continued improvements in these fields that a foundry can keep up with the increased requirements of this space age.

J. G. MEZOFF:[8] The authors and their companies are to be commended for their enlightened program in the design, procurement and use of metal castings. One hopes that others will follow this leadership.

**High Strength Aluminum Casting Alloy M517-T61, A preliminary evaluation,** by W. A. BAILEY, E. N. BOSSING and F. H. ROEBUCK *(paper printed beginning on page 660 of this volume).*

G. D. CHANDLEY:[42] The figure concerned with the ring type casting does not show the casting dimensions. What were the section thicknesses?

AUTHORS' REPLY: The test bars were taken from sections ¼ to 2½-in. thick.

**Aluminum Alloy 356 + Be Permanent Mold Casting, Centrifugal force effect on mechanical properties,** by A. J. ILER *(paper printed beginning on page 394 of this volume).*

F. L. ARNOLD[6] *(written discussion):* This comment concerns the section of Mr. Iler's paper under the heading Metallographic Analysis. Here "grain size" and "grain refinement" are discussed as a function of the location of the sample, whether from the chill or riser end of the casting. What is shown in the microstructures of Fig. 6, however, appears to this writer to be differences in dendrite spacing and dendritic cell size rather than grain size. It is true that both grain size and dendritic spacing are influenced in the same direction by solidification rate, but no grain boundaries appear in the microstructures (which appear unetched); and the areas bounded by the eutectic silicon constituents ordinarily outline the solidification of primary aluminum dendrites and are more properly called dendritic cells.

If this is true, it would be interesting to etch these samples in order to bring out the grains and compare the effects of the parameters in the centrifugal casting process with regards to their influence on both dendrite spacing and grain size.

AUTHOR'S REPLY: Mr. Arnold's comments are certainly correct. The series of photomicrographs only indicate dendrite arm spacing, and a correct interpretation would be measurements made of these spacings to indicate rate of solidification. The reference to grain size was used for clarity for nontechnical readers, intending that observation of differences in dendrite spacing would be an indication of a corresponding grain size which was a fact and it was hoped would be assumed.

## Malleable Division

**Shooter, Ceroxide Cermet or Slag** by C. A. SANDERS, H. J. HEINE and R. F. MARANDE *(paper printed beginning on page 340 of this volume).*

T. E. BARLOW:[43] Twenty-five years ago Carl Joseph wrote a paper in which it was pointed out that pin holes resulted from skulls remaining in ladles following their use on the previous day. Iron oxide is effective against these defects.

J. B. CAINE:[44] Do not confuse discussions and designations by using the term "Cermet." Cermets are something different and have a definite composition and application. What we are talking about in the subject paper is a "visible nonmetallic inclusion."

Adding iron oxide will not prevent this inclusion and adding aluminum will not prevent it. One source is the droplets of metal in the side of a ladle. When it is tilted, they oxidize, react with the refractory and manufacture a low melting point eutectic.

R. W. HEINE:[45] The addition of 0.15 per cent aluminum to white iron for malleable production gives a dry crusty slag which persists for 5 min. Oxidation is more spontaneous as temperature decreases. Patches of slaggy material develop below the surface. With more slag they get larger and sometimes penetrate along dendrites. They may result from using more steel in the charge, thus supplying an oxidizing material if a reducing material (more fuel) is not added to offset it. They may have a black iron oxide, a material penetrating the ladle lining. If you add 0.15 per cent aluminum, you will have only a few. The ladle deteriorates in part in relation to the type of iron melted in the first place.

D. WILLIAMS:[46] Radine of Moscow, pointed out that soluble gases will not come out unless you have a solid-liquid interface on which you will find your blow, or whatever. Prevent the formation of a solid-liquid interface and you will be all right. If you have a slag inclusion, you will get a blow along with it. Get rid of the finely divided slag or the solid-liquid interface.

F. JACOBS:[47] Reline ladles every day, with a silica sand lining bonded with Bentonite.

Use a brick in the ladle to coagulate the inclusion. If you cannot trap in the gate, additions of 0.02 per cent aluminum will deoxidize.

MR. SMALL:[48] At high temperatures the metal holds inclusions which cannot get out at low temperatures. When you pour at a high temperature into the mold and have a reaction in the mold you have two types, that from the outside and that developed in the mold.

G. DISYLVESTRO:[49] Use 5 per cent iron oxide and the inclusion comes to the surface and flattens out. Build up the sand so it will not react with the iron.

MR. LEWIS:[50] How about $CO_2$ ladle lining?

AUTHORS' REPLY: In our experience with them they were not good although some report they are all right. The lining was eaten out and more trouble created. It is good to have a new molding sand bonded lining every day.

## Pattern Division

**Design and Engineering of Hot Core Boxes,** by E. E. HARKESS *(paper printed beginning on page 193 of this volume).*

M. RONNE:[51] On shell cores, when the core is contoured, will it cure in time before ejection pins are raised?

E. E. HARKESS: As stated in the paper, if the equipment is designed properly for making a contour or complicated core, there should be no problem in curing.

MEMBER: Do you use baffle plates for dry sand mixes?

E. E. HARKESS: As I stated, baffle plates are used over the top of the blow plate holes in the reservoir so sand can drain out.

F. PLAIN[52] When your core is ejected high enough for removing, how is it handled?

E. E. HARKESS: The pick-off fingers raise the cores clear of the box and ejection pins and then retract for pick-off by the operator.

J. HENRY:[53] Do you have any problem with overcuring?

E. E. HARKESS: Not unless something goes wrong.

W. MASON:[54] How do you control temperatures for heavy masses of cores? We have some trouble with our gas heat control.

E. E. HARKESS: There must be the proper temperature control system for the type of fuel used.

J. FOLSOM:[55] What kind of resins should be considered?

E. E. HARKESS: Most widely used thermosetting sand binders are the phenolic resins.

MEMBER: What is the maximum temperature?

E. E. HARKESS: Based on a 450 F operating temperature, a 0.003 in. expansion factor is incorporated.

**Aluminum Pattern Castings,** by W. E. SICHA *(paper printed beginning on page 479 of this volume).*

R. L. OLSON:[56] Has vacuum been tried in making pressure cast castings?

W. E. SICHA: Use of a mild vacuum would be helpful.

W. J. DOELKER:[57] Has any development been made regarding thickness of matchplates?

MEMBER: Thickness depends on pressure applied and span between supporting flask.

MEMBER: Most plates in the east are ½-in. thick.

M. RONNE:[51] If you squeeze both plates at the same time even a thin plate will not deflect.

## Sand Division

**Granular Movement During Squeezing,** by D. C. WILLIAMS *(paper printed beginning on page 8 of this volume).*

H. W. DIETERT:[58] You indicated that tensile strength was difficult to determine. Tensile and shear strengths tests may be termed as the accurate ones whereas the compressive strength may be described as the more inaccurate one, due to the fact that the sand specimen is too short – which causes the shear planes to intersect. Would it not be better to increase the length of the compressive test specimen? Further, do you feel that such a change would cause the strength values to fall within values shown by the Mohr circle of stress?

D. C. WILLIAMS: At this time I have only the opinion that if the length over diameter ratio of our compression test specimen was changed to two, we would find that test data relating various properties would agree much closer with predicted values from Mohr's circle.

However, with the l/d ratio of two, we would be faced with compaction problems, but it may be that a slight procedural change in compacting the 1⅛-in. x 2 in. specimen would meet the requirements.

J. B. CAINE:[44] Can the relationship between green compression tensile and shear strengths, as expressed by Mohr's circle, be used as a measure of "flowability"?

D. C. WILLIAMS: I would think so, although I have not yet been able to secure a copy of the only paper I know of that attempts to develop such a relationship; an abstract of this paper "The Friction Angle in Green Molding Sands and Its Relationship to Technological Properties and Hydration of Clays," by J. Navarro Alsaser, *Foundry Trade Journal* (December 17, 1959).

There is also the idea that the flow value used in soil mechanics, namely $N\phi = \tan^2 (45° + \phi/2)$, could be used to evaluate flowability. Since $\phi$ is the only unknown, which can readily be determined by constructing the abbreviated Mohr's circles as shown in the paper, $N\phi$ can be readily calculated. If someone is interested in this he might try to correlate $N\phi$ with experience or laboratory tests.

E. C. ZIRZOW[59] *(written discussion):* In the conclusion the following statement is made, "For the squeeze board to descend the two movements (horizontal and vertical) occur simultaneously." In my opinion this statement is not correct. The greatest movement is vertical. Vertical movement always occurs first, especially in a loose packed sand such as you would find in a mold. You do not get any horizontal movement until you approach maximum compaction. At that point, and not until that point is reached do you get simultaneous vertical and horizontal movement. Hence your greatest amount of horizontal movement occurs next to the squeeze board where the greatest compaction exists and not adjacent to the pattern.

Furthermore, the amount of horizontal movement is small compared to the amount of vertical movement. The amount will vary dependent upon the composition of the molding sand.

As proof of the above statements, I can cite a series of tests which were made. It was suspected at one time that we were not using enough bars in our flasks. The flasks were of rugged steel construction. We rigged up a caliper-like instrument using a dial to determine if we were getting any movement of sides and ends of the flasks during the squeeze. The squeeze board was mounted, and provided a minimum amount of clearance between the vertical sides and ends of the flasks. This was a jolt, squeeze operation, and hence the sand was compacted fairly well before any squeeze was applied. Measurements were taken at the end of the jolting period. After the jolting period, the sides and ends of the flasks showed no evidence of bulging. The squeeze was then applied. The squeeze pressure was about 100 psi. The sides and ends of the flasks were measured with the squeeze pressure on. This indicated that the sides and ends had bulged between 0.040 and 0.050 in. However, the vertical movement of the squeeze board was between 2.0 and 2.5 in. In other words, a horizontal movement of between 0.040 and 0.050 in. was obtained compared to a vertical movement of 2.0 to 2.5 in. Bear in mind that this was after jolting. If we had not jolted the vertical movement would have been much greater. As further proof that the vertical movement is greater we reduced the squeeze pressure and obtained a vertical movement of 1.25 to 1.75 in., and no measurable horizontal movement.

The suggestion that the dimensions of the squeeze board be decreased is also incorrect. Following this suggestion might lead to serious consequences. In all probability it would lead to run-outs, because due to the limited horizontal movement of the sand adjacent to the pattern the sand would not be compacted beyond the squeeze board. The mold would be soft in the areas not covered by the squeeze board. Proof of this is in the common practice of leaving a hole in the squeeze board where minimum compaction is desired or by removing a portion of the sand over the area where minimum compaction is desired.

D. C. WILLIAMS: I appreciate Mr. Zirzow's comments concerning his investigation, and I would like to reply, but first, all references to the literature will be taken from *Theoretical Soil Mechanics* by Karl Terzaghi.

In the discussion a squeeze pressure of about 100 psi was used. It is not clear as to whether this pressure refers to that on the air line or to the psi actually applied by the squeeze board. Since it is often customary to refer to the line pressure when discussing squeezing, this condition will be assumed until details are obtained regarding area of squeeze board and total effective pressure developed. Therefore it will be considered that the squeeze pressure applied by the board will be some fraction of the line pressure.

The horizontal movement of the flask sides, as indicated by Mr. Zirzow, is small. An estimation of the lateral forces applied against the flask walls can be obtained from equation (2) of Terzaghi's article no. 7 on page 22. Rearranging this equation for lateral pressure we find

$$\sigma_{111} = \frac{\sigma_1 - 2c \tan (45° + \phi/2)}{\tan^2 (45° + \phi/2)}$$

where $\sigma_{111}$ = laterial pressure in psi.
$\sigma_1$ = vertical pressure delivered by squeeze board in psi.
$2c \tan (45° + \phi/2)$ = green compressive strength.
C = cohesion.
$\phi$ = angle of internal friction.

Assuming $\phi$ at a level of 50 degrees, then the denominator calculates to a value of about 7.6. It is readily seen that such a denominator value leads to relatively low pressures against the flask sides and the lateral movement reported by Mr. Zirzow are very reasonable, even if he used a delivered squeeze board pressure of 100 psi.

In order for the squeeze board to descend below the top level of the cope the pressure applied must displace the sand located above the surface of sliding *bcde,* as in Fig. No. 4 of the paper. In squeezing a rigid restriction (the wall of the flask) is interposed and the sand moving along the surfaces of sliding above *bcde* will develop compaction along flask walls which are parallel to the longer dimension of the squeeze board.

Regarding run-outs, the deeper the squeeze board descends the lower the level of the surface of sliding along the line *bcde,* and increased compaction closer to the parting line providing no part of the pattern intersects the surface of sliding. The proof suggested by Mr. Zirzow might be reviewed from the point of view of the opening in the squeeze board. Such an opening establishes additional points comparable to point *b* changing the relation between the length and breadth of the squeeze board, which containing an opening must be considered as a combination of squeeze boards. The direction of the surfaces of sliding will be changed depending upon the contour of the opening. Surfaces of sliding along *bcde,* which are opposed in direction, will reduce the descent distance of the squeeze board yielding a softer mold at the parting line. Depending (upon the location of the hole this could be one method of producing compaction against the flask wall of the shorter dimension. This brings out the important point that certain edges at the perimeter of the squeeze board and at opening within the squeeze

board are the loci for the start of granular movement along surfaces of sliding.

E. C. ZIRZOW: The 100 psi, I refer to is the pressure at the machine. The line pressure ran between 120 and 125. I am not prepared to argue the possibility that soil mechanics can be applied to the making of a mold. I am objecting to your conclusion or inferences.

I made two points in my discussion of your paper:

1) I do not agree with you that at the initiation of the squeeze you get simultaneous vertical and horizontal movement. It is my contention that at the start of the squeeze you only get vertical movement. If we take a theoretical example of sand three grains in depth, loosely packed as in the case when the sand is placed in the mold. The first movement of these three grains when the squeeze is applied is vertical or in the direction of the applied force. This vertical movement continues until the grains have made contact with each other. After making contact they will slide off in a horizontal direction so that it would be theoretically possible to reduce the depth from three to one grain of sand. Hence I reiterate that initially you do not get simultaneous horizontal and vertical movement. It is only after the grains have come into contact with each other that you get both vertical and horizontal movements or simultaneous vertical and horizontal movement.

2) My second point was that if you reduce the dimensions of the squeeze board the mold would be soft around the edges especially at the parting lines, and that in all probability you would get run-outs. The mold would be hardest adjacent to the squeeze board, and get softer as the distance from the squeeze board was increased. Actually the softest part of the mold would not be adjacent to the pattern because of the slight ramming effect which is obtained when the sand is placed into the mold but the softest part of the mold would be very close to the pattern parting line. The above can be proved by checking hardness. The packing which would be obtained would be least from the slight horizontal movement and greatest from the vertical movement. The horizontal movement would be the least, adjacent to the pattern and the greatest, adjacent to the squeeze board.

Terzaghi's formula proves this. Hence, unless the distance between the pattern was abnormally large to permit unlimited horizontal movement there would be the possibility of run-outs. Incidently, for economical reasons, the amount of sand between the pattern and outside of the mold is usually held to the absolute minimum. It all gets back to the fact that the greatest compaction is in the direction of the applied load. Where you do not have any load being applied, such as on the section of the mold not covered by the squeeze board, the compaction is the least. Hence, by reducing the dimensions of the squeeze board the compaction is reduced in the areas not covered by the squeeze board or the outside of the mold where the compaction should be the greatest. If it is not good practice to have good hardness of the mold at the parting line, then a lot of time has been wasted peening molds or applying peening strips to the outer diameter of the squeeze board.

**Steel Foundry Sand Clay Bonding, A microscopic study,** by J. B. CAINE, E. H. KING and J. S. SCHUMACHER *(paper printed beginning on page 59 of this volume).*

G. DESLYVESTRO:[49] What has been your experience with a wetting agent in a bentonite water sand system.

AUTHORS' REPLY: Within the limits of the tests, there has been no observable effect of wetting agents on clay distribution. Wetting agents seem to increase the mobility of the water, but not of a solid such as a clay.

**Systematic Approach to Sand Design and Control, Progress report 3 — The mulling effect,** by G. J. VINGAS and A. H. ZRIMSEK *(paper printed beginning on page 101 of this volume).*

V. ROWELL[60] *(written discussion):* Several years ago a subcommittee of the AFS Sand Division undertook the study of mulling techniques. This effort bogged down in disagreement on so many details that the effort was abandoned. Any attempt to establish fundamental principles on mulling techniques and their effects on sand properties and behavior is quite complicated. The authors have done a commendable job on this difficult subject.

Molding sands that are relatively low in moisture content are the most important if casting quality is considered. Most accurate dimensional tolerances and maximum freedom from sand defects on castings are achieved with molding sand that is low in moisture, high in clay and rammed uniformly hard. The authors have pointed out that this type of sand requires the most attention to mulling efficiency.

The green properties depicted in the paper were all obtained on freshly mulled sand. This information is pertinent only when new sand mixtures are used. Sand properties may change rather drastically with aging. Green compressive strength, for example, may be only 5 psi after inefficient mulling, and may rise to 15 psi after tempering for 24 hr protected from moisture loss.

In a typical iron foundry system, sand mulling is far less critical if the sand-clay combination is exposed continuously to water, without completely drying out, than in a steel foundry where facing sand is compounded of dry new or reclaimed sand and dry clay with moisture mulled in and the sand used immediately.

The amount of mulling required to develop optimum properties from the clay and water added is tremendously affected by the moisture content of the sand-clay mixture for several hours prior to mulling. It is hereby suggested that tests made on freshly mulled new, dry sand and clay mixtures be compared with test results on the same mixtures after several hours. If this tempering time is possible under production conditions, the effects of mulling time and efficiency tend to be minimized.

Any interpretation of the results cited in this paper should be modified by consideration of individual foundry conditions and further test made simulating the time-moisture conditions prevailing in the individual foundry.

AUTHORS' REPLY: At this stage of the research project, the authors cannot state factually whether low or high moisture per se is important for the production of good castings. There are foundries today making castings considered accurate dimensionally and free from defects, with sands ranging in moisture content from 1.7 to 8.0 per cent. This contradicts, of course, the theory held by some that low moisture is necessary for production of good castings.

As an example, a sand of 7.45 per cent western bentonite and 3.9 per cent water, mulled for 2 min (Fig. 3), will feel and react identically to a sand of 7.45 per cent western and 3.0 per cent water, mulled for 6 min (Fig. 4). This paper factually explains and concludes that moisture content is not absolute but relative to clay content and, most of all, mulling time.

The authors are in agreement with V. Rowell that the data presented can be applied at present to new sand mixtures only. In the initial stage of this project, the principles of mulling new sand, clay and water are presented. Data on system sands will be gathered later. The same principles which apply to new sands will in all probability be involved in system sands to a different degree. It should be noted, however, that at no time during this experiment did the properties of sand change when it was stored in polyethylene bags for extended periods of time. The properties might be expected to change, however, in reused sands that are hot and undermulled.

**Systematic Approach to Sand Design and Control, Report 2—dry properties of clay bonded sand,** by G. J. VINGAS and A. H. ZRIMSEK *(paper printed beginning on page 33 of this volume).*

D. C. WILLIAMS:[46] The two progress reports presented by A. H. Zrimsek and G. J. Vingas have been stimulating to this writer, because they have caused me to participate in some reflective evaluations. The two reports have raised more questions than answers provided. The actual values obtained and comparisons between clay minerals seem relatively minor as this discussion is prepared. I do not expect the authors to prepare an exhaustive reply to the questions set forth below. These questions are presented mainly to stimulate thinking by anyone who so desires to become interested.

1. Is there any real good reason for making the dry compression strength test?
2. What causes the development of dry compressive strength? When an answer is provided for this question, then possibly there will be sufficient justification to give an affirmative answer to question 1.
3. In progress report 2, indicates the results when western and southern bentonites are used quite similar results are obtained. Then we can ask what contribution a clay-mineral makes to the dry compressive strength? Is dry compressive strength due alone to the clay-mineral? If the development of

dry compressive strength due to a combination of factors some of which are unknown at this time?

4. Is the dry compressive strength value, as is commonly measured, the same as that developed in the test specimen at a temperature level of approximately 220 F? If not, then what happens during cooling from 220 F to room temperature?
5. Why is "air dried" compressive strength lower in value than dried compressive strength? In both tests water is removed, the main difference being in the rate at which the water leaves the test specimen. What takes place because of the heat treatment during the oven drying process?
6. Is the difference in strength values due to the clay minerals or some effect of heat? What understanding is necessary to develop mixtures whose dry compressive strength does not depend upon the type of clay mineral used?

AUTHORS' REPLY: The authors wish to thank Dr. Williams for his discussion. The purpose of this research project is the presentation of facts about sand and elimination of opinions and theories based on speculation that exist about it. If this series of papers stimulates thinking among foundrymen and raises questions in their minds, one of the goals of this research project has been accomplished.

The authors will discuss the questions Dr. Williams has posed in the order he presents them:

1. Dr. Williams questions the dry compression strength test. The authors question not only the validity of that test but all sand tests. The first part of this research project has been designed to determine variables affecting casting quality and evaluate them qualitatively and quantitatively. In other words, the assumption is made that the foundryman knows how to make a good casting. If the molding medium is kept constant, manufacture of consistently good castings can be expected. This research project will, therefore, either reject or substantiate the present testing methods for control purposes, and will eventually determine whether these same test methods can be used as a design tool for sand formulation.

At this stage of the project the only factual conclusion that can be reached is that test results can be used to control a property in relation to five others. Specifically, it can be stated that there is nothing magic about any value of dry compression strength, as long as the number is constant in relation to green compression strength, green shear strength, density and water content. It is the conclusion of the authors, therefore, that if satisfactory castings are produced in a sand of certain quality described by a series of physical tests in a given period, the same casting results will be obtained if the physical properties and formulation are maintained.

The actual importance as already shown is not in the physical properties themselves, but specifically in clay content, water content, mulling intensity and additive content. The testing of physical properties is simply used to measure the above.

2. The reason for the development of dry compression strength is a question beyond the scope of this investigation.

3. Although the dry compression strength results of southern and western bentonite, presented in progress report 2, were shown to be equal, they were substantially different with regard to surface texture. The authors failed to report the observations made during the testing. What contribution, if any, clay makes to dry compression strength cannot be answered by the authors for this also is beyond the scope of this investigation.

4. No difference was noted when samples were tested at room temperature or 220 F.

5. Dr. Williams' question regarding the differences encountered between air set and dry strength results cannot be answered comprehensibly, since the former property was not included in the design of the test.

6. The authors wish to refrain from answering the last of Dr. Williams' questions since the answer would be based purely on speculation.

AUTHORS' CLOSURE: The authors are guilty of omitting two graphs from the original text of their progress report 2. The graphs are those referred to on page 39, column 1, where the text should be corrected to read, "As shown in comparing Figs. 13 and 14, . . ."

The authors wish to apologize for this omission which undoubtedly caused confusion for readers. *(see page 923)*

**The Carbon Dioxide Process,** by G. D. HALEY and J. L. LEACH *(paper printed beginning on page 189 of this volume).*

MEMBER: What period after gassing did you test strength?

AUTHORS' REPLY: Immediately after gassing. Actual tests should be based on actual flow of gas and not on pressure. There is no definite relationship between time and gas pressure.

**Systematic Approach to Sand Design and Control, Progress report 4 — wood flour,** by A. H. ZRIMSEK and G. J. VINGAS *(paper printed beginning on page 141 of this volume).*

V. ROWELL:[60] The term "Wood Flour" really should be broadened to include finely ground cellulosic materials of nonwood origin. Certain pulverized grain fibers may, for example, resemble certain wood flours more than does one particular wood flour resemble another.

The authors state that wood flour type materials are often added to molding sands to increase green strength, flowability and workable moisture range and to decrease dry strengths. They also concede that these reasons are a bit oversimplified.

Many foundrymen employ these materials to gain casting benefits through modification of elevated temperature characteristics of molding sands. The normally expected benefits are reduction or elimination of expansion defects due to sand. Hot strength in the range up to 2000 F is reduced, hot deformation is increased, particularly in the critical range around 1200 to 1500 F, and improved flowability results in more uniform mold density. The improved sand flowability resulting from the proper addition and control of wood flour type materials of proper quality will definitely improve casting finishes on vertical walls and in pockets of the mold.

The findings and conclusions of the authors are sound and informative. However, it should be clearly understood that wood flours and similar cellulosic materials will differ from one another in details of performance just as much as will clays of different types and origins.

AUTHORS' REPLY: The authors wish to thank V. Rowell for his comments. The authors are in agreement that the details in performance of wood flours, cellulosic materials and grain fibers differ. During the experiment a few commercially available wood flours and ground cobs were spot checked for their effect on molding sand. The effects were found to be similar to the data gathered with the wood flour used in this research project. Cellulosic materials and grain fibers should be checked in the same manner as wood flour was in this project to definitely determine their effect on molding sands.

The authors stated that the reason for adding wood flour to sands today is to increase green strength, lower dry strength and improve flowability. They also stated that this reason is a concept that is oversimplified. The conclusions reached and presented in this paper show factually what wood flour does to clay bonded sands.

The views of the authors on hot properties of sands were presented previously,[1] and there is no need to discuss them further. In the near future, casting results will be presented using the various sand mixtures which were used in gathering the data published in the series thus far. A more comprehensive discussion can then be made regarding casting finish, packing in pockets, and flowability.

In conclusion, the only recommendation the authors can make to foundrymen at this time regarding the difference in details of performance of wood flours, cellulosic materials and grain fibers is that foundrymen request this information from their suppliers.

MEMBER: What kinds of sand did you use in your tests?

AUTHORS' REPLY: Portage silica sands.

R. KOTTKE:[61] Does the method of addition of water, wood flour and bentonite effect the tests and graphs shown?

AUTHORS' REPLY: The progress report on the mulling effect covers this question. You add clay and wood flour first, then water.

1. "Is a New Concept Needed for Hot Sand Properties?" MODERN CASTINGS (Nov. 1960).

## Steel Division

**Factors Affect Metal-Mold Reactions,** by G. A. COLLIGAN, L. H. VANVLACK and R. A. FLINN *(paper printed beginning on page 52 of this volume).*

O. J. MYERS:[62] This work represents that excellent approach to

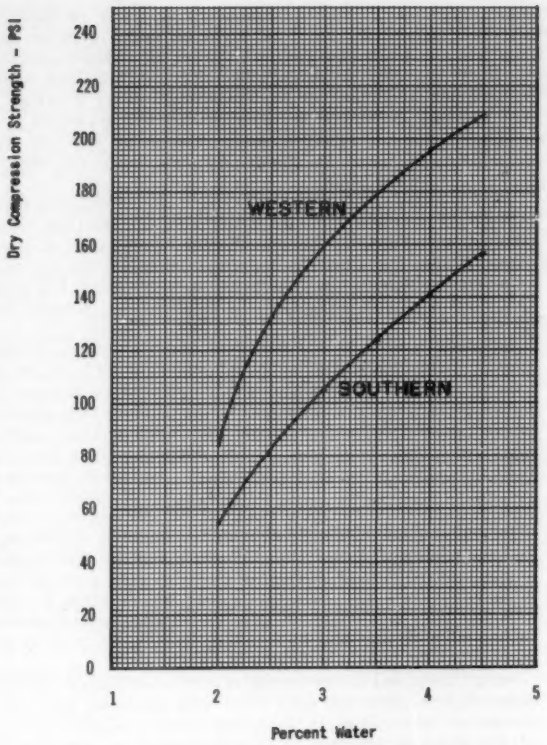
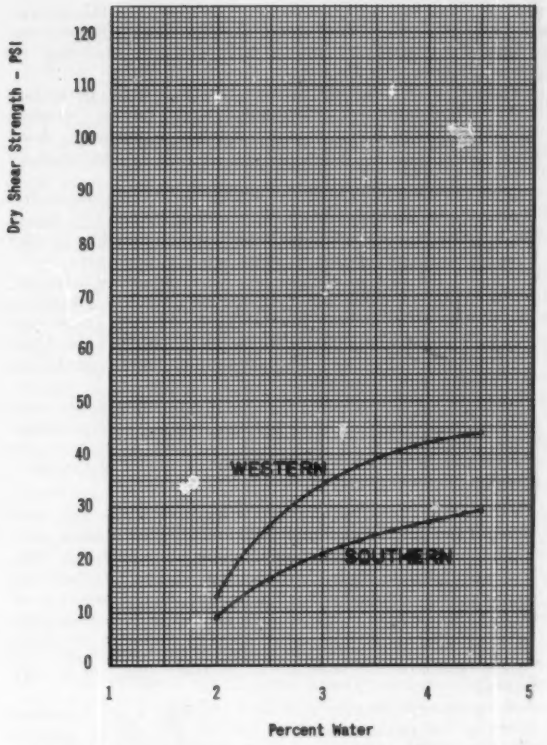



Fig. 13 — Water content effect on dry properties of sands bonded with 4.75 per cent western or 4.75 per cent southern bentonite mulled 6 min in 12 in. muller.




Three ram properties shown compare with results from 18 in. muller (Figs. 1 and 2).

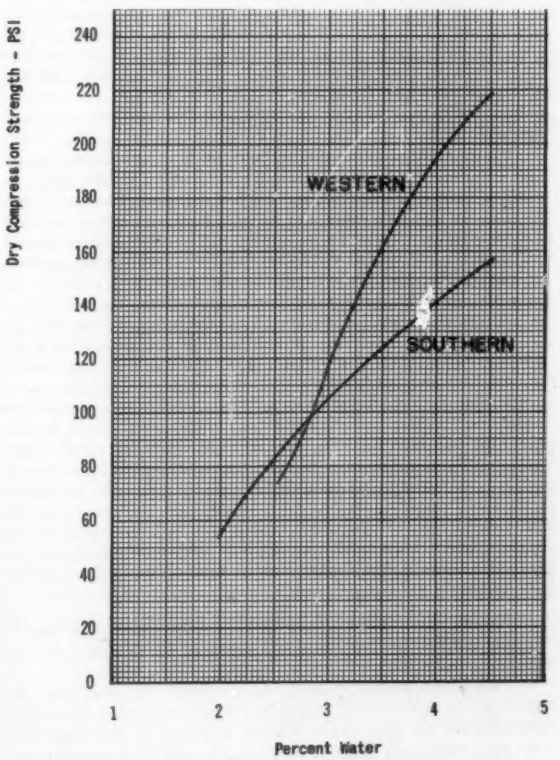
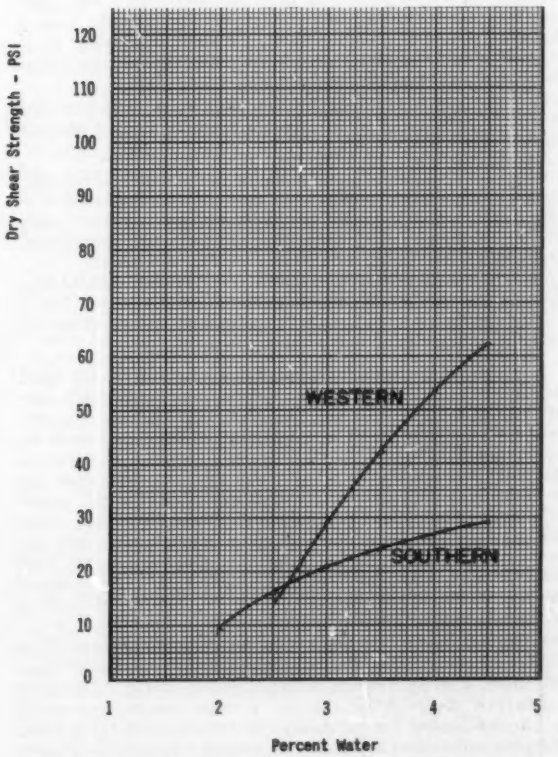



Fig. 14 — Water content effect on dry properties of sands bonded with 7.45 per cent western or 7.45 per cent southern bentonite mulled 6 min in 12 in. muller.




Three ram properties shown compare with results from 18 in. muller (Figs. 1 and 2).

a complex problem which satisfies both the theoretician and practical foundryman. Detailed data present factual evidence for the conclusions reached. Castings are presented verifying the hypothetical data.

It is not clear whether the authors took into account the fact that the green sand molds contained water during founding while the shell molds were dry. A test using a completely dried green sand mold might prove valuable as well as one based on baked core sand.

Another significant experience would be to coat (and dry) the green sand molds with a mold wash bound by liquid phenolic resins rather than the usual oxygen bearing binders such as vegetable oils, proteins, sugars and/or dextrines.

It may be possible to further minimize the mold-metal reaction by using phenolic resins of various ratios of phenol to formaldehyde, ranging from the simple "one step" type containing no hexamethylenetetramine "catalyst" through the whole suite of novolac "two-step" types containing from 5 to 15 per cent hexamethylenetetramine.

I would also like to suggest that some confusion might exist in studying Fig. 9 where the proprietory term "Bakelite" might lead one to believe that resin from the shell mold is in the microstructure, when actually the mounting media of the specimen is meant.

AUTHORS' REPLY: Green sand molds containing water were used for the purpose of producing an atmosphere at the metal-mold interface with the highest possible oxidizing potential. The authors agree that extensive investigations should be performed to evaluate other mold binders and mold refractories. We regret that lack of support for this work prevented its continuation.

H. HENDERSON:[13] The authors have done a good job in pointing out the reactions at the mold-metal interface and how to cope with them.

While I am not familiar with steel castings, I am interested in knowing the methods used to identify the various mineral components found in the sintered areas. For example, in Fig. 10 the cristobalite (C) looks the same to me as the quartz (Q), so my question is: How were these constituents identified?

I would also like to ask if the manganese effect on mold-metal interface reactions is present in gray iron and ductile iron castings. There is indication that it may be, for example, with a gray iron containing 1.50 to 2.00 per cent manganese, it is almost impossible to cast a clean surface especially if the silicon content is lower than the manganese.

It is also noted that ductile iron of 0.30 to 0.40 per cent manganese gives an extremely smooth finish either case in shell molds or green sand molds.

AUTHORS' REPLY: Initially, x-ray diffraction as well as metallographic study of these phases is necessary for positive identification. As experience is gained, identification may be made using only the microscope. Such properties as reflectivity, morphology, hardness and color permit these phases to be identified.

The manganese effect on reactions in gray and ductile iron may be quite significant, and we would intuitively expect this to be true. Unfortunately, we have performed no experimental studies in this range of metal composition.

P. J. NEFF:[63] This work attacking the mechanism of the mold interface reactions that lead to penetration contributes substantially to the knowledge of the subject. In a general way, we agree with the conclusions and many of the comments made by the authors. On the other hand, there are a number of questions regarding the manner in which some of the conclusions are reached. There are repeated references to the effect or the role of the so-called "oxidizing" or "reducing" conditions. These terms have been used in connection with mold reactions by nearly any one who has worked with the subject. In many cases, they have probably been used incorrectly. The authors seem to recognize this, but still leave an element of doubt in use of terms such as "more oxidizing" or "less reducing", etc.

On page 2, there is reference to "oxidizing conditions" in a green sand mold and to the "strongly reducing conditions" in a shell mold. The equilibrium diagram, shown in Fig. 1, seems to suggest that the interface reaction at high temperature would not proceed unless the percentage of $CO_2$ were above a value somewhat more than 3 per cent. Attention is drawn to a statement on page 9 pertinent to the oxidizing conditions prevailing at 3.9 per cent $CO_2$. This is further illustrated in Fig. 14, illustrating mold wall attack. Thus, there is question at what stage does the atmosphere become reducing. A statement on page 10 allows that "under reducing conditions, a sample of the same size exhibited no reaction." It might be useful to have these reducing conditions described. This takes on added significance in view of the comment to the effect that "the oxidizing nature of the green sand mold produces extreme interface reaction . . . The reducing nature of the shell sand mold permitted little interface reaction." In other words, it is stated that the shell mold produces a reducing atmosphere, and yet is not immune to the penetration effect.

This question of oxidation might be further explored, either in light of comments that are made or by examination of the photomicrographs. Attention is called to the presence of higher oxides of iron that occurred in green sand molds. On the other hand, the micrographs of shell molds show no oxides, but rather point out that iron is present as a result of the reducing conditions.

It has been pointed out that even small additions of manganese greatly lower the reducing tendencies of a given gas mixture. This apparently makes it impossible to prevent the formation of a liquid silicate melt in the presence of manganese. While the authors have not claimed the elimination of penetration with a shell mold, it is of some interest to note that even with the highest manganese addition, the amount of reaction at the interface was less than occurred in green sand with the smallest manganese addition. Would it be unreasonable to have expected progressively increasing penetration with higher manganese in the shell mold, since this is the case with green sand?

It would be of some interest to know by what technique the authors obtained the analysis of gas at the interface. It seems that this may be a matter of great importance, yet the method of sampling at the correct location and at the proper time would seem to present a difficult situation.

In the discussion of the casting results, a comparison has been made of both green and shell sands with increasing amounts of manganese in the steel. In discussing the microscopic evaluation of the interface reactions, it is stated there is "extensive evidence" of the formation of a liquid silicate melt. Similarly there is a statement relative to the small amount of silicate occurring in the shell mold. Is quantitative estimation possible on the basis of microscopic examination? The photomicrographs themselves would not permit such a conclusion. We assume that such a quantitative evaluation would not be obtainable from the particular illustrations contained in the report. Was this opinion derived from a more extensive examination of specimens on the microscope, or was it influenced by the amount of reaction products adhering to the castings and evident on a macro-examination?

In at least two places, the authors point out that the possibility of preventing interface reactions by the use of neutral or reducing gases is most impractical. We are inclined to agree that this is entirely true, but it is questionable whether the statement follows from the information that has been presented. For instance, it is stated that "observation of the phases present . . . indicates that normal cast metal practice operations lie above the boundaries between the liquid silicate melt field and the solid and liquid iron field." This statement is followed by another statement that the possibilities of using neutral gases, therefore, are impractical. Would it not be well to expand on the basis for this conclusion? It is our opinion that the use of such gases would merely serve to displace air that would exist in the mold cavity prior to pouring. Such gases would be displaced by the metal during the pouring operation. The liquid metal then coming in contact with the mold wall would be subjected to whatever gas reactions or atmosphere were generated at the mold interface by reactions taking place in the sand itself. The possibilities of controlling mold interface reactions by the use of organic materials thus are not apparent simply on the basis of recognizing that normal cast metal practices produce penetration.

AUTHORS' REPLY: The most expedient method of answering this discussion seems to be in a tabular form with reference to each question individually.

1. The references on page 2 with respect to "oxidizing and reducing atmospheres" are qualitative and are based upon published results of other investigators, as noted in the paper.

2. The comments relating to pages 9 and 10 indicate a lack of understanding of the equilibrium data presented. Under carefully controlled conditions of an equilibrium investigation of the system Fe-Si-O-C, there is no question about the nature of the atmosphere. If the $CO_2/CO$ ratio permits a silicate melt to

form, it is oxidizing; if the $CO_2/CO$ ratio does not permit a silicate melt to form, it is reducing. This is clearly evident in Fig. 1. These conditions are obviously not those of actual shop practice, but we can use the equilibrium data to indicate how severely oxidizing an atmosphere exists during solidification and subsequent cooling. Based on this comparison we can say that under the conditions of this investigation the green sand mold produces a much more oxidizing atmosphere than the shell sand mold. Both, however, exhibit an oxidizing condition.

3. With respect to penetration Mr. Neff must define which type he means, metal or silicate melt penetration. It should be pointed out that the silicate melt is an $SiO_2$ saturated FeO liquid. The iron present is a result of Fe precipitation from this melt as it cools through the iron stable field during solidification.

4. The analysis of the gas mixture refers to a sample from the phase equilibria studies as noted in the text. These studies were performed in a closed vertical tube furnace using analyzed gas mixtures of $CO_2$ and CO. It was not obtained from any of the castings, also as noted in the text.

5. Quantitative metallography is quite an accepted technique in the field. The results of this investigation are presented as a qualitative evaluation of an extensive microstructural and macrostructural study.

6. The fact that green sand mold practice results in higher oxides of iron at the metal-mold interface, and that shell sand molds yield a lower state of oxidation at the interface indicate to these authors that any additions to the molding sand mixture that can produce a less oxidizing atmosphere (i.e., phenolic resins, organics) may reduce surface reaction. This is the basis for our conclusions.

**Feeding Distance of Bars in Investment Molds,** by H. PRESENT and H. ROSENTHAL *(paper printed beginning on page 138 of this volume).*

C. M. ADAMS, JR.[64] *(written discussion):* The authors are to be congratulated for their contribution to a field of casting technology almost entirely lacking in quantitative information. The development of new techniques and materials is extending the field of application of hot mold procedures, and much more of this type of research will be needed in order that they can be used to full advantage.

Anything which slows freezing should presumably slow the rate of interdendritic liquid flow needed to accomplish feeding, and thereby promote soundness. However, the authors' work indicates there may be a bigger effect on the nature rather than the incidence of shrinkage, since slow cooling also promotes mushy solidification. The conclusion that high pouring temperature and high mold temperature tend to disperse as much as to eliminate shrinkage, seems quite justified. The same reasoning applies to the effect of carbon content. Irrespective of thermal events, increasing carbon concentration also promotes mushy freezing, as well as lowering the melting poin . A previous study on centerline shrinkage in room temperature sand molds indicated that carbon had a profound influence on dispersing shrinkage and neither promoted nor interfered with feeding in tapered sections.[1]

Although the authors have addressed themselves to feeding distance and solidification in uniform sections, it is of interest that increasing the mold temperature greatly amplifies the effects of taper on directional freezing. Certain conclusions can be drawn from the fact that the freezing time of a round section depends on the square of its dimensions, and also upon the square of the temperature difference between the metal and the mold. For example, based on the way in which the freezing time of steel varies along a tapered section, a given amount of taper will be three times as effective in a mold at 2000 F as at 1200 F. And then, in a hot mold, there is always the possibility of thermal taper, variation of mold temperature along a uniform section. To take a second example, in a mold at about 2000 F, a mold temperature gradient of 10 F/in. along a 0.5 in. diameter bar is equivalent to geometric taper of 0.05 in./ft (and this in turn, is equivalent to over one in./ft at room temperature). Do the authors have any reason to feel that temperature gradients in the mold prior to casting could have had any influence on their observations?

1. E. J. Sullivan, C. M. Adams, Jr., and H. F. Taylor, "Transport of Feed Metal During Solidification of Tapered Steel Bar," AFS TRANSACTIONS, *vol. 65,* (1957).

In view of the profound influence of mold temperature on thermal events in solidification, there clearly will have to be a new set of rules developed for this phase of the foundry field, and the authors have made a valuable step in that direction.

E. M. BROAD[65] *(written discussion):* To the best of our knowledge, this paper is the first published manuscript relating feeding distance of bars in investment molds to certain significant parameters. It is our hope that this publication will provide the impetus for more scientific and technical work of this nature. The investment casting industry had been hampered by lack of good material on gating and risering, and if it is to continue its phenomenal growth pattern, it must recognize this deficiency.

Experience in our foundry indicates agreement with the findings presented. Apparent feeding distance increased with increased mold temperature, increased metal pouring temperature and increased carbon content. We do, moreover, recognize an important effect of thermal capacity of molding material and rate of heat dissipation. In this regard we do question the rather heavily loaded flask of especially the ¾-in. bars. It is our feeling that the rate of heat transfer, and this the thermal gradients and solidification pattern, would be markedly different between those bars on the outside of the cluster and those bars in the interior.

It is also well to bear in mind that the findings presented are based upon a radiographic sensitivity of 1.5 per cent. We believe that the sensitivity used has a bearing on the results obtained, especially in the evaluation of mold temperature. At a constant metal pouring temperature, an increase in mold temperature will reduce thermal gradients during solidification of the bar from the unrisered (or chilled) end to the risered (or hot) end. Any reduction of thermal gradients would tend to promote dispersed shrinkage at the expense of existing centerline shrinkage. We would, therefore, offer that the observed improvement in feeding distance with increased mold temperature could be, in effect, a finer dispersion of centerline shrinkage not detected at 1.5 per cent sensitivity.

Figure 2 shows that increase in apparent feeding distance as mold temperature is raised. As noted by the authors, raising mold temperature from 1200 F to 1600 F has a much smaller effect than when raising temperature from 1600 F to 2000 F. This is in accordance with theory, for if the heat transfer coefficient remains constant, then freezing time is inversely proportioned to the square of the difference between melting point and mold temperature. Therefore, a casting which freezes in 1.00 min when poured into a 1200 F mold, will take 1.95 min to solidify in a 1600 F mold, but 5.50 min to solidify in a 2000 F mold. The 5½ min increase in solidification time from 1200 F to 2000 F could cause much more dispersed shrinkage than the twofold increase from 1200 F to 1600 F. If this dispersed shrinkage were fine, then it might not have been detected at 1.5 per cent sensitivity, and results would be as indicated by Fig. 2. The authors do suggest undetected shrinkage as a possible partial explanation for their findings.

It is also possible that the extended feeding range is due to the greater heat transfer from the riser at the higher mold temperature. Perhaps microradiography would indicate the contributing effects of actual greater feeding range due to heat transfer and apparent greater feeding range due to finer dispersed shrinkage.

In regard to the increase in feeding distance attributed to higher metal pouring temperature, we believe this to be a true increase and not the result of finer dispersed shrinkage. Raising pouring temperature at a constant mold temperature will promote directional solidification with a decrease in both centerline and dispersed shrinkage.

One more point should be raised. Figure 2 is drawn to show the curves passing through the origin. It is true that at zero diameter, a zero length is fed, however a 1/16-in. section can be fed sound quite a distance. Some of our work indicated that while a ½-in. section of 303 stainless steel can be fed 4 times or 2 in., a 1/16-in. section can be fed 10 to 20 times or ⅜-in. to 1¼-in., depending upon metal and mold temperatures and gating arrangements. This would indicate more of a parabolic shape as the origin is approached. A closer look at Fig. 2 does show that the curves could have been drawn this way.

AUTHORS' REPLY: The authors wish to thank both Dr. Adams and Mr. Broad for their discussion of the paper. Their analysis adds greatly to the data presented.

Dr. Adams has raised the question as to possible temperature gradients in the mold prior to pouring. It is recognized that the outer part of the mold will cool fairly rapidly when the

mold is withdrawn from the preheating furnace. The thermal gradient so produced would promote an effect during pouring which would be similar to a chill at the end of the bar. To avoid this, the molds were withdrawn from the furnace immediately prior to pouring. In addition, the number of molds poured from one ladle were limited to three so that total pouring time could be held to a minimum.

Mr. Broad has called attention to the fact that bars on the perimeter of the cluster have a different thermal environment than those on the interior of the cluster. This type of cluster was used for the sake of economy. The data obtained did not include the identity of bars from these different positions. Thus it is not possible to determine what, if any, effect there was on feeding distance.

In regard to feeding distance of small diameter bars, such as $\frac{1}{16}$-in., we feel that the technique of x-ray examination for soundness is obviously unsuitable. The fine shrinkage one might find in thin bars would require detection by a technique such as microradiography. The whole question of the soundness and feeding of thin sections is one which requires special treatment outside the scope of this program.

**Specialty Melting as a Service Function in a Research Laboratory,** by W. F. Moore *(paper printed beginning on page 483 of this volume).*

Member: Is there any difference in surface contour from vertical vs. horizontal melting?

W. F. Moore: No.

C. Stanley:[66] Have you experimented with electron beam melting?

W. F. Moore: No, however, we are now setting up for this type of melting.

**Meeting the Challenge of Controlling Quality,** by L. Krueger *(paper printed beginning on page 549 of this volume).*

H. Gould:[67] Where are your quality control people drawn from? How many do you have?

L. Krueger: Our quality control people are drawn from the Pattern, Inspection and Foundry departments. We have three quality control people, excluding regular inspection personnel.

W. Giff:[68] What are the functions of the quality control committee?

L. Krueger: They serve in an advisory capacity, to establish and maintain quality control procedures. The committee also acts as a clearing house on quality control problems.

Member: Is the committee advisory only?

L. Krueger: No. The committee also has authority.

**High Integrity Steel Castings Foundry Procedure Development,** by R. J. Ely and J. E. Fries *(paper printed beginning on page 813 of this volume).*

J. T. Berry:[69] What is the kyanite material?

Authors' Reply: Kyanite is a complex aluminum silicate. When calcined it expands and forms a hollow grain which provides the insulating characteristics.

**Radiography of Cast Nuclear Components by means of Kilocurie Cobalt Sources,** by R. Lent *(paper printed beginning on page 526 of this volume).*

A. J. Kiesler:[70] What is the cost of a cobalt 60 source such as described here?

R. Lent: $30,000.

A. J. Kiesler: How many people present have either cobalt or x-ray in their shops?

R. Lent: By a count of the hands raised here, 7 out of the 30 present have one or the other as a shop tool.

**Betatron Use for Radiographic Examination of Heavy Section Castings,** by L. J. Venne *(paper printed beginning on page 743 of this volume).*

Member: What is the cost of equipment for use on 10 in. of steel?

L. J. Venne: About $300,000.

W. Koppi:[71] Did your company consider using cobalt 60?

L. J. Venne: Yes, but the company decided on the betatron because of a projected need for over 12 in. sections.

W. Koppi: Could you give an example of comparative exposure times for 10 - 12 in. sections for the betatron vs. cobalt 60?

L. J. Venne: Cobalt 60 — $1\frac{1}{3}$-million volts — 175 min. Betatron — 24 million volts — 5 min.

**Press Forging of High Alloy Steel,** by J. T. Berry, T. Watmough and P. R. Gouwens *(paper printed beginning on page 701 of this volume).*

H. Hailey:[30] How does this process compare with ausforming?

Authors' Reply: Where the matrix being deformed is austenitic, the process is comparable. However, with much of the work on both CF8C (347) type steels and 17-4 PH precipitation hardening steels, the formation and deformation during pressing of low carbon martensite (i.e., martenforming) is partly responsible for the strengthening.

W. Barber:[33] World formation of martensite have a detrimental effect on corrosion resistance?

Authors' Reply: Generally speaking the overall corrosion resistance of high chromium steels is a function more of chromium content (i.e. stability of oxide film) than metallographic structure. However, under certain specific corrosion conditions it could well be that the major metallographic phase present effects corrosion rate, e.g., pitting type corrosion in certain environments and possible certain aspects of stress corrosion. Specific cases would thus appear to warrant cursory study.

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